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PREFACE.

The present volume contains nineteen papers—six of which were issued under the auspices of the Joint Research Committees of this Institute and the British Iron and Steel Research Association, and one of which reported on an investigation carried out with the aid of a grant from the Andrew Carnegie Research Fund—that were presented at the Annual General Meeting of this Institute held in London on 11th and 12th July, 1945; the discussion and correspondence on them and the authors' replies are also included.

Section I. of this volume contains the above-mentioned matter and also Minutes of the Proceedings of the Meeting and the Report of Council and Statement of Accounts for 1944.

Section II. is devoted to a survey of the literature of the manufacture and properties of iron and steel, and kindred subjects, and consists of a collection of abstracts of articles from the Transactions and Proceedings of scientific societies and from the technical press. This section also contains reviews of recent books and bibliographies of literature dealing with the manufacture and properties of iron and steel. The matter included in this Section has already appeared in the *Bulletin of The Iron and Steel Institute*, which is issued monthly.

In front of the title-page are inserted lists of Bibliographies and Special Reports issued by the Institute, together with a list of Translations made available between 1st May and 30th November, 1945.

4, Grosvenor Gardens,
London, S.W. 1.
28th February, 1946.

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LIST OF PLATES.

Abbreviations and Symbols.

Å.	Ångstrom unit(s) = 1×10^{-10} m.
A.C.	air-cooled; alternating current.
A.H.	air-hardened.
amp.	ampere(s).
amp.hr.	ampere-hour(s).
approx.	approximately.
at.-%	atomic per cent.
at.wt.	atomic weight.
atm.	atmosphere(s) (pressure).
A.W.G.	American wire-gauge.
Bé.	Baumé (scale).
b.h.p.	brake horse-power.
B. & S.	Brown and Sharpe (gauge).
B.o.T.	Board of Trade.
b.p.	boiling point.
B.T.U.	Board of Trade unit(s).
B.Th.U.	British thermal unit(s).
B.W.G.	Birmingham wire-gauge.
C.	centigrade (scale).
cal.	calory (-ies).
c.c.	cubic centimetre(s).
c.d.	current density.
c.g.s.	centimetre-gramme-second unit(s).
cm.	centimetre(s).
coeff.	coefficient(s).
conc.	concentrated.
const.	constant(s).
cu.	cubic.
cwt.	hundredweight(s).
D.C.	direct current.
dia.	diameter.
dil.	dilute.
dm.	decimetre(s).
e.m.f.	electromotive force.
e.v.	electron volt(s).
F.	Fahrenheit (scale).
ft.	foot, feet.
ft.lb.	foot-pound(s).
g.	gramme(s).
gal.	gallon(s).
H.F.	high-frequency.
h.p.	horse-power.
h.p.hr.	horse-power-hour(s).
hr.	hour(s).
in.	inch(es).
in.lb.	inch-pound(s).
I.S.W.G.	Imperial standard wire-gauge.
K.	absolute temperature (Kelvin scale).
kg.	kilogramme(s).
kg.cal.	kilogramme-calory(-ies).
kg.m.	kilogramme-metre(s).
km.	kilometre(s).
kV.	kilovolt(s).
kVA.	kilovolt-ampere(s).
kW.	kilowatt(s).
kWh.	kilowatt-hour(s).
kX.	crystal Ångstrom(s) ≐ 1000 Siegbahn X-units.
lb.	pound(s).
L.F.	low-frequency.
M	molar (solution).
m.	metre(s).
m.amp.	milliampere(s).
max.	maximum.
mg.	milligramme(s).
min.	minimum; minute(s).
ml.	millilitre(s).
mm.	millimetre(s).
m.m.f.	magnetomotive force.
m.p.	melting point.
mV.	millivolt(s).
$m\mu$	millimicron = 1×10^{-9} m. = 10 Å.
N.	normal (solution).
N.T.P.	normal temperature and pressure.
O.H.	open-hearth; oil-hardened.
O.Q.	oil-quenched.
oz.	ounce(s).
p.d.	potential difference.
*p*H	hydrogen-ion concentration.
p.p.m.	parts per million.
r.p.m.	revolutions per minute.
sec.	second(s).
sp.gr.	specific gravity.
sq.	square.
S.W.G.	standard wire-gauge.
T.	tempered.
temp.	temperature.
V.	volt(s).
VA.	volt-ampere(s).
W.	watt(s).
Wh.	watt-hour(s).
W.G.	water-gauge.
W.Q.	water-quenched.
wt.	weight.
wt.-%	weight per cent.
yd.	yard(s).
γ	microgramme(s) = 1×10^{-6} g.
μ	micron(s) = 1×10^{-6} m.
$\mu\mu$	1 millionth micron = 1×10^{-12} m. = 0·01 Å.
Ω	ohm(s).
$^\circ$	degree (arc or temperature).
$'$	minute of arc; foot (feet).
$''$	second of arc; inch(es).
$<$	less than.
$>$	greater than.
$\nless$	not less than.
$\ngtr$	not greater than.
$\leqq$	equal to or less than.
$\geqq$	equal to or greater than.
$\neq$	not equal to.
$\equiv$	identically equal to.
$\cong$	approximately equal to.
$\propto$	proportional to.

SECTION I.

MINUTES OF PROCEEDINGS
AND PAPERS OF
THE IRON AND STEEL INSTITUTE.

ANNUAL GENERAL MEETING
1945

MINUTES OF PROCEEDINGS

AND

PAPERS AND DISCUSSIONS

AT THE

ANNUAL MEETING IN LONDON, 1945.

The Seventy-Sixth Annual Meeting of The Iron and Steel Institute was held at The Institution of Civil Engineers, Great George Street, London, S.W. 1, on Wednesday and Thursday, 11th and 12th July, 1945. The Wednesday sessions, with the President (Mr. Arthur Dorman) in the Chair, started at 10.30 a.m. and 2.30 p.m.; at the Thursday session, which commenced at 9.45 a.m., the Senior Vice-President (Dr. C. H. Desch, F.R.S.) presided in the President's unavoidable absence. The Institute of British Foundrymen had been invited to take part in the proceedings.

The Minutes of the previous Meeting, held in London on 23rd and 24th November, 1944, were taken as read and signed.

Obituary.

The President (Mr. Arthur Dorman): I regret to have to announce the death of two Past-Presidents. One is Mr. W. R. Lysaght, C.B.E., who died on 27th April. We all knew him as a Grand Old Man of the industry and a great friend of many of us. The other is Sir William Ellis, G.B.E., who died a few days ago. Both did yeoman work for our great industry.

At the request of the President, the Members stood in silence for a few moments.

Welcome to Members and Visitors.

The President: I should like to offer a cordial welcome to all the members and visitors here to-day. We are very glad to know that there are a number of Americans present, and we also offer a warm welcome to Dr. Magnus Tigerschiöld, Chief Engineer of Jernkontoret, who has also attended our Council meeting.

Presentation of the Report of Council and Statement of Accounts for 1944.

The President: I have now to present the Report of Council for 1944 (*see* pp. 19 p–35 p), and there are several comments which I should like to make.

Staff.—It is right that some reference should be made in the Report to the staff. Mr. Elsdon, our Librarian, has completed 40 years' service with the Institute; Miss Davison, the Assistant Librarian, and Miss Dowd, our Accountant, have been with us for 26 and 25 years, and Mr. Chattin has been with us for 20 years. It is gratifying that our staff stay with us

in this way, and I should like to thank them all, not excluding the Secretary, who has done so much work for the Institute.

Membership.—At the end of last year the total membership was 3034. It is now 3240, an increase of over 200 since 1st January. That is very gratifying, and shows that interest in the Institute is sustained.

Co-operation with the Institute of Metals.—I am pleased to say that the friendly co-operation which has so long characterized our relations with the Institute of Metals has continued. On the retirement of the Secretary of that Institute at the end of June, 1944, the two Institutes agreed to have the same Secretary. The two Councils also agreed that as from 1st January of this year a number of the officials should be joint, and that Mr. Chattin should serve as Assistant Secretary of both Institutes and Mr. Vaughan, the Assistant Editor of the Institute of Metals, should help The Iron and Steel Institute in the same way. Both Councils believe that these arrangements will work well. We are doing a great deal in other ways to secure the closer working together of the two Institutes, and I think that this is as it should be; we have common interests and a thousand Joint Members.

Local Societies.—Many joint meetings have been held with local societies, and we thank all those who have made this possible. A new development of considerable potential importance is the affiliation which has been arranged with the Lincolnshire Iron and Steel Institute, the Sheffield Society of Engineers and Metallurgists and the Staffordshire Iron and Steel Institute. We expect that similar affiliation will be arranged with other societies. Your Council feel that some part of Members' subscriptions should be spent in the districts in which they live and work, and that it is much better to make use of well-established local societies than to compete with them by forming local branches. We are very glad to work with these local societies, which have their own traditions and local enthusiasms, and some of which are older than this Institute. We think that they ought to remain independent, and we believe that they have done much useful work. By affiliation we believe that we can help each other.

Proposed Institution of Metallurgists.—Some months ago, a preliminary notice was sent to all Members informing them of the steps which the Council have taken, jointly with the Council of the Institute of Metals, to encourage metallurgists to form a professional institution. I understand that the new institution will be registered very soon, and that applications for membership will then be invited. Joint membership rates have been agreed, and this should be helpful. We are sure that this new body will work in the closest harmony with the Institute of Metals and ourselves, and that there will be no rivalry or overlapping; indeed, we think that by encouraging its formation we have done something of real value for our professional members. We wish the new Institution of Metallurgists every success.

National Certificates in Metallurgy.—Not unconnected with the foregoing is the recent formation of a Joint Committee for National Certificates in Metallurgy, on which, in addition to representatives of our Institute, there sit representatives of the Ministry of Education, the Institution of Mining and Metallurgy and the Institute of Metals. A full explanatory memorandum is being issued to all Members. We hope that the principals of technical colleges in all suitable districts will encourage their pupils to join courses, and that employers will give practical encouragement to young men on their staffs to work for these certificates. This is an important step in spreading knowledge of the science on which our industry is based.

Research.—I have left to the last reference to the new British Iron and Steel Research Association. This Association has been formed by the British Iron and Steel Federation to take over responsibility for co-operative

research, and the Federation has agreed to set aside a substantial annual sum for its work. The Institute nominates nine Members of the Council of the new Association. It is proposed that our Joint Research Committees, which have done so much during recent years for research, shall continue, and that each shall be responsible to the appropriate Divisional Panel of the Association. We have also made it clear that the Institute will be glad to give all possible other help to the Research Association, notably by making available our library and information services and by publishing reports. This is an important development, and I am sure that we all wish every success to the new Research Association.

The Hon. Treasurer (The Hon. R. G. Lyttelton): I beg to present the Accounts for the year 1944; they are in the usual form (*see* pp. 36 P–43 P). The Balance Sheet needs little comment. The Council have continued their policy of increasing reserves against the expenditure which is likely to be required after the war. The income and expenditure both reflect the increased membership. Thus, revenue from subscriptions has reached a new record at £7576, an increase of £500 over last year's figure; on the other hand, we have had to spend more on services, publishing, stationery and so on. The final result is, however, satisfactory; after transferring £1625 to Reserve and Suspense Accounts, or £550 more than last year, the excess of income over expenditure transferred to the Balance Sheet has increased from £215 to £571. Both the Williams Prize and the Carnegie Research Trust Funds have again accumulated substantial sums, and this will enable useful increases in awards to be made in the near future.

I wish again to record our sincere thanks to all companies from whom we have received special subscriptions, and especially to those who have agreed to continue these payments for another three years. With this help we can look forward without too much anxiety to the next few years, during which we shall have to incur considerably increased expenditure in order to carry through projects intended to provide improved service, but during which also we hope to put the finances of the Institute on a permanent basis. Our reserves are by no means too large, and we require a good deal of working capital, but by the efforts of the Secretary and his staff we are getting our finances into very good shape.

The President then moved the adoption of the Report of Council and Statement of Accounts for 1944, and the motion was carried unanimously.

Presentation of a Portrait of Mr. James Henderson.

The President: I wish to inform you that yesterday evening a portrait of Mr. James Henderson, Past-President, by Captain Oswald Birley, M.C., R.P., was presented to the Institute by Sir Walter Benton Jones, Bt., on behalf of the Board of the United Steel Companies, Ltd. (*Applause.*) Sir Walter, in making the presentation, referred to the great services that Mr. Henderson had rendered to the industry and to the Institute during a period of over fifty years, in the course of which he had held almost every possible position of influence and authority and lent distinction to them all; it was a pleasure to his Company (Sir Walter added) to be able to make the presentation to the Institute. He also paid a tribute to Mrs. Henderson and thanked and congratulated the artist. I am confident that the Members generally will cordially endorse these views. (*Applause.*)

Presentation of the Bessemer Gold Medal to Mr. Harold Wright.

The President : The next item on the programme is the presentation of the Bessemer Gold Medal to Mr. Harold Wright, the chief metallurgist of Messrs. Dorman, Long & Co., Ltd. Having known him for forty years, it is a great pleasure to me to be in the Chair at a time when he is to receive this honour, the greatest that the Institute can bestow.

Mr. Wright started at the North Eastern Steelworks 61 years ago, and worked with Mr. C. H. Ridsdale, and also with Mr. E. H. Saniter, who ultimately himself was awarded the Bessemer Gold Medal and who was then Mr. Ridsdale's assistant. Mr. Saniter left shortly afterwards, and Mr. Wright took on the position of assistant in the laboratory there. He also attended classes with another Bessemer Gold Medallist, Dr. J. E. Stead. Dr. Stead's name is known throughout the world, but he was especially loved and respected in Middlesbrough, which was his home.

Mr. Wright then went to Samuelsons, and worked on Sir Bernhard Samuelson's blast-furnaces, and that is where I first met him. My first meeting with him was 40 years ago, when I was a very young man, and we were struggling at our Britannia Steelworks, using molten Cleveland iron in the basic open-hearth process. The furnaces were too small and the iron was what might be called patchy. Mr. Wright, who was then the chief chemist at Samuelsons, was very helpful to us in our difficulties, and he induced Mr. Samuelson to improve the quality of the iron by additions first of all of Swedish ore and then of other ores, with the result that in due course we obtained quite a respectable basic iron and made much better progress with this then more or less new molten-metal process. In those days we used hardly any scrap; only iron oxide was put in the furnace, and the molten iron was poured in, and the reactions, as you may imagine, were considerable.

Mr. Wright's chief activity after that was in connection with by-product coke-ovens. Samuelsons were one of the very first to instal by-product coke-ovens. He saw the waste of a million cubic feet of gas a day, and he at once made suggestions for utilizing it. The town was consulted, but the town authorities—I am talking now of the early years of this century—were a little suspicious, and difficulties were possibly magnified. In 1913, however, Middlesbrough became the first municipality to utilize coke-oven gas. I have no hesitation in saying that that was entirely due to the sustained efforts of Mr. Wright. In 1920 he read a paper before this Institute, of which he had been a Member since 1902, on the chemical and thermal side of blast-furnace practice. He has written a number of other papers, including one on the application of coke-oven gas to towns, read before the Society of Chemical Industry in Newcastle in 1923.

In 1918 he became chief metallurgist to Dorman Longs. At first he gave his attention chiefly to coke-ovens and blast-furnaces, but we soon switched him over to the steel side as well. He was very helpful in assisting the steel furnace managers in making the special steel required for the Sydney Harbour Bridge a number of years ago. He is a Past-President of the Cleveland Institution of Engineers and the Cleveland Technical Institution.

The outstanding feature of Mr. Wright's career is that he has been a real and true practical friend of any manager in trouble. Even to-day, though no longer so young as he was, he will lend a helping hand to any manager whose blast-furnace or steel furnace is causing him a little trouble. When they are in trouble they at once call for Mr. Wright. Mr. Wright goes there, and does not say a great deal for a few hours, but in due course

the furnace is nursed back to its normal health. His practical experience is, of course, enormous.

Like so many good men, Mr. Wright has had a great deal of help from his wife. Mrs. Wright is a very wonderful woman; she is an M.B.E., she works for the Women's Voluntary Service, and she is no mean speaker on a political platform. They have a large family, two of whom are with Dorman Longs, while one, Mr. Leslie Wright, who is here to-day, is Managing Director of Messrs. Gjers Mills & Co., Ltd. One of the causes of Mr. Wright's success has been, I think, the happiness of his home life. For a long period he has served well the firm with which he is connected, and also the trade as a whole, and he takes a worthy place in the list of famous men who have had the Bessemer Gold Medal bestowed upon them.

The President then, amid applause, handed a replica of the Medal to Mr. Wright.

Mr. Harold Wright: I should like to thank the President for the very kind words he has used in handing me this Medal. I appreciate it fully as the greatest distinction that can come to a metallurgist, especially in the iron and steel trade, and I am also deeply impressed by the fact that my name is to be included in the list of those very eminent men to whom this Medal has been awarded in the past. It is a very great honour that you have done me to-day.

As the President has said, I have spent the whole of my life in solving those urgent, practical problems that the works managers meet with in their daily life—almost in their hourly life. Starting many years ago, for ten years at least at the North Eastern Steelworks I was personally responsible for investigating every defect that occurred in connection with the steel. That is very important work, which has to be carried on at all steelworks. I must not be understood as saying anything which would seem to minimize the importance of long-term research, but this work which I have been doing has been, and will continue to be, essential to the progress of the industry.

Since I first started in the steel trade, there has been a wonderful improvement in the quality of what I can best describe as commercial steel. It was in the experimental stages in the early days, and was, I think, looked on rather dubiously, but to-day it stands on an entirely different plane, and can be relied on to be a very good, sound material.

My work has to a large extent kept me from being associated with other activities. Before the present war, my firm was producing 30,000 tons of steel ingots a week, which brings a few problems to people like myself, and we have collieries, mines, sheet works and constructional activities, including bridge-building, all of which have furnished problems, so that we have a tremendous amount of work to do to solve the problems which arise every day. In all my work I have been assisted by colleagues, and have had the help of the firm's large staff of very excellent analysts and other scientists, who have helped me to a very great degree. In addition to that, I have received the greatest help from Mr. Arthur Dorman and from the other directors and the executive. I have been helped by the firm in every new undertaking that I have brought forward. In thanking you, therefore, for the great honour which you have done me to-day, I feel that I must say that it is due in part to the help that I have received, and all those whom I have mentioned have a share in the great honour which has been conferred upon me.

I feel less embarrassed than might otherwise be the case because the applause of the Members has been interpreted by me as suggesting that they also approve of my receiving this Medal. I thank you all. My

knowledge of basic English is not at all equal to my knowledge of basic steel, and I may have failed to find suitable words to express my feelings, but I want you to realize that I do feel that this is a great honour that has been done to me. It is appreciated in the same way by my wife, and I thank the President for his kind reference to her and to my family. (*Applause.*)

The PRESIDENT: I should like to mention that Messrs. Dorman, Long & Co., Ltd., have endowed a lecture called the Harold Wright Lecture. Dr. C. H. Desch, F.R.S., gave the first lecture last autumn. It is perhaps unique for a lecture to be endowed in this way in the lifetime of the man whom it is intended to honour. We hope that Mr. Wright will be spared to hear many more Harold Wright Lectures given by eminent people.

WILLIAMS PRIZES.

The PRESIDENT: I have to announce that the Council have awarded the Williams Prize for 1944 to Mr. G. D. Elliot, of the Appleby-Frodingham Steel Co., Ltd., for Special Report No. 30, "Ironmaking at the Appleby-Frodingham Works of the United Steel Companies, Limited." We shall all remember that excellent Report, and the very good discussion to which it gave rise. Mr. Elliot richly deserves this Prize, and I have great pleasure in handing him the certificate.

The presentation was then made, amid applause.

The PRESIDENT: The Council decided to-day to award another Williams Prize. This award goes to Mr. R. W. Evans, of Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd., for his paper on "The Heating of Open-Hearth Furnaces with Mixed Coke-Oven and Blast-Furnace Gas," which was printed in the *Journal* for 1944, No. I., pp. 395 P–417 P. This also is a first-class paper of a thoroughly practical character, and aroused much interesting discussion. Mr. Evans is not here to-day, so I shall hope to present his Prize to him at the Autumn Meeting later this year.

CHANGES ON THE COUNCIL.

The PRESIDENT: At the meeting of the Council this morning we decided to nominate as President for next year Dr. C. H. Desch, F.R.S. His name will therefore be submitted to you at the Autumn Meeting. (*Applause.*) We all know Dr. Desch, and the Council feel that we shall be fortunate to have him as our President. I see by your applause that you agree with us.

We also appointed Sir Peter Brown an Honorary Vice-President, in recognition of his long and distinguished service to the Institute. We do not want to lose Sir Peter's help in our activities, but he feels that now he is retiring from active business he ought to make room for a younger man. His election as an Honorary Vice-President affords an excellent solution. I hope that he will continue to come to our Council meetings. We should like to congratulate him on his knighthood, and to congratulate Sir Allan Macdiarmid on the same honour.

We have elected our old friend Professor Thomas Turner an Honorary Member of the Institute. It is interesting to note that it is sixty years this year since Professor Turner read his first paper, on the effect of silicon on cast iron. It was printed in the *Journal of the Chemical Society* in 1885. He became engaged to his wife at the same time, and married two years later; they have "lived happily ever after," and we wish them the best of luck.

The SECRETARY (Mr. K. Headlam-Morley) said that in accordance with Bye-law No. 10, the names of the following Vice-Presidents and Members of Council were announced at the Autumn Meeting, 1944, as being due to retire by rotation at the present Meeting; they were eligible for re-election:

Vice-Presidents: Dr. C. H. Desch, F.R.S., Captain H. Leighton-Davies, C.B.E., and Mr. J. S. Hollings, C.B.E.

Members of Council: Mr. D. R. Lysaght, Mr. N. H. Rollason, Mr. E. J. Fox, Mr. J. Sinclair Kerr and Mr. J. Mitchell.

No other Members having been nominated up to one month previous to the present Annual Meeting, they were now presented for re-election. (*Agreed.*)

The following were now Honorary Members of Council in their capacity as Presidents of the Societies mentioned:

Mr. C. A. Julius Behrendt (Lincolnshire Iron and Steel Institute), in succession to Mr. J. N. Kilby.

Captain H. Leighton-Davies, C.B.E. (Swansea and District Metallurgical Society), in succession to Mr. O. J. Thomas.

Mr. A. A. Munro (Sheffield Metallurgical Association), in succession to Mr. R. Staton.

BALLOT FOR THE ELECTION OF MEMBERS AND ASSOCIATES.

Mr. W. GEARY (Scunthorpe, Lincs.) and Mr. L. RIPLEY, O.B.E. (Middlesbrough), were appointed scrutineers of the ballot for the election of Members and Associates, and in due course reported that the following 147 Members and 96 Associates had been elected.

MEMBERS.

Anderson, Arthur Leslie (Bilston, Staffs.). **Angus,** Harold Thomas, Ph.D. (Birmingham). **Auld,** Duncan John (Haverton Hill, Co. Durham). **Bailey,** George Leo, M.Sc. (Boxmoor, Herts.). **Banks,** Allan Pearson (Workington). **Barnard,** George Henry (London). **Beauchamp,** Herbert Reginald (London). **Belcher,** Ronald, F.R.I.C. (Sheffield). **Bell,** Leonard Percy (Shotton, Chester). **Bird,** Albert Charles (Haverton Hill, Co. Durham). **Birks,** H. (Saltburn-by-the-Sea). **Bond,** Alfred Ernest (Wannerton, Kidderminster). **Briggs,** Janet Zaph (New York, U.S.A.). **Buckley,** John Hunter, B.Sc. (Melbourne, Australia). **Burdon,** Andrew Murray, Ph.D. (Eng.), (London). **Cameron,** R. C. W. (Saltburn-by-the Sea). **Capper,** Horace (London). **Catton,** Alfred Hebert (Leeds). **Catton,** Douglas E. (Leeds). **Challoner,** Charles (Stockton-on-Tees). **Clark,** J. (Saltburn-by-the-Sea). **Collin,** Kenneth Lovell (Sheffield). **Colton,** William James (Sheffield). **Connor,** Francis George (London). **Cooke,** William (Stoke-on-Trent). **Craig,** William (Sheffield). **Cutbush,** Eric George, B.Sc.(Hons. Met.) (Luton). **Dadswell,** Cyril John, Ph.D. (Sheffield). **Davies,** Reginald Ewart (Port Talbot). **Derge,** Gerhard, Ph.D. (Pittsburgh, Pa., U.S.A.). **Desai,** N. L., B.Sc. (Burnpur, India). **Dews,** Joseph Walter (Willenhall, Staffs.). **Dixon,** Edgar Ogle (Cudahy, Wisconsin, U.S.A.). **Dixon,** Frederick Cecil (Karabuk, Turkey). **Donaldson,** Archibald Ian (Sheffield). **Drogseth,** Arne Gulbrand (London). **Eaton,** Wyman (Chicago, U.S.A.). **Fairbairn,** Joyce, B.Sc. (Sheffield). **Findlay,** Frank Brown, A.R.T.C. (Glasgow). **Fitzjohn,** Robert (Wellingborough). **Fox,** Gordon (Chicago, U.S.A.). **Gandhi,** Prakash Nagardas, B.Sc. (Ishapore, Bengal, India). **Gibson,** Stanley (Workington). **Gilbert,** Francis Charles, A.R.C.S., M.Sc. (Hartlepool). **Goodeve,** Charles Frederick, O.B.E., D.Sc., F.R.S. (London). **Gordon,** Lt.-Col. Lord Dudley Gladstone, D.S.O. (Sheffield). **Grange,** H. (Saltburn-by-the-Sea). **Grant,** B. E. (Saltburn-by-the-Sea). **Gray,** Alexander Reith (Shotton, Chester). **Haines,** George William (Birmingham). **Halupka,** Herman Augustyn (London). **Hanson,** John Harry Arthur (Coventry). **Haut,**

F. J. G. (London). **Heasman,** Noel, B.Sc., A.R.C.S. (Hartlepool). **Hill,** Erle G. (Wheeling, West Virginia, U.S.A.). **Hockings,** Reginald Arthur (Altrincham). **Hodgson,** Matthew Mitchell (Wallsend-on-Tyne). **Holden,** Alfred Harold, B.Sc. (Chesterfield). **Holden,** Athole Frederick (Melton Mowbray). **Holman,** Charles (Bristol). **Holt,** Eric Maximillian (Sheffield). **Hoult,** Frederick Herbert (Strood, Kent). **Howie,** Robert Charles, B.Sc. (Cardiff). **Hue de la Colombe,** Jean (Denain, France). **Humphries,** Harold (Sheffield). **Hunt,** John P. (Sheffield). **Hurst,** Thomas William (South Bank, Yorks.). **Ihre,** Nils E. (London). **James,** Reginald Sidney (Bilston, Staffs.). **Kenneford,** Arthur Spencer, M.Sc. (Loughborough, Leics.). **Kilpatrick,** William, M.I.M.E. (Sheffield). **Knight,** Harold, B.Sc. (Hons.) (Scunthorpe, Lincs.). **Ko,** Tsun (Birmingham). **Kohli,** Bhim Sen, B.Sc. (Hons.) (Jamshedpur, India). **Lajus,** Henry Jean (Enfield). **Lee,** Oscar Martin (Portsmouth). **Levy,** Felix Lewis (London). **Li,** Pao-Sen, B.Sc. (Birmingham). **McCollum,** William Stuart (Chesterfield). **McGregor,** Douglas S., M.Sc. (Huddersfield). **McGuire,** Francis T., Ph.D. (Chicago, U.S.A.). **McHarg,** William (Bilston, Staffs.). **Mackay,** Murray (Dagenham, Essex). **Mackenzie,** Murdo (London). **McLaren,** John (Corby). **Mardon,** Henry Herbert, B.Sc. (Scunthorpe, Lincs.). **Mehta,** Yeshwant Muljibhai, B.Sc. (Bombay, India). **Meikle,** George, B.Sc. (London). **Menon,** Pallath Madhava, B.Sc. (Calcutta, India). **Mitchell,** James Ferguson (Bilston, Staffs.). **Morfey,** Frederick Henry (London). **Murty,** Duvvuri Suryanarayana, B.A., M.Sc. (Ishapore, Bengal, India). **Myers,** Denis Wharmby (Oldham). **Nicholls,** Christopher (Wallsend-on-Tyne). **Nicholls,** Herbert (Shirley, Surrey). **Nichols,** Lawrence Edgar (London). **Noakes,** John Edmund (Queensferry, Chester). **Parakh,** Ardeshir Kaikhusru, L.E.E., L.M.E. (Hons.) (Jamshedpur, India). **Parish,** Charles Henry (West Bromwich, Staffs.). **Payne,** Charles Arnold, B.Sc. (Hons.) (Derby). **Pendred,** V. (Saltburn-by-the-Sea). **Phimister,** Donald (Bilston, Staffs.). **Phipps,** Frederick George (Sheffield). **Po,** M. H., B.Sc. (Birmingham). **Podgorski,** Stefan J. B. (London). **Pople,** Arthur Haydn (Cardiff). **Pratt,** Ronald Stanley (Manchester). **Prichard,** Basil Stennett (Wednesbury). **Raine,** Thomas, Ph.D. (Manchester). **Ramsay,** James Kennedy (Bromley, Kent). **Rice,** Owen R. (Chicago, U.S.A.). **Richards,** Robert Layland, B.Met.Eng. (Melbourne, Australia). **Robinson,** N. W. F. (Saltburn-by-the-Sea). **Rogers,** Thomas Charles (Abadan, Iran). **Rollinson,** Colin Vincent, B.Sc. (Nilgiris, South India). **Royce,** Howard Mark William (London). **Rudberg,** Erik Gustaf, Ph.D. (Stockholm, Sweden). **Ruffle,** Thomas William (Kingston, Surrey). **Rustay,** Arnold L. (Worcester, Mass., U.S.A.). **St. Claire-Johnson,** Garry Henry George (Manchester). **Saunders,** William Robert (Birmingham). **Seaver,** Jay John (Chicago, U.S.A.). **Sentance,** Henry G. V. (Glasgow). **Sieveking,** H. A., M.Sc. (Eng.), (Glasgow). **Smith,** Alfred James Nicol, B.Sc. (Hons.) (London). **Smith,** Peter A. (Halifax). **Stant,** Leonard, B.Sc.Tech. (Wolverhampton). **Stanton,** J. (Muradnagar, Meerut District, India). **Stanton,** Leonard Radford, Ph.D. (Manchester). **Stanworth,** Stephen, A.M.I.Mech.E. (Burnley). **Stokes,** Alexander John (Lymington, Hants.). **Stokowiec,** Zygmunt (Sheffield). **Summers,** Richard Felix (Shotton, Chester). **Sweet,** John M., B.Sc. (Ishapore, Bengal, India). **Sylwestrowicz,** Witold (London). **Tomlinson,** Arthur (London). **Tranter,** Frederick John (Pinner, Middlesex). **Vorster,** J. H., B.Sc. (Vereeniging, South Africa). **Wainwright,** Clifford, M.Sc. (Teddington, Middlesex). **Warier,** T. K., B.Sc.(Met.) (Burnpur, India). **Warriner,** C. (Saltburn-by-the-Sea). **Webb,** Leslie Wilfrid Thomas (Swansea). **Weeks,** George Wentworth (Hatfield, Herts.). **Wells,** Geoffrey Weston (Scunthorpe, Lincs.). **Wheeler,** Marcus Alan (Derby). **Wiles,** John Roland (Coventry). **Williams,** L. R., B.Sc. (London).

Associates.

Bagchi, Asoke, B.Sc. (Calcutta, India). **Baker,** Peter (Doncaster). **Ball,** Charles Selwyn (Letchworth). **Bateman,** Jack (Leeds). **Bayly,** Jean Rosemary (Leeds). **Beach,** Norman (Chelmsford). **Bills,** Kenneth Maurice (Darlaston). **Birkhead,** Thomas Kilner (Stocksbridge, Sheffield). **Brain,** Alan Gordon (Cambridge). **Brigham,** William Burton (Leeds). **Buckley,** Alice Brenda (Leeds). **Bullock,** Gordon (Durham). **Burton,** Donald (Scunthorpe, Lincs.). **Carroll,** John Bernard (Manchester). **Churchman,** Arthur Trevor (Birmingham). **Cina,** Bernard (Glasgow). **Clements,** Peter Graham (Cambridge). **Clough,** Graham (Leeds). **Cockaday,** R. E. (Leeds). **Colley,** Frederick Arthur David (Wolver-

hampton). **Connolly,** Brian Jolliffe (Sheffield). **Corbitt,** Reginald (Willenhall, Staffs.). **Cribb,** Robert James Preston (Cambridge). **Cush,** James J. (jun.) (Saltburn-by-the-Sea). **Davies,** Don Eurof (Swansea). **Dawson,** David Whitfield Outram (Leeds). **Dearden,** Melvin (Sheffield). **Dennison,** John Philip (Leeds). **Dutt,** Nikhil Chandra (Calcutta, India). **Earley,** Cyril Charles (Birmingham). **Elson,** Geoffrey Percy (Manchester). **Fairbank,** Laurence Haydn (Birmingham). **Forsyth,** Peter Joseph Edward (Farnborough, Hants.). **Ghosh,** Debabrata (Calcutta, India). **Gilmour,** James McMillan (Glasgow). **Goult,** Jean M. (Leeds). **Grover,** Sydney F. (Farnborough, Hants.). **Hamilton,** Ian Gordon (Glasgow). **Hannam,** John Charles (Staines, Middlesex). **Harper,** Sydney (Birmingham). **Hoyle,** Geoffrey (Leeds). **Hughes,** Ian Charles Herbert (Swansea). **Jackson,** Brian (Leeds). **Jackson,** Brian Richard (Corby). **Johnson,** Charles Stanley (Bradford). **Krishnamurthy,** B. (Trichinopoly, India). **Laidler,** Douglas Stuart, Ph.D., B.Sc. (Watford). **Lang,** Cameron (Sheffield). **Larn,** John David (Kingswood, Bristol). **Lechem,** Sonia (Cambridge). **Leja,** Jan (Birmingham). **Lewis,** Michael Frederick Paul (Scunthorpe, Lincs.). **Lund,** Harry, B.Sc. (Sheffield). **Lyons,** John Vianney (Birmingham). **Mantle,** Bryan Sutton (Darlaston). **Martin,** Donald Luther (Schenectady, New York, U.S.A.). **Mason,** Colin Frank (Leeds). **Millward,** Ronald (Leeds). **Mobey,** Frank, B.Sc. (Hons.) (Manchester). **Muir,** Archibald Robert (Sheffield). **O'Sullivan,** Ronald (Newport, Mon.). **Pallot,** Lawrence Ernest (Poyle, Colnbrook, Bucks.). **Parkinson,** Edgar Harold (Wednesfield, Staffs.). **Parr,** J. G. (Leeds). **Peplow,** Douglas Boraston (Bilston, Staffs.). **Raine,** Edward Morgan (Scunthorpe, Lincs.). **Reynard,** John Kingsley (Burnley). **Richardson,** James George (Scunthorpe, Lincs.). **Robertson,** Andrew Dick (Glasgow). **Roy,** Saradindu, B.Sc. (Birbhum, Bengal, India). **Sarkar,** Satyabrata, B.Sc. (Calcutta, India). **Simpson,** Philip Ross (Leeds). **Smith,** George Allan (Scunthorpe, Lincs.). **Snelson,** Donald H. (Birmingham). **Stearn,** Francis Michael (Leeds). **Stephenson,** Norman (Sheffield). **Subramaniam,** V. (Calcutta, India). **Swindale,** John Douglas (Birmingham). **Taylor,** William Bower (Durham). **Thomas,** Ivor John (Netherton, Dudley). **Thompson,** Laurence Stephen (Wolverhampton). **Thorneycroft,** Dennis Raymond (Birmingham). **Thornton,** Jonathan (Scunthorpe, Lincs.). **Tomlinson,** Jeffrey Ernest (Warrington). **Tull,** E. V. (Leeds). **Vaughan,** Thomas Bernard (Birmingham). **Wain,** Richard Arthur (Scunthorpe, Lincs.). **Waters,** Gordon Frank (Wolverhampton). **Welsh,** Norman Clifford (Newcastle-on-Tyne). **Wilkes,** Thomas Orlando George (Darlaston). **Williams,** T. L. (Farnborough, Hants.). **Willsmere,** Lawrence Allan (Kidderminster). **Wilson,** Cecil Vincent (Sheffield). **Wilson,** Robert William (Cambridge). **Workman,** George Monkhouse (Workington). **Worth,** Derek (Leeds).

Presentation of Papers.

A list of the papers included in the programme of the Meeting is given on p. 10 P. The following papers were presented for verbal discussion.

Wednesday, 11th July :

Morning Session :

"The Application of Radiography to the Improvement of Foundry Technique," by R. Jackson. (Paper No. 13/1945 of the Steel Castings Research Committee.)

"First Report of the Foundry Practice Sub-Committee." (Paper No. 12/1945 of the Steel Castings Research Committee.)

Afternoon Session :

"The Influence of Centrifugal Casting upon the Structure and Properties of Steel," by L. Northcott and D. McLean.

"Examination of Two Ingots of Free-Cutting Steel, One Containing Lead and the Other Lead-Free," by C. S. Graham. (Paper No. 31/1945 of the Committee on the Heterogeneity of Steel Ingots.)

"The Microscopical Examination of Samples of Lead-Bearing

Complete List of Papers Presented at the Annual General Meeting in London, 1945.

"First Report of the Foundry Practice Sub-Committee." (Paper No. 12/1945 of the Steel Castings Research Committee.)

W. E. Bardgett and R. E. Lismer: "Mode of Occurrence of Lead in Lead-Bearing Steels and the Mechanism of the Exudation Test." (Paper No. 33/1945 of the Committee on the Heterogeneity of Steel Ingots.)

M. Balicki: "A Study of Work-Hardening and Re-annealing of Iron." (Andrew Carnegie Research Report.)

R. J. Box and B. A. Middleton: "Corrosion of Steel Salt-Bath Pots by Molten Alkali Nitrates."

A. H. Cottrell: "Tensile Properties of Unstable Austenite and Its Low-Temperature Decomposition Products."

The late Dr.-Ing. Leopold Frommer and A. Murray: "The Influence of the Heat Treatment of Steel on the Damping Capacity at Low Stresses."

C. S. Graham: "Examination of Two Ingots of Free-Cutting Steel, One Containing Lead and the Other Lead-Free." (Paper No. 31/1945 of the Committee on the Heterogeneity of Steel Ingots.)

P. Hamer, L. Powell and E. W. Colbeck: "Emulsions of Oil in Water as Corrosion Inhibitors."

C. C. Hodgson and H. O. Waring: "A Note on the Relationship between Preliminary Heat Treatment and Response to Nitriding of some Nitriding Steels."

J. E. Hurst and R. V. Riley: "A Further Note on the Microstructure of High-Silicon Acid-Resisting Iron."

R. Jackson: "The Application of Radiography to the Improvement of Foundry Technique." (Paper No. 13/1945 of the Steel Castings Research Committee.)

A. H. Jay: "A Study of the Basic Open-Hearth Process, with Particular Reference to Slag Constitution."

L. Northcott and D. McLean: "The Influence of Centrifugal Casting upon the Structure and Properties of Steel."

J. R. Rait and H. J. Goldschmidt: "The Constitution of Basic Steel Furnace Slags."

T. H. Schofield: "The Microscopical Examination of Samples of Lead-Bearing and Lead-Free Steels and Ingot Irons." (Paper No. 32/1945 of the Committee on the Heterogeneity of Steel Ingots.)

F. C. Thompson and L. R. Stanton: "Some Observations on the Austempering and Isothermal Transformation of Steels, with Special Reference to the Production of Martensite." (Paper No. 16/1944 of the Alloy Steels Research Committee.)

K. Winterton: "A Note on the Physical Properties of an Austenitic Weld-Metal and Its Structural Transformation on Straining."

K. Winterton: "The Effect of Overheating on the Transformation Characteristics of a Nickel–Chromium–Molybdenum Steel."

Y. K. Zea: "The Phosphorus Reaction in Basic Open-Hearth Practice."

and Lead-Free Steels and Ingot Irons," by T. H. SCHOFIELD. (Paper No. 32/1945 of the Committee on the Heterogeneity of Steel Ingots.)

"Mode of Occurrence of Lead in Lead-Bearing Steels and the Mechanism of the Exudation Test," by W. E. BARDGETT and R. E. LISMER. (Paper No. 33/1945 of the Committee on the Heterogeneity of Steel Ingots.)

The latter three papers were discussed jointly.

Thursday, 12th July :

Morning Session :

"The Constitution of Basic Steel Furnace Slags," by J. R. RAIT and H. J. GOLDSCHMIDT.

"The Phosphorus Reaction in Basic Open-Hearth Practice," by Y. K. ZEA.

"A Study of the Basic Open-Hearth Process, with Particular Reference to Slag Constitution," by A. H. JAY.

The above three papers were discussed jointly.

LUNCHEON.

Following the Annual Meeting a Luncheon was held at the Connaught Rooms, Great Queen Street, Kingsway, on the 12th July, 1945, at 1 P.M. The PRESIDENT (Mr. Arthur Dorman) occupied the Chair, and among the four hundred members and guests present were the following: His Excellency M. A. J. Clasen, Luxemburg Minister, The Rt. Hon. the Earl of Dudley, M.C., Lieut.-Col. Sir John Greenly, K.C.M.G., C.B.E., The Rt. Hon. Viscount Greenwood of Holbourne, P.C., Mr. W. J. Jordan, High Commissioner for New Zealand, Mr. Kenneth G. Lampson, Deputy Controller, Iron and Steel Supplies, His Excellency M. Réné Massigli, French Ambassador, Mr. G. Heaton Nicholls, High Commissioner for the Union of South Africa, His Excellency M. Bjorn G. Prytz, Swedish Minister, Mr. Spencer Summers, Secretary, Department of Overseas Trade, Mr. M. Tigerschiöld, Chief Engineer, Jernkontoret, Stockholm, and Mr. C. R. Wheeler, Controller, Iron and Steel Supplies.

The loyal toast having been honoured, Mr. SPENCER SUMMERS (Secretary, Department of Overseas Trade) proposed the toast of

"*The Iron and Steel Institute and Industries.*"

He said: It is with the utmost pleasure that I rise to propose this toast. During the course of the last few weeks I have had occasion to make a good many speeches, as one who is anxious to return to Westminster. They have been very much alike, and from force of habit I have tended to get into a groove; if, therefore, during the course of my remarks to-day I inadvertently use some of the phrases which I have been in the habit of employing, I ask for your indulgence; and if I end with a powerful peroration culminating in an earnest request for your support, I hope you will put it down to habit and not to abuse of your excellent hospitality.

The brief remarks which I propose to make I wish to divide into two parts: those which concern the work of my Department and those which concern your industry. Since I have had the privilege of serving the Department of Overseas Trade I have met many people who have travelled the world, both before and during the war, and who have had occasion to make contact with our officers abroad and learn something of the service which we are anxious to render. We are in fact a "service" department; and I say without fear of contradiction that the overseas officers of His

Majesty, whether in the Trade Commissioner service or in foreign countries, have a service to render to the business community when travelling which they can ill afford to ignore. Those who have seen fit to use them have given a good report of the result; but I find that there are too many who do not consider it worth their while to make contact with our overseas officers. I would ask that those who have occasion to travel should do what they can, when they get to their destination, to make use of the help which our officers can give, and which is offered—and this should make a particular appeal to Scotsmen—free of charge.

We hear much in these days of the relationship which should exist between Government and industry. We are anxious to put at the disposal of industry as much information as we are able to give, and we hope that industry will keep in the closest touch with the Government and let us know what they are doing and what their plans and ambitions are, so that we can evolve an effective working partnership based on mutual confidence, which alone can provide a satisfactory basis for the relationship between Government and industry. It is the ambition of His Majesty's Government to be entitled by our actions to enjoy the confidence of industry. If you have plans, I hope you will not hesitate to let us know them, though if you prefer to wait until 26th July, I shall not quarrel with that!

During the past few years there have been produced a number of reports on the cotton trade, the coal trade and other trades which have suggested, and quite rightly, the need for a substantial expenditure of capital. It is only to be expected that after six years of war, during which capital expenditure on modernisation and replacement was virtually impossible, great programmes of modernisation and expansion should be desired. But one result of those reports, and a result which I for one deplore, is the impression which they give that we are lagging behind in the economic race, and that private enterprise has failed.

I hope that that impression will not overlay the undoubted fact of the immense contribution which your industry has made not only to the welfare of this country but to the winning of the war in Europe. During that time you have produced for that war 72,000,000 tons of steel. You have collected from divers places (including the square in which I live, where I am glad to see the end of the railings!) some 5,000,000 tons of scrap. In various other ways you have made a notable contribution. Tool steels used to need 18% of tungsten, which came from Burma. When those supplies were lost to us, you found ways and means to replace them which enabled you to provide the tool steel for the munitions industries of this country without which our programme could not have been carried out. During that time you have had to use some strange raw materials, as is well known to His Excellency the Swedish Minister, who is sitting on my right. You have frequently had to employ plants to make products for which they were not designed. You have had transport difficulties and fuel difficulties. You have had to contend with A.R.P. and to find men for the Home Guard. All these things have thrown on the management, on the foremen and on all those whose daily life is spent in the works a burden which they have borne manfully, and it is fitting on this occasion to pay a great tribute to them for the work which they have done.

Your industry has passed through difficult times. When you acquired protection, you were told to put your house in order, and one way in which you did that was to more than double your output in the six years preceding the war. Where would this country have been if your output had not been doubled in those years? During that period, while you were engaged in putting your house in order, much time was spent on organization, and particularly on the selling side of your commercial activities. I hope that when this war is over a greater emphasis than was then possible will be

placed upon the reduction of cost rather than upon the commercial organization and the selling and all the rest of it; and in that sphere The Iron and Steel Institute has a notable contribution to make.

One effect of the war has been to stimulate the production of goods and services abroad in markets to which you and others were accustomed to send your goods. That development is inevitable; it will go on when this war is over, and signs are not lacking in the Dominions, in India, in the Argentine and in South America generally which it would be fatal to ignore. It is no use trying to keep back the tide. It is with satisfaction that I, from the knowledge I have of the many friends that I have in this room, am able to say that you recognize the proper ambitions of those countries that have been consuming countries in the past to work out their own ideas and to develop their own industrial future. I do not believe that it is contrary to the interests of Britain that the markets overseas should become industrialized. It may entail changes in the form of the exports which we have sent from this country in the past, but the raising of standards of living and the demands which industrialization brings with it cannot in the long run do other than redound to the advantage of this country.

Those ambitions and those desires to industrialize will require the help of the older industrialized countries, and in particular of this country and of the United States of America. I see here to-day several people from the U.S.A., and we welcome them very heartily. (*Applause.*) I am sure they will not object or misunderstand me when I say that we do not take kindly here to the plans of distant manufacturers to go to the United States for business purposes and to come to our island for a little relaxation! (*Laughter.*) We in this country pride ourselves on our modesty, and we are not good at blowing our own trumpet; but I hope that our modesty will not react against us, for I am sure that the accomplishments of Britain, particularly in the field of science and applied science, have not as yet had proper recognition in the minds of the peoples of the world. (*Applause.*)

I would therefore ask you to bear with me for a moment while I recall to you some of the directions in which Britain has made a notable contribution in the field of science, both for the winning of the war and for raising the standard of living in the days that are ahead. Take aircraft. The Mosquito carries the same bomb-load as the Flying Fortress, at a hundred miles an hour faster, and with a crew of two instead of a crew of ten. The Lancaster carries a greater bomb load than any other existing aircraft, and is the only one capable of carrying a 10-ton bomb. Those are British developments. Then look at the engines. Where will you find an engine comparable to the Rolls Royce Merlin, the power of which has been doubled in five years? President Roosevelt, whose death all of us in this country deplored, paid a notable tribute to those who had made it possible for the British sparking plug for aircraft to be used in most American machines, as the result of the development of the material called Sintox, without which it would scarcely have been possible.

The electrical industry has an equally fine record. The Battle of Britain might well have been lost if we had not had the warning which radiolocation gave us, and no country can claim greater credit than Britain for the development of radiolocation. The use of that invention in aeroplanes was made possible only by the part which Imperial Chemical Industries played in the development of polythene, an insulating material. Many of you will remember in the early days of the war the announcements in the Press about the magnetic mine, and later on how grave was the submarine menace, which might well have starved us out during the course of the war. It is to British science and research and invention that we owe the means to combat the magnetic mine and the submarine. The

same story can be told of bomb sights and predictors. In plastics we have Perspex instead of glass.

If we turn to medicine, where publicity has already been rightly given to Penicillin and D.D.T., we find the same story. It has been possible to impregnate the shirts of the troops, particularly in the East, to guard them against lice and the diseases which they carry. Last summer I was privileged to go to Australia and New Zealand, and I spent a few days in New Guinea. In order that we might not be subjected to malaria, we were dosed for weeks beforehand and for weeks afterwards with one yellow pill per day of Atebrin, which is the substitute which the British medical profession discovered to take the place of quinine, the supplies of which had been cut off by the Japanese. That interesting discovery had, we were then told, reduced the casualties from malaria to one-hundredth of what they had formerly been.

The cotton industry has produced a waterproof material which excludes liquid but admits air. The woollen industry is said to have overcome the problem of shrinkage. The tins of food which we take with us on picnics will in future have the means of heating them up contained in the bottom of the tin. There has been invented a means of blowing up the back tyre of a cycle as it goes along. Last, but by no means least, on this limited list, to which many more things could be added and will be added when security considerations no longer apply, we have the products in which your industry has played such a notable part, and in which many of you have played an individual part—Mulberry and Pluto and Fido. If I may single one out for especial mention, it is to the steel industry above all others that the credit is due for the production of the Bailey bridge, the tolerances and accuracies of which surpass the normal, and which on the Italian front was used as frequently as one per mile of the ground covered.

I mention these things merely to emphasize the importance which we should attach to our own accomplishments during this war, and with no intention of saying anything to the detriment of our Allies; I wish to show that our modesty is no criterion of our accomplishments. I have no use for those people who see good in every country but their own. We are too prone to self-disparagement. We have done great things. We have made a marked contribution to the civilization of the world, and we must make sure that our modesty does not react against us. Your industry has made a notable contribution. There remains much to be done, but you have the brains, you have the ingenuity, you have the resources, and I hope that you will be provided with the Government which will give you the scope to develop those resources which you have at your disposal! We ask merely for your collaboration and a willingness to accept Government as an aid to your own efforts. (*Applause.*)

The President (Mr. Arthur Dorman), who responded, said: We are all very gratified by the speech which Mr. Spencer Summers has just made. This is the first public function which The Iron and Steel Institute has held since 1941. A year before that, in 1940, one of our guests was the late Sir William Bragg, whose name will always be honoured in the scientific world. I remember one of his remarks. He defined science—not an easy thing to do—as "the knowledge of Nature's materials and laws." Those of us who are in the iron and steel trade feel, I think, that that description is a very adequate one. We have to have a knowledge at any rate of Nature's materials, and sometimes we wish that we were more richly endowed with them.

We in the iron and steel trade have no reason to be ashamed of the part that we have played in the war. Recently the Chairmen of individual

concerns have been allowed for the first time to make public some of our activities. I do not think that we feel inclined, in the presence of such an assembly as this, to boast unduly about them, but surely we should take adequate credit for them. A great many experiments were made, many of which were successful. Special qualities of alloys were necessary for aeroplanes and other uses, special armour plate had to be produced, special forgings were wanted for tanks, and every ship that sailed the seas and every weapon of war had bestowed on it much highly scientific research. Mr. Spencer Summers has mentioned the Bailey bridge and floating piers and undersea pipe-lines. We must all admire the tremendous enterprise and skill which they entailed, but I suggest that it is the accuracy of the workmanship which made those projects so successful. The Bailey bridges in particular had to be most scientifically and accurately made, and many firms in this country successfully took a share in the good work.

We have had our troubles. We started the war cut off from all our rich foreign ore, and we had to substitute the leaner local ironstone. I think we must say that the local ironstone producers came up to scratch well by increasing their output to an extent which was not believed possible a few years before. The use of the local material, however, involved difficulties for us. Working under black-out conditions was very unpleasant for the workmen, and also for some of the management. The coke-ovens were unpleasant places to work in. What was worst of all was the impossibility of doing the ordinary maintenance. We had to leave modernization entirely on one side, and were behindhand with ordinary maintenance for reasons which were beyond our control.

I am glad to say that the position is now altering, and as soon as labour is available (which will not be for some short time yet) we in the trade are determined to embark on large new schemes, costing many millions of pounds. New works will be built, and we fully intend to increase the output of steel in this country above anything that we have heretofore achieved. That means, of course, that some of the old works will have to be scrapped, but it is our firm intention to expand and not to retreat.

If I may now say a word about our Institute, we have always recognized the necessity for research work. We are 75 years old, and we have a record of which we need not be ashamed. We have international connections, and I should like to say how pleased I am to see here to-day representatives from the United States of America, among whom I should like to mention Professor Bradley Stoughton. We are also very pleased to have with us His Excellency the French Ambassador, M. Massigli, the Ministers of Sweden and Luxemburg, and the High Commissioners of South Africa and New Zealand. (*Applause.*) Yesterday the Director-General of Supplies of South Africa gave us some very interesting details of the steel trade in that country, and no one knows more than he does about it. I am sorry that he cannot be here to-day. We welcome many other guests, and it is a great compliment to the Institute that they should be present. We want to go forward in co-operation with our competitive neighbours and friends from overseas. (*Applause.*)

There has recently been formed the British Iron and Steel Research Association, and our Institute is naturally working in the closest co-operation with it. Great determination and enterprise are essential, and I believe that the organization of our trade in the future will exhibit these qualities. The result will be an industry which will hold its own in the competitive world which in due course we shall have to face. After six years of starvation, we must not be led astray if at the moment there is a great influx of inquiries and orders; that is only natural; but, taking a long view, we must make our industry competitive. We hope that the coal trade will settle its differences and allow us to obtain our coal at a price which

will make it possible, given efficient workmanship and efficient management, to compete in the markets of the world. Efficiency and enterprise are clearly the goals at which we must aim.

I should like to mention two small but important ways in which we hope to achieve this end. One is the proposed Institution of Metallurgists, which is in process of being formed, with the idea of giving metallurgists the same status in the world as engineers and others. The other is that, in conjunction with the Ministry of Education, the Institution of Mining and Metallurgy and the Institute of Metals, the Presidents of which we are pleased to welcome here to-day, we have formed a Joint Committee to award National Certificates to young students in metallurgy. We have to try to give some inducement to the younger people to study and give their quota to the future development of our great industry.

Before I conclude, I wish to refer to the presence here of my friend Professor Turner. It is sixty years this year since he read his first paper—and I am told it was a very good paper—on the influence of silicon on cast iron. (*Applause.*) Two years later he married, and he and his wife are still "living happily ever after." (*Applause.*) I must also mention my friend Mr. Harold Wright, who yesterday received the Bessemer Gold Medal, the greatest honour that the Institute can bestow. (*Applause.*)

It gives me great pleasure to propose the toast of

"*The Guests.*"

I have already mentioned some of them by name. We are very pleased to see them here, and I am glad to be able to leave the response to the toast in the able hands of my friend and colleague Lord Greenwood. (*Applause.*)

The Rt. Hon. The Viscount Greenwood, P.C., who responded, said: It is naturally a great honour to respond to the toast of the Guests before an audience almost as large as the San Francisco Conference. The guests are numerous and distinguished, headed by His Excellency the French Ambassador (*applause*), the Ministers of Sweden and Luxemburg, the High Commissioners of South Africa and New Zealand, and representatives of many of the Allies and of the Services of the King. Do not let us ever forget what we owe to the Services of our King. (*Applause.*)

I agree with every word said by Mr. Spencer Summers. He is not a Member of Parliament at the moment; for the moment the House of Commons is sealed up in tins, and not a word comes from it. I hope that Mr. Spencer Summers will return to the House. He is Secretary of the Department of Overseas Trade, and as far back as 1919 I occupied that post myself. With him, representing the third generation of a famous steel family, at the head of that Department, we have every reason to congratulate ourselves on the position there.

We wish to give an especial welcome to our American friends, and for this reason. We lost by the death of the late great President Roosevelt the first and best friend that this country and Empire ever had. (*Applause.*) Long before America came into the war he helped us in every way he could, and especially by using the services of the Prime Minister of Canada, Mr. Mackenzie King. Only the War Cabinet know how great was the threat to our existence at that time, and we should never forget that the man who stood by us first and to the end was that great President, who put the cause of civilization ahead of the politics of his own country. (*Applause.*)

I am speaking to the scientific branch of the steel industry. You scientists are unpopular with many people, but not with the industry. Scientists always irritate the complacent, spur the lazy, disturb the dogmatic and show up the inefficient. No wonder you are unpopular with many people! But I support you, because we do not want the complacent,

the lazy, the dogmatic or inefficient. As far as the steel industry is concerned, as Chairman of Dorman Long I can say that we lean on the people who advise us in the construction that is pending of new layouts on a big scale; without your help we could not start on them, let alone bring them to success.

My favourite scientist has always been Darwin, and I will tell you why. About the time when Professor Turner was writing his first paper, on the effect of silicon on cast iron, I was teaching in Sunday school in a small town in Canada; and I can assure those who have not done it that the most difficult task in a man's life as a teacher or speaker is to explain to a class of boys and girls the story of the Garden of Eden and the sinful possibilities of the fig leaf and the apple. Happily, Darwin washed that out, and generations of Sunday school teachers regard him in consequence as the most helpful scientist that there has ever been. (*Laughter.*)

To come to our own great interest, I agree with Mr. Spencer Summers that it is a negation of the truth to suggest that we are a decadent or a complacent body of people. The plans had been prepared for years and are nearing fruition to make our production and distribution of steel as efficient as those of any country in the world, but they were held back owing to the war. When the war started, it was thought that we should be doing our maximum if we could produce 10,000,000 tons of ingots a year, but we got up to nearly 14,000,000 tons, and we never failed to meet every demand that the Services made upon us. When it comes to Mulberry and other great constructions, it was the steel trade under its own leaders and with its own foremen and craftsmen who made those things possible, and on the due date, which is always a critical matter in all military affairs. (*Applause.*)

I wish your Institute well, and I wish the steel industry well, because I believe—and I speak as a veteran now—that politics divide but commerce unites. Whatever squabbles we may have with our Allies—I am not interested in our enemies, though I suppose as an old Sunday school teacher I ought to pray for them, in a whisper at any rate (*laughter*)—and whatever politicians may have to say to friendly countries abroad, let us in the commercial world work with our old friends such as France and Sweden and Luxemburg and Belgium and America and the Dominions and others; because I believe that a prosperous world working well together is the best guarantee for peace. (*Applause.*)

I wish the members of the Institute well in continuing the splendid help which they are giving to our industry, to our country and to the civilization of mankind. (*Applause.*)

REPORT OF COUNCIL FOR 1944.

THE Council submit this, their Annual Report and Statement of Accounts for the year 1944, to Members for their approval at the Seventy-Sixth Annual General Meeting of The Iron and Steel Institute. Information up to 31st March, 1945, has been included in some sections of the Report.

ROLL OF THE INSTITUTE.

Membership at the end of the year totalled 3034, an increase of 234 over the record figure of the previous year. Particulars are given in Tables I. and II. and in Fig. 1.

The current Roll of Members does not include those who have been placed on suspense lists. These consisted of 53 Members and Associates absent on military service, &c., and 249 Members and Associates resident in occupied territory or in neutral countries on the Continent.

TABLE I.—*Membership at* 31*st December*, 1944, *and Preceding Five Years.*

	31/12/'39.	31/12/'40.	31/12/'41.[1]	31/12/'42.	31/12/'43.	31/12/'44.
Patron	1	1	1	1	1	1
Hon. Members	15	14	15	14	12	13
Life Members	68	67	60	60	64	65
Ord. Members: [2]						
Home	1622	1673	1698	1753	1883	2017
Overseas	785	663	515	460	471	465
Associates	213	248	262	295	369	473
Total	2704	2666	2551	2583	2800	3034

[1] Revised.

[2] The above figures include six Members for 1942, seven for 1943 and eight for the other years whose names were retained in the List of Members in an honorary capacity by order of the Council.

TABLE II.—*Composition of Membership at* 31*st December*, 1944.

	Home Members.		Overseas Members.		Total.	
	Ordinary.	Joint.[1]	Ordinary.	Joint.[1]	Ordinary.	Joint.[1]
Patron	1	...	...	...	1	...
Hon. Members	4	2	6	1	10	3
Life Members	37	7	17	4	54	11
Ord. Members	1483 [2]	533	335	131	1818	664
Total (Members)	1525	542	358	136	1883	678
Associates	182	256	16	19	198	275
Total (classified)	1707	798	374	155	2081	953
Total Membership					3034	

[1] Members who are also Members of the Institute of Metals.

[2] Includes eight Members whose names were retained in the List of Members in an honorary capacity by order of the Council.

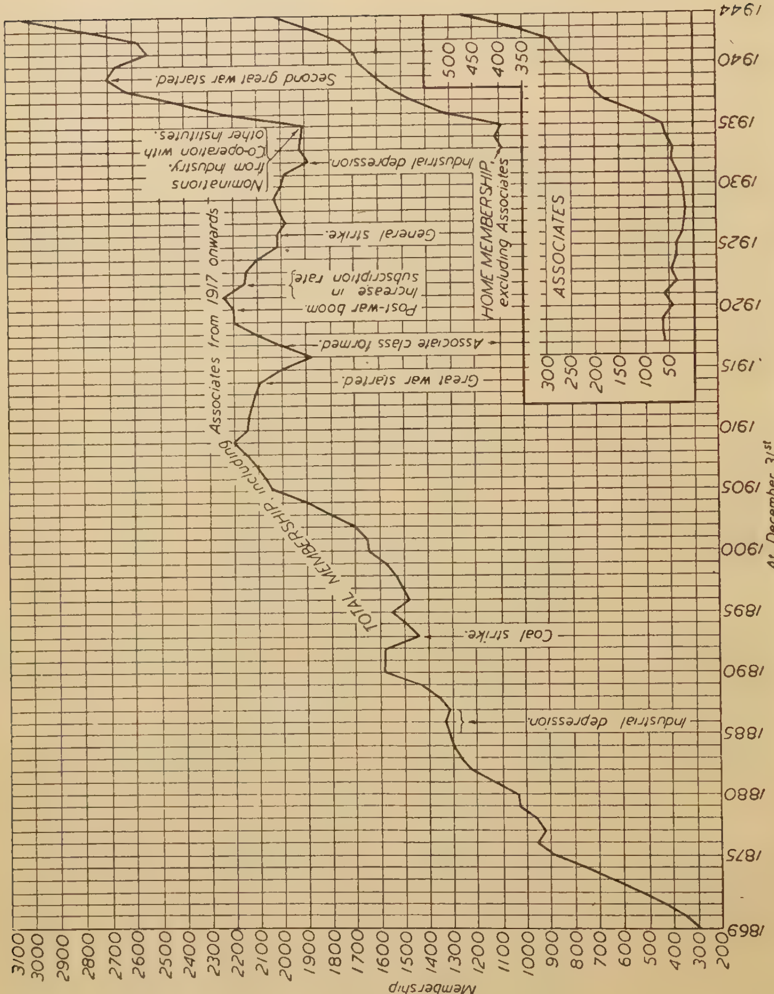

Fig. 1.—Variation in Membership since the Foundation of the Institute in 1869.

Twenty-six deaths were reported during the year and 26 Members resigned. Two hundred and eleven Members and 152 Associates were newly elected or reinstated.

OBITUARY.

The Council regret to record the deaths of the following twenty-four Members which occurred during the year 1944:

AIREDALE OF GLEDHOW, The Right Hon. Lord (London) (*Hon. Vice-President*) . . .	11th March.
ARMITAGE, R. (Leeds)	2nd February.
BELL, Lieut.-Col. Sir MAURICE H. L., Bt., C.M.G. (Northallerton) (*Hon. Vice-President*) . .	18th November.
BEYNON, Sir JOHN WYNDHAM, Bt., C.B.E., D.L. (Newport, Mon.)	13th October.
CARTWRIGHT, A. (Bombay)	25th November.
CRAMER, C. M. (Pittsburgh, Pa., U.S.A.) . .	19th April.
CUTHILL, J. (Stockport)	14th September.
DAVIES, H. A. (Wednesbury)	December.
DAYSON, W. (Doncaster)	22nd April.
DOUGLAS, H. (Rotherham)	?
FORNANDER, A. M. (Fagersta, Sweden) . .	18th October.
GRIFFITHS, P. (Gorseinon, Glam.) . . .	12th August.
HAWKE, R. S. (London)	26th July.
KEEGAN, C. C. (Dublin)	20th January.
KUJUNDZIC, N. (Leeds)	4th March.
LAW, S. W. (Corby)	18th June.
McINTOSH, F. F. (Sewickley, Pa., U.S.A.) . .	April.
MILROY, A. G. (Feltham, Middlesex) . . .	October.
SAKLATWALLA, Dr.-Ing. B. D. (Pittsburgh, Pa., U.S.A.)	4th November.
SANDERS, A. (Avonmouth)	24th July.
SCARF, F. (West Bromwich)	29th March.
STANSFIELD, Professor A. (Montreal, Canada) .	5th February.
SWINDEN, Dr. T. (Stocksbridge, near Sheffield) (*Vice-President*)	27th October.
WESTMAN, G. B. (London)	15th April.

The deaths of the following two Members took place earlier than 1944, but were not previously reported:

MICHAEL, R. G. (Swansea).	1942.
RUSSELL, D. (Rotherham)	30th May, 1943.

Lord Airedale of Gledhow and Sir Maurice Bell were Honorary Vice-Presidents, and Dr. T. Swinden was a Vice-President. Professor Stansfield joined the Institute in 1899, Mr. Armitage in 1898 and Lord Airedale in 1889. Obituary notices will be found in the No. II. volume of the *Journal* for the year 1944.

HATFIELD MEMORIAL LECTURE FUND.

A Hatfield Memorial Lecture Fund was raised by subscriptions from the iron and steel industry and individuals. At the request of the subscribers the University of Sheffield consented to act as Trustees. A Lecture Committee, consisting of representatives of The Royal Society, the University of Sheffield and the Institute, was formed, and it is proposed that an annual Memorial Lecture shall be held.

FINANCE.

(*The Statement of Accounts for* 1944 *is attached to this Report.*)

General Fund Balance Sheet.—The Balance Sheet is presented in the usual form.

General Fund Income and Expenditure Account.—Receipts from subscriptions and sales of publications were higher than the record figures of the previous year. After allocation to reserve and suspense accounts, the excess of income over expenditure transferred to the Balance Sheet was £571.

Trust Funds.—The financial position of the Trust Funds was well maintained.

Investments.—The value of the investments of the General and Trust Funds at the end of the year was £79,190 (an increase of £4885), or £11,408 in excess of their cost at which they are taken into the Balance Sheets.

House Fund and Industrial Subscriptions.—Income from special subscriptions was £4712. The Council acknowledge with thanks receipt of new subscriptions from The Brymbo Steel Co., Ltd., Guest, Keen and Nettlefolds, Ltd., The Kennedy Press, Ltd., The Kettering Iron and Steel Co., Ltd., Lamberton & Co., Ltd., Samuel Osborn & Co., Ltd., and Skinningrove Iron Co., Ltd., and wish to take this opportunity of again expressing their thanks to the following subscribers: Edgar Allen & Co., Ltd.; Ashmore, Benson, Pease & Co., Ltd.; Babcock and Wilcox, Ltd.; Bairds and Scottish Steel, Ltd.; Baldwins, Ltd.; Arthur Balfour & Co., Ltd.; Frederick Braby & Co., Ltd.; Bradley and Foster, Ltd.; The Briton Ferry Steel Co., Ltd.; Burnell & Co., Ltd.; The Butterley Co., Ltd.; Bynea Steel Works, Ltd.; Colvilles, Ltd.; Consett Iron Co., Ltd.; The Darlington Forge, Ltd.; The Darwen and Mostyn Iron Co., Ltd.; Darwins, Ltd.; Dorman, Long & Co., Ltd.; English Steel Corporation, Ltd.; Thos. Firth and John Brown, Ltd.; General Refractories, Ltd.; Gillette Industries, Ltd.; Guest Keen Baldwins Iron and Steel Co., Ltd.; J. J. Habershon & Sons, Ltd.; Hadfields, Ltd.; N. Hingley & Sons, Ltd.; William Jessop & Sons, Ltd.; Kayser, Ellison & Co., Ltd.; The Lancashire Steel Corporation, Ltd.; Arthur Lee & Sons, Ltd.; The Llanelly Steel Co. (1907), Ltd.; John Lysaght, Ltd.; McCall & Co. (Sheffield), Ltd.; The Millom and Askam Hematite Iron Co., Ltd.; The Mond Nickel Co., Ltd.; Neepsend Steel and Tool Corporation, Ltd.; Newton Chambers & Co., Ltd.; The Oughtibridge Silica Firebrick Co., Ltd.; The Park Gate Iron and Steel Co., Ltd.; The Patent Shaft and Axletree Co., Ltd.; Raine & Co., Ltd.; Round Oak Steel Works, Ltd.; Simon-Carves, Ltd.; Walter Somers, Ltd.; The South African Iron and Steel Industrial Corporation, Ltd.; South Durham Steel and Iron Co., Ltd.; South Wales Siemens Steel Association; The Stanton Ironworks Co., Ltd.; The Steetley Co., Ltd.; John G. Stein & Co., Ltd.; Stewarts and Lloyds, Ltd.; John Summers & Sons, Ltd.; Tata, Ltd.; Taylor Bros. & Co., Ltd.; Richard Thomas & Co., Ltd.; The Tinsley Rolling Mills Co., Ltd.; The Union Steel Corporation (of South Africa), Ltd.; The United Steel Companies, Ltd.; The Upper Forest and Worcester Steel and Tinplate Works, Ltd.; Vickers, Ltd.; The Wellman Smith Owen Engineering Corporation, Ltd.; The Welsh Plate and Sheet Manufacturers' Association; Whitehead Iron and Steel Co., Ltd.; The Woodall-Duckham Vertical Retort and Oven Construction Co. (1920), Ltd.

The Council wish also to record their thanks to Broken Hill Proprietary Co., Ltd., Davy and United Engineering Co., Ltd., and Partridge Jones and John Paton, Ltd., from whom subscriptions have been received since the end of the year, as well as to those Companies which have agreed to continue their subscriptions for a further period.

Changes on the Council.

(*To* 31*st March*, 1945.)

During the year Sir Arthur B. Winder, J.P., was nominated an Honorary Vice-President, and Mr. J. R. Menzies-Wilson a Vice-President. Mr. James Mitchell (Messrs. Stewarts and Lloyds, Ltd.), Mr. H. H. Burton (English Steel Corporation, Ltd.) and Mr. Gerald Steel (The United Steel Companies, Ltd.) were elected Members of Council. Dr. J. E. Hurst and Mr. A. A. Munro accepted invitations to become Honorary Members of Council during their presidencies of the Staffordshire Iron and Steel Institute and the Sheffield Metallurgical Association in succession to Mr. B. Thomas and Mr. R. Staton, respectively.

In accordance with Bye-Law No. 10, the names of the following Vice-Presidents and Members of Council were announced at the Autumn Meeting, 1944, as being due to retire at the Annual Meeting in 1945 :

Vice-Presidents.—Dr. C. H. Desch, F.R.S., Captain H. Leighton-Davies and Mr. J. S. Hollings, C.B.E.

Members of Council.—Mr. D. R. Lysaght, Mr. N. H. Rollason, Mr. E. J. Fox, Mr. J. Sinclair Kerr and Mr. J. Mitchell.

No other Members having been nominated up to one month previous to the Annual Meeting, the retiring Members are presented for re-election.

Bessemer Gold Medal.

The Bessemer Gold Medal for 1944 was awarded to Mr. Essington Lewis, C.H., Director-General of Munitions and Aircraft Production, Australia, in recognition of his outstanding services to the iron and steel industry of Australia.

The Bessemer Gold Medal for 1945 has been awarded to Mr. Harold E. Wright, Chief Metallurgist of Messrs. Dorman, Long & Co., Ltd., in recognition of the valuable contributions which he had made over many years to improving the technique of iron and steel manufacture.

Andrew Carnegie Medal.

No award of the Andrew Carnegie Medal was made during the year.

Williams Prize.

The Williams Prize for 1944 has been awarded to Mr. G. D. Elliot, of the Appleby-Frodingham Steel Co., Ltd., for Special Report No. 30, "Ironmaking at the Appleby-Frodingham Works of The United Steel Companies, Limited."

Andrew Carnegie Research Scholarship.

No grant was made by the Council during 1944.

The Worshipful Company of Blacksmiths.

Mr. Norman Iles was recommended for admission to the Worshipful Company of Blacksmiths.

Presentations.

The Council record their thanks to the following, among other donors : Members of the Joint Research Committees for the presentation of a portrait of the late Dr. W. H. Hatfield, F.R.S., Vice-President, by James Bateman, R.A.; Mr. R. A. Kirkby for a water-colour painting of the

Institute's building, by Henry Rushbury, R.A.; and Mr. Guy Barrett for a miniature anvil made by Mr. Chambers, head roller, from the first ingot made in the Bessemer shop at Ebbw Vale, together with photographs taken in 1906 on the dismantling of the original Bessemer shop.

The Council also thank those who have presented books to the Joint Library, including the American Iron and Steel Institute for the presentation of "Iron and Steel Industry: Position, Descriptions and Classifications" in 23 volumes.

Education.

A Joint Committee on Education has been formed with the Institution of Mining and Metallurgy and The Institute of Metals. Evidence was given before the Higher Technological Committee of the Board of Education.

National Certificates in Metallurgy.

A Joint Committee for National Certificates in Metallurgy was formed between the Institute, the Institution of Mining and Metallurgy, The Institute of Metals and the Ministry of Education, as a result of which it has been agreed to establish a scheme for National Certificates in Metallurgy. It is proposed that courses shall begin in September, 1945, and the first examinations for the Ordinary Certificates be held in 1946.

Proposed Institution of Metallurgists.

The Council, jointly with the Council of The Institute of Metals, decided to take steps to enable metallurgists to form a professional Institution. As a result, leading metallurgists on the two Councils, in collaboration with others, have formed a body which is representative of different interests and districts throughout Great Britain. After due consideration, these metallurgists have decided to form a professional Institution, to be known as the "Institution of Metallurgists." The Council look forward to working in close collaboration with this new Institution and extend their sincere good wishes for a prosperous future.

Joint Library and Information Department.

Joint Library.

The work of the Joint Library with The Institute of Metals was well maintained. The Council wish to thank those from whom presentations were received. A list of additions made to the library is issued quarterly and copies will be sent to Members on request. Demands for the loan of books and periodicals again showed an increase; 6953 publications were borrowed in 1944, as compared with 6430 in 1943.

Micro-Film Service.

The service for the supply of micro-films in conjunction with ASLIB and the Science Library, South Kensington, continued in operation. A reading instrument is available in the Joint Library. Members in possession of micro-film copies of foreign technical journals are invited to deposit them in the Library and to give permission to make them available to others wishing to inspect them.

Collaboration with the Institution of Civil Engineers and the Science Library.

The valuable collections of scientific works included in the Science Library and the Library of The Institution of Civil Engineers are available for consultation or loan under certain conditions.

Information Department.

The number of enquiries dealt with during the past year again increased. Members seeking information on technical subjects are invited to avail themselves of the service provided.

MEETINGS.

Annual Meeting.

The Annual Meeting was held at the offices of the Institute, 4, Grosvenor Gardens, London, S.W.1, on Thursday, 11th May, 1944; there were two sessions, at 10.45 A.M. and 2.30 P.M. Mr. James Henderson, President, was in the Chair at the commencement and was followed by Mr. Arthur Dorman, the President-Elect. Twenty-seven papers, including twelve submitted under the auspices of Joint Research Committees of the Institute and the British Iron and Steel Federation, were presented, and nine papers were discussed at the meeting.

Autumn Meeting.

The Autumn Meeting was held at The Institution of Civil Engineers, Great George Street, London, S.W.1, on Thursday, 23rd November, at 11.0 A.M. and 2.45 P.M., and Friday, 24th November, 1944, at 10.30 A.M.; the President, Mr. Arthur Dorman, was in the Chair. Twelve papers, including seven submitted by Joint Research Committees of the Institute and the British Iron and Steel Federation, and one Special Report were presented; the Special Report and one paper were selected as the basis for a Discussion on Blast-Furnace Operation and Problems.

RELATIONS WITH OTHER SOCIETIES AND TECHNICAL INSTITUTIONS.

Friendly relations were maintained with scientific societies and technical institutions in Great Britain, the Dominions and allied and neutral countries.

Co-operation with The Institute of Metals.

The close feeling of friendly co-operation which has distinguished relations with The Institute of Metals during recent years was continued. On the retirement of Mr. G. Shaw Scott, Secretary of that Institute, on 30th June, 1944, Mr. K. Headlam-Morley (Secretary of The Iron and Steel Institute) was, by agreement between the Councils, appointed to be also Secretary of The Institute of Metals. Subsequently, arrangements for further close working were made. As a result certain members of the staffs have been appointed to serve both Institutes as from 1st January, 1945. Among others, Mr. A. E. Chattin (Assistant Secretary of The Iron and Steel Institute) was appointed to be also the Assistant Secretary of The Institute of Metals; he will also serve as Executive Editor (instead of Assistant Editor) of The Iron and Steel Institute. Mr. N. B. Vaughan, hitherto Assistant Editor of The Institute of Metals, was appointed to serve in the same capacity also for The Iron and Steel Institute.

The Council are confident that these arrangements will promote efficiency, reduce overhead expenditure and enable better service to be given to Members.

Co-operation with Local Societies.

The Council record their pleasure at the continued friendly relations existing with the following societies on the same basis as in previous years:

Cleveland Institution of Engineers.
Ebbw Vale Metallurgical Society.

Lincolnshire Iron and Steel Institute.
Manchester Metallurgical Society.
Newport and District Metallurgical Society.
Sheffield Metallurgical Association.
Sheffield Society of Engineers and Metallurgists.
Staffordshire Iron and Steel Institute.
Swansea Technical College Metallurgical Society.
West of Scotland Iron and Steel Institute.

Affiliation with Local Societies.

The Council are pleased to announce that as from 1st January, 1945, affiliation has been arranged with three of the above societies, namely :

Lincolnshire Iron and Steel Institute.
Sheffield Society of Engineers and Metallurgists.
Staffordshire Iron and Steel Institute.

The societies retain their complete independence, but have agreed to provide facilities for Members of the Institute in the districts in which they operate on terms which, it is believed, will prove mutually advantageous. Details will be included in the Report of Council for 1945.

Joint Meetings.

(*For the period from* 1*st April,* 1944, *to* 30*th April,* 1945.)

The following are particulars of Joint Meetings held with Local Societies (up to 30th April, 1945) :

Monday, 17th April, 1944 : Cleveland Institution of Engineers.

Place and Time : Cleveland Scientific and Technical Institute, Corporation Road, Middlesbrough, at 6.30 P.M.

Chairman : Mr. L. F. Wright, President of the Cleveland Institution of Engineers.

Paper : " The Heating of Open-Hearth Furnaces with Mixed Coke-Oven and Blast-Furnace Gas," by R. W. Evans, B.Met.

Wednesday, 19th July, 1944 : Sheffield Branch of the Institute of British Foundrymen.

Place and Time : Royal Victoria Hotel, Sheffield, at 7 P.M.

Chairman : Mr. W. J. Dawson, Chairman of the Steel Castings Research Committee.

Paper : " Developments in the Design and Use of Side-Blown Converter Plants," by Mr. P. C. Fassotte.

Friday, 8th September, 1944 : Middlesbrough and District Branch of the Institute of British Foundrymen.

Place and Time : Cleveland Scientific and Technical Institute, Corporation Road, Middlesbrough, at 7.30 P.M.

Chairman : Mr. H. Foster, President of the Middlesbrough and District Branch of the Institute of British Foundrymen.

Paper : " Developments in the Design and Use of Side-Blown Converter Plants," by Mr. P. C. Fassotte.

Wednesday, 18th October, 1944 : Manchester Metallurgical Society and The Institute of Metals.

Place and Time : The Engineers' Club, Albert Square, Manchester, at 6.30 P.M.

Chairman : Mr. N. S. Hubbard, B.Sc., Vice-President of the Manchester Metallurgical Society.

Lecture : " Deformation of Metals," by Professor H. W. Swift, M.A., D.Sc.

Tuesday, 24th October, 1944 : Lincolnshire Iron and Steel Institute.

Place and Time : Modern School, Cole Street, Scunthorpe, at 7.30 P.M.

Chairman : Mr. J. N. Kilby, President of the Lincolnshire Iron and Steel Institute.

Paper : " Some Notes on Slags and Slag Control in Basic Open-Hearth Tilting Furnaces Using Phosphoric Iron," by Mr. A. Jackson.

Monday, 27th November, 1944 : Sheffield Branch of the Institute of British Foundrymen.

Place and Time : Royal Victoria Hotel, Sheffield, at 7 P.M.

Chairman : Dr. C. J. Dadswell, Past-President of the Sheffield Branch of the Institute of British Foundrymen.

Paper : " The Side Feeding of Steel Castings. A Note on the Influence of the Mechanism of Freezing," by B. Gray, B.A. (Cantab.).

Saturday, 2nd December, 1944 : Sheffield Metallurgical Association and the Sheffield Society of Engineers and Metallurgists.

Place and Time : Royal Victoria Hotel, Sheffield, at 2.30 P.M.

Chairman : Mr. R. Staton, President of the Sheffield Metallurgical Association, assisted by Mr. F. Wardrobe, President of the Sheffield Society of Engineers and Metallurgists.

Papers :

" Tensile Properties of Unstable Austenite and Its Low-Temperature Decomposition Products," by A. H. Cottrell, B.Sc., Ph.D.

" The Effect of Overheating on the Transformation Characteristics of a Nickel-Chromium-Molybdenum Steel," by K. Winterton, B.Sc., Ph.D.

Thursday, 14th December, 1944 : Staffordshire Iron and Steel Institute.

Place and Time : Dudley and Staffordshire Technical College, The Broadway, Dudley, at 7 P.M.

Chairman : Dr. J. E. Hurst, President of the Staffordshire Iron and Steel Institute.

Paper : " Developments in the Design and Use of Side-Blown Converter Plants," by Mr. P. C. Fassotte.

Monday, 18th December, 1944 : Cleveland Institution of Engineers.

Place and Time : Cleveland Scientific and Technical Institute, Corporation Road, Middlesbrough, at 6.30 P.M.

Chairman : Mr. L. F. Wright, President of the Cleveland Institution of Engineers.

Paper : " Considerations on Blast-Furnace Practice," by T. P. Colclough, D.Sc., M.Met., F.R.I.C.

Thursday, 18th January, 1945 : Staffordshire Iron and Steel Institute.

Place and Time : Dudley and Staffordshire Technical College, The Broadway, Dudley, at 7 P.M.

Chairman : Dr. J. E. Hurst, President of the Staffordshire Iron and Steel Institute.

Paper : " Some Observations on the Austempering and Isothermal Transformation of Steels, with Special Reference to the Production of Martensite," by Professor F. C. Thompson, D.Met., M.Sc., and L. R. Stanton, Ph.D.

Monday, 22nd January, 1945 : Sheffield Branch of the Institute of British Foundrymen and the Refractories Association of Great Britain.

Place and Time : Royal Victoria Hotel, Sheffield, at 7 P.M.

Chairman : Mr. E. Barron, President of the Sheffield Branch of the Institute of British Foundrymen.

Paper : " The Effect of Grain Shape on the Moulding Properties of Synthetic Moulding Sands," by W. Davies, M.Sc., F.G.S., and W. J. Rees, D.Sc. Tech., F.R.I.C.

Tuesday and Wednesday, 23rd and 24th January, 1945 : Lincolnshire Iron and Steel Institute.

Place and Time : Scunthorpe Technical School, Cole Street, Scunthorpe, at 7.30 P.M. each day.

Chairman : Mr. J. N. Kilby, President of the Lincolnshire Iron and Steel Institute.

Paper : " Ironmaking at the Appleby-Frodingham Works of The United Steel Companies, Limited," by Mr. G. D. Elliot and the Staffs of the Appleby-Frodingham Ironworks, Scunthorpe, and of the Central Research Department, Stocksbridge (The United Steel Companies, Ltd.). (Special Report No. 30.)

Wednesday, 7th February, 1945: Cardiff Local Members.

Place and Time : South Wales Institute of Engineers, Park Place, Cardiff, at 6 P.M.

Chairman : Mr. J. S. Hollings, C.B.E., Vice-President of The Iron and Steel Institute.

Paper : " Ironmaking at the Appleby-Frodingham Works of The United Steel Companies, Limited," by Mr. G. D. Elliot and the Staffs of the Appleby-Frodingham Ironworks, Scunthorpe, and of the Central Research Department, Stocksbridge (The United Steel Companies, Ltd.). (Special Report No. 30.)

Thursday, 8th February, 1945 : Ebbw Vale Metallurgical Society.

Place and Time : Tabernacle Vestry, Armoury Hill, Ebbw Vale, at 6.30 P.M.

Chairman : Mr. G. A. Young, Assistant General Manager of Messrs. Richard Thomas and Baldwins, Ltd., Ebbw Vale.

Lecture : " Blast-Furnace Design, Operation and Problems," by Mr. G. D. Elliot.

Monday, 19th February, 1945 : Cleveland Institution of Engineers.

Place and Time : The Cleveland Scientific and Technical Institute, Corporation Road, Middlesbrough, at 6.30 P.M.

Chairman : Mr. L. F. Wright, President of the Cleveland Institution of Engineers.

Paper : " Ironmaking at the Appleby-Frodingham Works of The United Steel Companies, Limited," by Mr. G. D. Elliot and the Staffs of the Appleby-Frodingham Ironworks, Scunthorpe, and of the Central Research Department, Stocksbridge (The United Steel Companies, Ltd.). (Special Report No. 30.)

Monday, 26th March, 1945 : Sheffield Branch of the Institute of British Foundrymen.

Place and Time : Royal Victoria Hotel, Sheffield, at 7 P.M.

Chairman : Dr. C. J. Dadswell, Past-President of the Sheffield Branch of the Institute of British Foundrymen.

Report : " First Report of the Foundry Practice Sub-Committee of the Steel Castings Research Committee." Presented by Mr. F. Cousans, Chairman of the Sub-Committee.

Saturday, 14th April, 1945 : Middlesbrough and District Branch of the Institute of British Foundrymen.

Place and Time : The Imperial Hotel, Darlington, at 3 P.M.

Chairman : Mr. H. Foster, President of the Middlesbrough and District Branch of the Institute of British Foundrymen.

Paper : " The Side-Feeding of Steel Castings," by B. Gray, B.A. (Cantab.).

The Council desire once again to record their thanks to the Presidents, Councils and Secretaries of the Local Societies, as well as to the authors of the papers.

The Council were pleased to accept invitations from other societies for Members of the Institute to take part in certain of their meetings :

Wednesday, 15th March, 1944 : The Institute of Metals.

Place and Time : The Institution of Mechanical Engineers, Storey's Gate, London, S.W.1, at 2.30 P.M.

Lecture : " Modern Views on Alloys and their Possible Application," by Dr. W. Hume-Rothery, F.R.S.

Monday, 17th April, 1944 : The Wolverhampton and District Engineering Society.

Place and Time : The Wolverhampton and Staffordshire Technical College, Wulfruna Street, Wolverhampton, at 7.15 P.M.

Lecture : " The Howrah Bridge, India," by Dr. H. P. Bugden.

Saturday, 20th January, 1945 : Swansea and District Metallurgical Society.

Place and Time : Royal Institution, Swansea, at 5 P.M.

Paper : " Slag Control in Open-Hearth Furnaces," by Mr. N. H. Bacon.

Research.

British Iron and Steel Research Association.

The British Iron and Steel Research Association was formed during 1944 under the scheme of the Department of Scientific and Industrial Research by the British Iron and Steel Federation. Under the constitution the Federation is entitled to nominate twenty-one and the Institute nine Members of Council. The Research Association took over responsibility for the direction of the industry's co-operative research activities from the Iron and Steel Industrial Research Council as from 1st January, 1945.

The Council wish the new Research Association every success in its important work for the industry.

Joint Research Committees.

Active collaboration with the Iron and Steel Industrial Research Council was continued on the same basis as in former years.

The Institute's Joint Research Committees have suffered further severe losses owing to the deaths of Dr. T. Swinden, who had succeeded the late Dr. W. H. Hatfield, F.R.S., as Chairman of the Heterogeneity of Steel Ingots Committee, the Corrosion Committee and the Alloy Steels Research Committee, and of Dr. G. D. Bengough, F.R.S., Chairman of the Marine Corrosion Sub-Committee (in January, 1945). Mr. H. H. Burton was elected Chairman of the Committee on the Heterogeneity of Steel Ingots and Mr. W. J. Dawson of the Alloy Steels Research Committee and the Corrosion Committee. Professor J. E. Harris became Vice-Chairman of the Marine Corrosion Sub-Committee (in January, 1945).

The following is a list of the Joint Research Committees and of their Sub-Committees; the number of meetings recorded in 1944 was 79 (56 in 1939; 59 in 1940; 64 in 1941; 75 in 1942; 62 in 1943) :

Alloy Steels Research Committee : Chairman, Dr. T. Swinden, later Mr. W. J. Dawson. Established June, 1934. Meetings held during 1944 : five.

Thermal Treatment Sub-Committee : Chairman, Mr. P. B. Henshaw. Established January, 1936. Meetings held during 1944 : three.

Hair-Line Crack Sub-Committee : Chairman, Mr. H. H. Burton. Established July, 1938. Meetings held during 1944 : five.

Special Aero-Components Sub-Committee : Established July, 1940. Meetings held during 1944 : (not recorded).

Corrosion Committee: Chairman, Dr. T. Swinden, later Mr. W. J. Dawson. Established July, 1928. Meetings held during 1944: two.

Atmospheric Corrosion Sub-Committee: Chairman, Mr. T. H. Turner. Established June, 1944. Meetings held during 1944: one.

Protective Coatings Sub-Committee: Chairman, Mr. T. M. Herbert. Established January, 1936. Meetings held during 1944: three.

Marine Corrosion Sub-Committee: Chairman, Dr. G. D. Bengough, F.R.S. Re-formed November, 1938. Meetings held during 1944: four.

Sub-Committee on the Corrosion of Buried Metals (working in collaboration with the Committee on Soil Corrosion of Metals and Cement Products of The Institution of Civil Engineers): Established October, 1937. No meetings held during 1944; activities carried on by correspondence.

Laboratory (Corrosion) Research Sub-Committee: Chairman, Dr. U. R. Evans. Established June, 1930. Disbanded June, 1944.

Sub-Committee on Low-Alloy Steels: Established June, 1938. Disbanded June, 1944.

Heterogeneity of Steel Ingots Committee: Chairman, Dr. T. Swinden, later Mr. H. H. Burton. Established May, 1924. Meetings held during 1944: five.

Ingot Moulds Sub-Committee (formed jointly by the Committee on the Heterogeneity of Steel Ingots and the Open-Hearth Committee of the Iron and Steel Industrial Research Council): Chairman, Mr. R. H. Myers. Established November, 1934. Meetings held during 1944: one. (Stresses in Moulds Panel, one.)

Joint Sub-Committee on the Physical Chemistry of Steelmaking (formed jointly by the Committee on the Heterogeneity of Steel Ingots and the Open-Hearth Committee of the Iron and Steel Industrial Research Council). Established 1938. No meetings held during 1944.

Liquid Steel Temperature Sub-Committee: Chairman, Mr. E. W. Elcock. Established March, 1929. Meetings held during 1944: four.

Sub-Committee on Gaseous and Non-Metallic Inclusions (including the Oxygen Sub-Committee): Chairman, Dr. T. Swinden (in 1945, Mr. W. W. Stevenson). Established January, 1936; title changed February, 1944. Meetings held during 1944: two. (Chemists' Panel, two.)

Inclusions Sub-Committee: Chairman, Mr. W. J. Dawson. Established November, 1936. Meetings held during 1944: five.

Standard Methods of Analysis Sub-Committee: Chairman, Dr. E. Gregory. Established September, 1939. Meetings held during 1944: eight.

Steel Castings Research Committee: Chairman, Mr. W. J. Dawson. Established November, 1934. Meetings held during 1944: five.

Moulding Materials Sub-Committee: Chairman, Dr. W. J. Rees. Established March, 1936. Meetings held during 1944: five.

Foundry Practice Sub-Committee: Chairman, Mr. F. Cousans. Established May, 1938. Meetings held during 1944: six.

Foundry Steel Temperature Sub-Committee: Chairman, Mr. D. A. Oliver. Established December, 1941. Meetings held during 1944: four.

Side-Blown Converter Practice Sub-Committee: Chairman, Dr. T. P. Colclough. Established September, 1943. Meetings held during 1944: four.

Refractories Panel (reporting through the Side-Blown Converter Practice Sub-Committee to the Open-Hearth Refractories Joint Panel of the British Refractories Research Association and the Iron and Steel Industrial Research Council): Chairman, Dr. R. J. Sarjant. Established December, 1943. Meetings held in 1944: three.

Metallurgical (Steel Castings) Sub-Committee: Chairman, Mr. W. J. Dawson. Established June, 1944. Meetings held during 1944: two.

PUBLICATIONS.

Papers and Committee Reports.—Two volumes of the *Journal*, Special Report No. 29, " Review of the Work of the Joint Research Com-

mittees, 1924–1943, of The Iron and Steel Institute and the British Iron and Steel Federation," and Special Report No. 30, "Ironmaking at the Appleby-Frodingham Works of The United Steel Companies, Limited," by G. D. Elliot and the Staffs of the Appleby-Frodingham Ironworks, Scunthorpe, and of the Central Research Department, Stocksbridge (The United Steel Companies, Ltd.), were published during the year. The two issues of the *Journal* contain twenty papers issued under the auspices of the Joint Research Committees of The Iron and Steel Institute and The British Iron and Steel Federation and twenty-one other papers, as follows :

D. Binnie : "A Study of a Shell Steel Ingot." (Paper No. 25/1943 of the Committee on the Heterogeneity of Steel Ingots.)

Blast-Furnace Materials Analysis Sub-Committee of the Blast-Furnace Committee : "The Determination of Sulphur and Phosphorus in Pig Iron."

T. P. Colclough : "Considerations on Blast-Furnace Practice."

W. Davies and W. J. Rees : "The Bonding Properties of Mixtures of Petroleum Extracts and Linseed Oil and of the Extracts themselves." (Paper No. 9/1944 of the Steel Castings Research Committee (submitted by the Moulding Materials Sub-Committee).)

W. Davies and W. J. Rees : "The Effect of Grain Shape on the Moulding Properties of Synthetic Moulding Sands." (Paper No. 8/1944 of the Steel Castings Research Committee (submitted by the Moulding Materials Sub-Committee).)

N. L. Evans : "The Use of Basic-Lined Ladles in the Desulphurisation of Cast Iron by Sodium Carbonate."

N. L. Evans : "Cleaning and Descaling Steel by Electrolytic Pickling in Molten Caustic Soda."

R. W. Evans : "The Heating of Open-Hearth Furnaces with Mixed Coke-Oven and Blast-Furnace Gas."

U. R. Evans : "Progress in the Corrosion Research Section at Cambridge University." (Paper No. 12/1943 of the Corrosion Committee.)

F. Fancutt and J. C. Hudson : "The Formulation of Anti-Corrosive Compositions for Ships' Bottoms and Underwater Service on Steel. The Effect of the Pigment and of the Medium." (Paper No. 13/1944 of the Corrosion Committee (submitted by the Marine Corrosion Sub-Committee).)

P. C. Fassotte : "Developments in the Design and Use of Side-Blown Converter Plants."

B. Gray : "The Side Feeding of Steel Castings. A Note on the Influence of the Mechanism of Freezing." (Paper No. 10/1944 of the Steel Castings Research Committee (submitted by the Foundry Practice Sub-Committee).)

J. A. Hall : "The Drift of Selenium Photo-Electric Cells in Relation to their Use in Temperature Measurement." (Paper No. 26/1943 of the Committee on the Heterogeneity of Steel Ingots (submitted by the Liquid Steel Temperature Sub-Committee).)

K. Hoselitz : "The Iron-Nickel Phase Diagram by Magnetic Analysis and the Effects of Cold-Work."

A. W. Hothersall, D. W. Hopkins and G. L. Evans : "Factors Affecting the Strength of Soldered Joints made from Electro-Tinned Steel Sheet."

A. U. Huddle and U. R. Evans : "Some Measurements of Corrosion-Fatigue made with a New Feeding Arrangement." (Paper No. 11/1943 of the Corrosion Committee.)

J. E. Hurst and R. V. Riley : "A Note on the Microstructure of High-Silicon Acid-Resisting Iron."

J. E. Hurst and R. V. Riley : "The Occurrence of the Carbide Phase in High-Silicon Iron-Carbon Alloys."

A. Jackson : "Some Notes on Slags and Slag Control in Basic Open-hearth Tilting Furnaces using Phosphoric Iron."

R. Jackson, R. J. Sarjant, J. B. Wagstaff, N. R. Eyres, D. R. Hartree, F.R.S., and J. Ingham : "Variable Heat Flow in Steel." (Paper No. 15/1944 of the Alloy Steels Research Committee (submitted by the Thermal Treatment Sub-Committee).)

T. Land : "Barrier-Layer Photo-Electric Cells for Temperature Measurement." (Paper No. 6/1943 of the Steel Castings Research Committee (submitted by the Foundry Steel Temperature Sub-Committee).)

F. László : "Tessellated Stresses.—Part III."

A. H. Leckie : "The Study of the Thermal Performance of Open-Hearth Furnaces by the Correlation of Operating Data."

H. Lipson and Audrey M. B. Parker : "The Structure of Martensite."

D. MANTERFIELD : "Survey of Liquid Steel Temperatures in Basic Open-Hearth Furnaces." (Paper No. 23/1943 of the Committee on the Heterogeneity of Steel Ingots (submitted by the Liquid Steel Temperature Sub-Committee).)

MARINE CORROSION SUB-COMMITTEE : "Fouling of Ships' Bottoms : Identification of Marine Growths." (Paper No. 14/1944 of the Corrosion Committee (submitted by the Marine Corrosion Sub-Committee).)

B. MASON : "The Constitution of Some Basic Open-Hearth Slags."

D. A. OLIVER and T. LAND : "A Thermocouple Method for the Measurement of Liquid Steel Casting-Stream Temperatures." (Paper No. 5/1943 of the Steel Castings Research Committee (submitted by the Foundry Steel Temperature Sub-Committee).)

N. J. PETCH : "The Interpretation of the Crystal Structure of Cementite."

A. PREECE and R. V. RILEY : "The Scaling Properties of Steels in Furnace Atmospheres at 1150° C." (Paper No. 14/1943 of the Alloy Steels Research Committee.)

D. J. PRICE and H. LOWERY : "The Emissivity Characteristics of Hot Metals, with Special Reference to the Infra-Red." (Paper No. 7/1943 of the Steel Castings Research Committee (submitted by the Foundry Steel Temperature Sub-Committee).)

H. T. PROTHEROE : "The Influence of Melting Conditions on the Physical Properties of Steel Castings." (Paper No. 11/1944 of the Steel Castings Research Committee.)

G. C. RICHER : "The Magnetisation of Polycrystalline Iron and Iron-Silicon Alloys."

T. H. SCHOFIELD : "Note on some Unusual Microstructures Observed in Mild and Medium-Carbon Steels."

W. STEVEN : "The Effect on the Hardenability of Small Additions of Chromium and Molybdenum to a Grain-Size-Controlled 0·9% Nickel Steel."

T. SWINDEN : "The Examination of a Rimming-Steel Ingot containing 0·29% of Carbon." (Paper No. 27/1944 of the Committee on the Heterogeneity of Steel Ingots.)

T. SWINDEN, W. W. STEVENSON and G. E. SPEIGHT : "Rimming-Steel.—An Examination of the Carbon and Oxygen Relationship in the Solidification of Basic Open-Hearth Steel." (Paper No. 24/1943 of the Committee on the Heterogeneity of Steel Ingots.)

T. SWINDEN, W. W. STEVENSON and G. E. SPEIGHT : "Rimming Steel.—Experiments on Melts of Rimming-Steel Composition in the Laboratory High-Frequency Furnace." (Paper No. 28/1944 of the Committee on the Heterogeneity of Steel Ingots.)

W. H. J. VERNON, F. WORMWELL and T. J. NURSE : "A Study of the Surface Film on Chromium-Nickel (18/8) Stainless Steel." (Paper No. 15/1944 of the Corrosion Committee.)

W. J. WRAŻEJ : "The Apparent Microstructure Produced by Hydrofluoric Acid Etching Reagents on Pure Iron and Iron-Silicon Alloys."

"The Training of Metallurgists, with Special Reference to the Iron and Steel Industries." Issued by authority of the Council of The Iron and Steel Institute.

The Bulletin of The Iron and Steel Institute was published monthly during the year; as usual, it was reprinted as Section II. of the *Journals* issued for the corresponding periods. The *Bulletin* is supplied free of charge to Members on application; the subscription rate to non-members is 30*s*. per annum ($6 to members of the American Iron and Steel Institute, the American Institute of Mining and Metallurgical Engineers and the American Society for Metals).

Translation Service.—Translations of thirty-eight foreign technical papers were included in the Institute's Translation Series; they are obtainable by Members at 10*s*. each (5*s*. for each additional copy of the same translation). The Council wish to record their appreciation of the collaboration received from a number of Companies, Research Associations and Government Departments.

STAFF.

Mr. R. Elsdon, Librarian, completed 40 years' and Miss W. F. Dowd, Accountant, 25 years' service.

The Council again wish to record their appreciation of the services rendered by the staff.

(*See also* paragraph on Co-operation with The Institute of Metals.)

Honours Conferred on Members.

(To 31st March, 1945.)

The Council tender their warmest congratulations to the following Members of the Institute for honours and appointments received during the period under review:

Allen, Dr. N. P.—Appointed Superintendent of the Metallurgy Department, National Physical Laboratory.

Bent, Quincey.—Awarded the Gary Memorial Medal by the American Iron and Steel Institute.

Brown, P. B.—Created a Knight Bachelor.

Bruce-Gardner, Sir Charles.—Created a Baronet. Appointed Chief Executive for Industrial Reversion to the President of the Board of Trade.

Cunniffe, C. H.—Created a Member of the Order of the British Empire.

Edwards, T. W.—Elected President of the Rotherham and District Chamber of Commerce; sixth successive year.

Gardom, J. W.—Elected President of the Institute of British Foundrymen. Awarded the British Foundry Medal.

Griffiths, Dr. W. T.—Re-elected President of the Institute of Metals.

Henderson, James.—Made an Honorary Member of The Iron and Steel Institute.

Holmstrom, C. E.—Elected President of the Sheffield Chamber of Commerce.

Hurst, Dr. J. E.—Elected President of the Staffordshire Iron and Steel Institute.

Jones, Lewis, M.P.—Created a Knight Bachelor.

Kayser, J. F.—Elected President of the London Branch of the Institute of British Foundrymen.

Larke, Sir Wm., K.B.E.—Succeeded Sir Percy Ashley as Chairman of the General Council of the British Standards Institution.

Lowe, R.—Elected Chairman of the Junior Institute of Engineers.

McCance, Dr. A., F.R.S.—Elected Chairman of the British Iron and Steel Research Association. Appointed a Justice of the Peace for Lanarkshire.

McCosh, A. K.—Elected President of the Mining Association of Great Britain (1944).

Macdiarmid, A. C.—Created a Knight Bachelor.

Mehl, Dr. R. F.—Awarded the Douglas Medal of the American Institute of Mining and Metallurgical Engineers. Awarded the Degree of Doctor *Honoris Causa* by the Escola Politecnica of the University of Sao Paulo, Brazil; received gold medal for his assistance in organising the first Brazilian Metallurgical Society.

Menzies-Wilson, J. R.—Appointed Director-General of Metallurgical Industries in the Economics Division of the Control Commission for Germany (British Element).

Munro, A. A.—Elected President of the Sheffield Metallurgical Association.

Rees, Dr. W. J.—Created an Officer of the Order of the British Empire.

Saniter, F. H.—Appointed Director of Research to The United Steel Companies, Ltd.

Stoughton, Professor Bradley.—Awarded the Honorary Degree of Doctor of Engineering by Lehigh University.

Swinden, the late Dr. T.—Elected Junior Vice-President of the Institute of British Foundrymen.

Wood, W. W., J.P.—Re-elected Master Cutler; sixth term of office.

Appointment of Representatives.

The following is a list of the Institute's representatives on various governing bodies and committees at 31st March, 1945 :

British Cast Iron Research Association : Professor T. Turner.

British Corporation Register of Shipping and Aircraft, Technical Committee : Dr. A. McCance, F.R.S.

British Electrical and Allied Industries Research Association,
- Sub-Committee J/E, Joint Committee, Steels for High Temperatures : Dr. J. W. Jenkin, Mr. L. Rotherham.
- Sub-Committee F/J, Earthing to Water Mains : Dr. J. C. Hudson.

British Iron and Steel Federation, Statistical Committee : Mr. K. Headlam-Morley.

British Iron and Steel Research Association, Council : Professor J. H. Andrew, Mr. G. Wesley Austin, Mr. Arthur Dorman, Principal C. A. Edwards, F.R.S., Dr. E. Gregory, Dr. W. T. Griffiths, Mr. James Henderson, The Hon. R. G. Lyttelton, Dr. W. J. Rees.

British Refractories Research Association, Council : Mr. W. J. Brooke, J.P.

British Standards Institution,
- General Council and Executive Committee: Mr. J. R. Beard (representing the Founder Institutions).
- Chemical Engineering Divisional Council : Mr. E. F. Law.
- Engineering Divisional Council E/- : The Hon. R. G. Lyttelton, Mr. K. Headlam-Morley, Mr. W. B. Baxter.
- Sub-Committee IS/35/3, Malleable Steel Castings : Mr. C. H. Kain.
- Sub-Committee C/25/10, Painting of Iron and Steel : Mr. F. Fancutt, Mr. R. A. Hacking.
- Sub-Committee CEB/6/1, Concrete Blocks : Mr. W. J. Brooke, J.P.
- Technical Committee CEB/1, Cement : Mr. W. J. Brooke, J.P.
- Technical Committee CH/17, Symbols used in Diagrams of Chemical Engineering Plant : Mr. A. E. Chattin.
- Sub-Committee CH/18/2, Metallic Finishes : Mr. T. W. Whiting.
- Sub-Committee CH/18/4, Phosphate Coatings : Dr. J. C. Hudson.
- Technical Committee EL/28, Fans : Mr. A. F. Webber, Dr. A. M. Burdon.
- Technical Committee HIB/15, Pressed Steel Galvanised Rainwater Goods : Dr. A. M. Burdon.
- Iron and Steel Industry Committee IS/- : *appointment open.*
- Technical Committee IS/-/1, Advisory Committee on Iron and Steel : *appointment open.*
- Technical Committee IS/1, Co-ordination of Iron and Steel Specifications : *appointment open.*
- Technical Committee IS/6, Steel Castings : Dr. R. H. Greaves.
- Technical Committee IS/8, Creep Properties : Mr. L. Rotherham.
- Technical Committee IS/11, Galvanised Corrugated Steel Sheets : Dr. J. C. Hudson.
- Technical Committee IS/15, Iron and Steel for Shipbuilding : Sir Edward J. George.
- Technical Committee IS/17, Cast Iron Columns for Street Lighting : Mr. J. G. Pearce.
- Technical Committee IS/28, Protective Coatings for Iron and Steel : Dr. J. C. Hudson.
- Technical Committee IS/35, Cast Iron : Mr. J. G. Pearce.
- Technical Committee ME/23, Brinell Hardness Testing : *appointment open.*
- Technical Committee ME/25, Testing of Metal Sheets and Strip : Dr. A. M. Burdon.

Technical Committee ME/32, Engineering Symbols and Abbreviations: Dr. A. M. Burdon.
Technical Committee PAM/160, Reels and Drums for Electrical Wires: Dr. A. M. Burdon.
Technical Committee PSM/2, Welders' Goggles: Dr. C. H. Desch, F.R.S.
Solid Fuel Industry Committee, SF/-: Mr. A. F. Webber.
Technical Committee SF/1, Nomenclature and Definitions: Mr. A. F. Webber.
Technical Committee SF/2, Underfed Screw Type Stokers: Mr. A. F. Webber.
Technical Committee SF/4, Heating Stoves: Mr. A. F. Webber.
Units and Technical Data Co-ordinating Committee: Sir Wm. Larke, K.B.E.
Conference M/4, Letter Symbols: Dr. A. M. Burdon.

CITY AND GUILDS OF LONDON INSTITUTE, Advisory Committee on Metallurgy: Dr. J. W. Jenkin.
CONSTANTINE COLLEGE, Advisory Committee: Mr. E. W. Jackson.
EMPIRE COUNCIL OF MINING AND METALLURGICAL INSTITUTIONS: Mr. K. Headlam-Morley, Mr. J. Sinclair Kerr.
ENGINEERING PUBLIC RELATIONS COMMITTEE,
Main Committee: Mr. James Henderson.
Executive Committee: Mr. K. Headlam-Morley.
HATFIELD MEMORIAL LECTURE COMMITTEE: Mr. K. Headlam-Morley, Dr. C. Sykes, F.R.S.
HONG-KONG UNIVERSITY, Home Committee: *appointment open.*
IMPERIAL COLLEGE OF SCIENCE AND TECHNOLOGY, Board of Governors: Mr. James Henderson.
IMPERIAL INSTITUTE, Mineral Resources Department, Iron and Ferro-Alloy Metals Committee: Mr. K. Headlam-Morley.
INSTITUTE OF FUEL, Council: Dr. R. J. Sarjant.
INSTITUTE OF WELDING,
Council and Representative of Patron Institution: Mr. K. Headlam-Morley.
Joint Library Committee (with British Welding Research Association): Mr. K. Headlam-Morley.
INSTITUTION OF CIVIL ENGINEERS, Committee on Soil Corrosion of Metals and Cement Products, Sub-Committee on Soil Corrosion of Metals: Dr. J. C. Hudson.
INSTITUTION OF MECHANICAL ENGINEERS, Research Committee on High-Duty Cast Irons for General Engineering Purposes: Dr. J. E. Hurst.
JOINT COMMITTEE ON MATERIALS AND THEIR TESTING: Mr. K. Headlam-Morley.
LIVERPOOL UNIVERSITY, Court of Governors: Sir W. Peter Rylands, Bt.
LLOYD'S REGISTER OF SHIPPING, Technical Committee: Mr. James Henderson, Mr. P. Baxter.
NATIONAL PHYSICAL LABORATORY, General Board: Professor J. H. Andrew, Mr. W. J. Dawson.
PARLIAMENTARY AND SCIENTIFIC COMMITTEE: Mr. K. Headlam-Morley.
RAMSAY MEMORIAL LABORATORY, Advisory Committee: *appointment open.*
ROYAL SCHOOL OF MINES, Advisory Board: Mr. Vernon Harbord.
ROYAL SOCIETY, General Board for Administering Government Grants for Scientific Investigations: The President.
SCHOOL OF METALLIFEROUS MINING (CORNWALL), Board of Governors: Mr. J. S. Hollings, C.B.E.
SCIENCE MUSEUM, Advisory Council: Mr. James Henderson.
SHEFFIELD UNIVERSITY, Court of Governors: Sir Arthur B. Winder, J.P.

[The Statement of Accounts for 1944 will be found in the following pages.]

THE IRON AND

BALANCE SHEET,

LIABILITIES.

1943. £	£			£	£
		Sundry Creditors :—			
512		Office Rent, Telephone, Heating, &c.		445	
808		Publishing (part estimated)		933	
173		Salaries, Insurance and Income Tax Staff Deductions		165	
80		Printing and Stationery, Library Books and Postage		160	
47		Travelling and Entertainment, Sundry Payments		105	
—		Autumn Meeting		43	
54		Subscriptions in Suspense		39	1,890
	1,674				
		Amounts received in Advance :—			
218		Members Subscriptions		108	
—		Special Subscriptions		155	
278		Sales of Publications		265	528
	496				
		Legacy by Sir Robert Hadfield, Bart., F.R.S.		268	
		Interest on Investment (Gross)		4	272
	268				
		Suspense Account as at 1st January, 1944 :—			
		Reserve for Ten-year Index	450		
		Add Further transfer, 1944	50	500	
450					
		Bessemer Gold Medal	75		
		Add Further transfer, 1944	25	100	
75					
		Library Account	956		
		Add Further transfer, 1944	50	1,000	
950					
		Repairs and Decorations	2,000	2,000	3,600
2,000	3,475				
		Entrance Fees Reserve Fund		2,934	
		Add Further transfer, 1944		270	3,204
	2,934				
		Life Composition Fund		3,459	
		Add Compounded during Year		179	3,638
	3,459				
		General Reserve		500	
		Add Further transfer, 1944		1,500	2,000
	500				
		Iron and Steel Institute :—			
23,882		Capital as per last Balance Sheet		23,882	
1,766		Accumulated Excess of Income over Expenditure ...	1,981		
215		*Add* Excess of Income over Expenditure for the Year...	571	2,552	26,434
	25,863				
	£38,669				£41,566

We have examined the above Balance Sheet

224 Regent Street,
London, W.1.
9th March, 1945.

STEEL INSTITUTE.

31st DECEMBER, 1944.

ASSETS.

1943. £	£		£	£
		Sundry Debtors :—		
—		Subscriptions in Arrear		N.V.
44		Publishing	76	
3,183		Income Tax Recoverable	3,397	
284		Sales of Publications	270	
24		Sundries	16	
—		Special Subscriptions	30	
—		National Certificates in Metallurgy	31	
580		Other Accounts	—	
	4,115			3,820
		Payments in Advance :—		
200		Staff Superannuation Fund	200	
146		Sundries	111	
—		Publishing Special Report No. 30	645	
	346			956
	—	Stock of Journals		N.V.
	—	Office Furniture and Library		N.V.
	1,433	Cash at Bank and in Hand		1,003
	1,372	Post Office Savings Bank Deposit Account		1,406
		Legacy by Sir Robert Hadfield, Bart., F.R.S. :—		
		Investment at Cost per Schedule	268	
		Cash at Bank	4	
	—			272
		(The Market Value of this Investment at 31st December, 1944, was £272)		
		Investments at Cost per Schedule :—		
	27,905	General Fund		29,905
		(The Market Value of these Investments at 31st December, 1944 was £33,983)		
		Joint Research Committees :—		
	39	Amount advanced		566
		Life Composition Fund :—		
		Investments at Cost per Schedule	3,601	
		Cash at Bank	37	
	3,459			3,638
		(The Market Value of these Investments at 31st December, 1944, was £3,853)		
	£38,669			£41,566

of the Institute and certify it to be correct.

W. B. KEEN & Co.,

Chartered Accountants.

N.V. = Not valued.

INCOME.

1943. £	£		£	£	£
153		Entrance Fees	195		
185		Do. Companies' Nominations	75		
	338			270	
	338	*Less* Transfer to Reserve Fund		270	
	—				—
		Annual Subscriptions :—			
5,036		Members, Home Current		5,568	
274		Do. Companies' Nominations		110	
230		Do. Arrears		165	
	5,540				5,843
937		Members, Overseas Current		1,030	
16		Do. Companies' Nominations		2	
236		Do. Arrears		251	
	1,189				1,283
306		Associates, Current		400	
24		Do. Companies' Nominations		38	
18		Do. Arrears		12	
	348				450
		Sales of Publications :—			
1,347		Journals, &c.		1,296	
65		Bulletin		157	
406		Translations		549	
	1,818				2,002
		Interest on Investments (Gross) :—			
1,143		General Fund		1,169	
124		Life Composition Fund		130	
16		Bessemer Medal Fund		16	
	1,283				1,315
	34	Interest on Deposit Account			34
	39	Sundry Receipts			40
		Institute of Metals :—			
750		Rent Receivable		750	
450		Contribution to Joint Library	475		
–50		*Less* Refund	50		
				425	
	1,150				1,175
		Iron and Steel Industrial Research Council :—			
750		Grant for Bulletin		750	
2,500		Grant for Secretarial Services		3,000	
500		Grant for Information Service		500	
	3,750				4,250
	2,625	Balance, being Excess of Expenditure over Income carried down ...			2,291
	£17,776				£18,683

£	£		£	£
		Special Subscriptions :—		
2,873		Contributions received during 1944	3,171	
1,541		Income Tax Recoverable	1,541	
	4,414			4,712
	£4,414			£4,712

FOR THE YEAR ENDED 31st DECEMBER, 1944.

EXPENDITURE.

1943. £	£		£	£
7,049		Salaries (including Pensions and Overtime)	7,653	
509		Staff in H.M. Forces	481	
750		War Bonus	1,155	
8,308			9,289	
688		*Less* Contribution from Ministry of Supply	—	
150		*Less* Contribution from Carnegie Research Fund	150	
	7,470			9,139
	63	National Insurance		74
	238	Staff Superannuation Fund		330
	1,200	Office Rent		1,200
	54	Repairs, Decorations and A.R.P.		54
	830	Cleaning, Heating, Lighting and Water		903
	395	Library Books and Binding		389
	34	Office Furniture		108
	60	Annual Meeting		68
	6	Autumn Meeting		49
		Publishing Expenses :—		
1,399		Journal : Printing and Paper	1,635	
172		Postage	141	
275		Advance Copies : Printing	301	
135		Postage	150	
453		Bulletin : Printing	477	
70		Postage	60	
11		Bibliographies	17	
119		Translations Service	748	
—		Special Publications	73	
	2,634			3,602
	580	Stationery and Printing		837
	652	Postage and Receipt Stamps		602
393		Travelling and Entertainment Expenses	370	
102		*Less* Recoverable from Ministry of Supply	—	
	291			370
58		Insurance	61	
90		Do. War Damage	31	
	148			92
	53	Auditors' Fee for 1943		52
	192	Telephone Rental and Calls		236
	169	Office Disbursements and Sundry Expenses		227
		Grants :—		
50		British Electrical and Allied Industries Research Association ...	50	
25		British Refractories Research Association	25	
1		Joint Committee on Materials and their Testing	1	
50		British Standards Institution	50	
10		Mellor Memorial Fund	10	
—		Hatfield Memorial Lecture Fund	105	
10		Parliamentary and Scientific Committee	10	
11		Foundry Prize—Sheffield University	—	
50		Sundry Researches	100	
	207			351
	2,500	Iron and Steel Industrial Research Council—Grant		—
	£17,776			£18,683

£	£			£	£
	2,625	Balance brought down			2,291
		Transfer to Reserve and Suspense Account :—			
300		Repairs and Decorations		—	
200		Library		50	
50		Ten-Year Index		50	
500		General Reserve		1,500	
25		Bessemer Gold Medal		25	
	1,075				1,625
		Companies' Nominations :—			
185		Entrance Fees		75	
		Subscriptions Account :—			
274		Home Members	110		
16		Overseas Members	2		
24		Associates	38	150	
	499				225
	215	Balance, being Excess of Income over Expenditure			571
	£4,414				£4,712

WILLIAMS

BALANCE SHEET,

LIABILITIES.

1943. £	£		£	£
	2,220	Capital Value representing Market Value of £3,000 3½% Conversion Loan at 21st September, 1926, when Fund was inaugurated		2,220
	72	Bank Overdraft—Current Account		—
		Income and Expenditure Account :—		
696		Balance as at 1st January, 1944	617	
—79		*Add* Excess of Income over Expenditure for the Year to 31st December, 1944	126	
	617			743
	£2,909			£2,963

INCOME AND EXPENDITURE ACCOUNT FOR

INCOME.

1943. £		£
121	Interest on 3½% Conversion Loan	121
5	Interest on Deposit	5
79	Excess of Expenditure over Income for the Year...	—
£205		£126

ANDREW CARNEGIE

BALANCE SHEET,

LIABILITIES.

1943. £	£			£	£
		Sundry Creditors :—			
220		Grants		90	
10		Auditors' Fees		10	
	230				100
	500	General Reserve		500	
	—	*Add* Further transfer, 1944		500	
					1,000
	136	Amount due to General Fund			—
	—	Bank Overdraft			23
21,241		Amount of Original Fund ($100,000)		21,241	
4,441		*Add* Amounts since Capitalised as per last Balance Sheet ...		4,441	
25,682				25,682	
4,196		*Add* Surplus Income as at 1st January, 1944	£4,596		
400		Excess of Income over Expenditure for the Year ...	559		
				5,155	
	30,278				30,837
	£31,144				£31,960

INCOME AND EXPENDITURE ACCOUNT FOR

INCOME.

1943. £		£
1,212	Interest on Investments (Gross)	1,238
4	Interest on Deposit	4
£1,216		£1,242

PRIZE FUND.

31st DECEMBER, 1944.

ASSETS.

1943. £		£
2,670	£3,452 15s. 7d. 3½% Conversion Loan (at cost)	2,670
239	Cash at Bank	293
£2,909		£2,963

THE YEAR ENDED 31st DECEMBER, 1944.

EXPENDITURE.

1943. £		£
200	Awards	—
5	Legal Expenses	—
—	Excess of Income over Expenditure for the Year...	126
£205		£126

RESEARCH FUND.

31st DECEMBER, 1944.

ASSETS.

1943. £		£
29,938	Investments at Cost as per Schedule	30,938
	(The Market Value of these Investments at 31st December, 1944, was £36,990)	
990	Cash at Bank	461
216	Income Tax Recoverable	561
£31,144		£31,960

THE YEAR ENDED 31st DECEMBER, 1944.

EXPENDITURE.

1943. £	£		£
200		Scholarship Grants	—
50		*Less* Grants not taken up	—
	150		
	—	Printing Reports	23
	150	Contribution to General Fund on account of Salaries	150
	6	Legal Expenses	—
	10	Audit Fee	10
	500	General Reserve—amount set aside	500
	400	Balance, being Excess of Income over Expenditure for the Year	559
	£1,216		£1,242

THE IRON AND STEEL INSTITUTE.

SCHEDULE OF INVESTMENTS AT 31st DECEMBER, 1944.

SHOWING NOMINAL VALUES, COST AND PRESENT MARKET VALUES.

Nominal Value. £	s.	d.	Nature of Security.	Market Value, 31st December, 1944. £	Cost. £
			GENERAL FUND OF THE INSTITUTE.		
2,197	7	0	3½% War Stock	2,288	2,161
1,324	7	4	3½% Conversion Loan	1,404	1,254
447	0	0	Southern Railway 4% Debenture Stock	514	449
1,872	0	0	London & North Eastern Railway 4% 2nd Guaranteed Stock	1,750	2,150
2,241	0	0	Do. do. 4% 1st do.	2,319	2,432
2,649	4	0	2½% Consolidated Stock	2,159	1,800
1,500	0	0	Buenos Ayres Great Southern Railway 4% Debenture Stock	930	1,595
2,954	1	0	4% Consolidated Stock	3,309	3,176
437	0	0	London & North Eastern Railway 3% Debenture Stock	374	376
12,242	19	3	4% Funding Loan, 1960/90	13,896	9,512
1,0000		0	3% Defence Bonds (Post Office Register)	1,000	1,000
2,000	0	0	3% Savings Bonds, 1955/65	2,030	2,000
2,000	0	0	Do. 1960/70	2,010	2,000
				33,983	29,905
			LIFE COMPOSITION FUND.		
587	13	10	3½% Conversion Loan	623	513
405	17	4	3% Local Loans	383	383
1,330	0	0	London & North Eastern Railway 3% Debenture Stock	1,137	1,255
594	0	0	London Passenger Transport Board 4½% "A" Stock	722	638
205	15	6	3½% War Stock	214	212
681	13	0	4% Funding Loan, 1960/90	774	600
				3,853	3,601

Nominal Value.	Nature of Security.	Market Value, 31st December, 1944.	Cost.
	ANDREW CARNEGIE RESEARCH FUND.		
	(Trustees: Sir Wm. Larke, K.B.E., James Henderson, The Hon. R. G. Lyttelton.)		
£ s. d.		£	£
6,897 14 11	3½% War Stock	7,182	6,896
800 0 0	Do.	833	794
2,693 12 0	3½% Conversion Loan	2,855	2,007
2,642 12 0	3% Local Loans	2,497	2,527
2,250 0 0	North Eastern Electric Supply Company 3½% Consolidated Debenture Stock	2,284	2,262
1,500 0 0	Great Western Railway 4% Debenture Stock	1,747	1,204
2,000 0 0	London Midland & Scottish Railway 4% Debenture Stock	2,160	1,693
1,312 0 0	London & North Eastern Railway 4% 1st Guaranteed Stock	1,358	881
5,000 0 0	Do. do. 4% Debenture Stock	5,375	3,540
6,000 0 0	Do. do. 3% do.	5,130	3,545
1,000 0 0	3% Savings Bonds, 1955/65	1,015	1,000
2,250 0 0	Do. do. 1960/70	2,261	2,250
2,813 2 7	2½% Consolidated Stock	2,293	2,339
		36,990	30,938
	WILLIAMS PRIZE FUND.		
	(Trustees: Sir Wm. Larke, K.B.E., James Henderson, The Hon. R. G. Lyttelton.)		
3,452 15 7	3½% Conversion Loan	£3,660	£2,670
	BESSEMER MEDAL FUND.		
	(Trustees: Sir Francis Samuelson, Desmond Lysaght, the Hon. R. G. Lyttelton and N. H. Rollason.)		
400 0 0	London Midland & Scottish Railway 4% Debenture Stock	£432	
	HADFIELD LEGACY.		
267 11 2	3% Savings Bonds, 1955/65	£272	£268

(*Signed*) R. LYTTELTON.
Hon. Treasurer.

(*Signed*) K. HEADLAM-MORLEY.
Secretary.

We have examined the foregoing Balance Sheets and Income and Expenditure Accounts with the Books and Vouchers of the Institute and certify them to be correct. We have also verified the Balances at the Bankers and the Securities for the Investments shown above.

W. B. KEEN & Co.,
Chartered Accountants.

224 REGENT STREET,
LONDON, W.1.
9th March, 1945.

THE INFLUENCE OF THE HEAT TREATMENT OF STEEL ON THE DAMPING CAPACITY AT LOW STRESSES.*

By the late Dr.-Ing. LEOPOLD FROMMER and A. MURRAY
(Research Laboratories, High Duty Alloys, Ltd.).

SYNOPSIS.

An electromagnetic method of inducing torsional oscillations in freely-suspended cylindrical steel bars was used for measurements of the damping capacity up to a maximum stress of 100 lb. per sq. in. There were six specimens, each 3 in. in dia. and 3 ft. long, all from the same melt, with a composition normal to a 0·6% carbon steel. Measurements were made in the normalised, 830° C.-oil-quenched and the fully-tempered conditions; by successive heat treatment these conditions were repeated.

The measurements showed that the damping value was a characteristic property which varied for each condition and was reproduced through two heat-treatment cycles. Within the experimental accuracy the torsional damping capacity is independent of oscillation frequency up to 7000 cycles per sec. and of stresses up to 100 lb. per sq. in. The damping values as expressed by the logarithmic decrement are $0{\cdot}5 \times 10^{-4}$ for the tempered condition, and $0{\cdot}7 \times 10^{-4}$ for the oil-quenched condition. The Brinell hardness values showed changes similar to these, and it is suggested that this correspondence is due to the state of aggregation of the carbide, resulting from the heat treatment given.

Introduction.

An investigation of the damping capacity, primarily of light alloys, resulted in the development of an experimental method which, it was felt, might provide information of value in respect of steel. Other investigators in this country have from time to time published data concerning the damping capacity of ferrous alloys determined by a mechanical vibration method usually described as the Föppl-Pertz method,[1, 2, 3] the stress range covered being approximately 1–5 tons per sq. in., in which range the damping capacity is found to be dependent on the stress.

The method, which will now be described briefly[4] with particular reference to the determination of the damping capacity of a 0·6% carbon steel, is a development of that employed by Förster,[5] Zener[6] and Wegel and Walther,[7] and is characterised by a limitation of the amplitude of a resonance vibration of a freely-suspended specimen, so that the surface shear stress never exceeds 90 lb. per sq. in., for which, within the limits of experimental error, the damping capacity is independent of such stress variation as may come up to this limit.

In view of the differences in stress range and frequency of vibration it is felt that it would be unwise to attempt to compare results obtained by these two different methods of measuring damping capacity. The Föppl-Pertz method has the merit of operating within the range of engineering stresses, whilst the resonance vibration method has the advantage of being relatively independent of stress within a limited yet sufficient range for the damping capacity to be considered indicative mainly of the metallurgical state of the material.

* Received July 18, 1944.

Definition of Damping Capacity.

In this work the logarithmic decrement has been used for expressing the damping capacity. This is defined by the relation:

$$\lambda = \log \frac{A_0}{A_1} = \frac{\delta A}{A_1} - \frac{1}{2}\left(\frac{\delta A}{A_1}\right)^2 + \frac{1}{3}\left(\frac{\delta A}{A_1}\right)^3 - \ldots \quad (1)$$

where A_0 = initial amplitude of vibration, A_1 = amplitude of vibration one cycle later and δA = loss of amplitude for the cycle from A_0 to A_1. Some workers have expressed their results in terms of the "specific damping capacity," which is expressed by $P = \Delta E/E$, where ΔE = energy lost during the cycle and E = maximum energy of that cycle.

Since the energy is proportional to the square of the vibration amplitude, then:

$$P = \left(\frac{A_0^2 - (A_0 - \delta A)^2}{A_0}\right) = \frac{2\delta A}{A_0} - \left(\frac{\delta A}{A_0}\right)^2 \quad \ldots \quad (2)$$

Thus the logarithmic decrement is half the specific damping capacity provided that δA is small compared with A_0.

In this work $\delta A/A_0$ was always less than 10^{-4} so that the second and higher terms in equations (1) and (2) are negligible.

In practice, the fall in amplitude over one cycle was too small for observation, so the average for a large number of cycles was observed. Usually the time for the amplitude to fall to half value was observed; thus, with a modified form of equation (1):

$$\lambda = \frac{1}{ft} \log \frac{A_0}{A_n},$$

where $A_0/A_n = 2, f$ = frequency and t = time in sec.

$$\lambda = \frac{0{\cdot}6932}{ft}.$$

Experimental Method.

Although the experimental method and apparatus employed in this work were developed with particular regard to measurements on aluminium alloys, no important modification was necessary for measurements on steel specimens. For any method of measuring damping to be successful the reduction of external friction or energy loss to a small order of the internal energy dissipation is of first importance. Secondly, it is an advantage if the specimen requires but little or no preparation, and size and shape are not too restricted. Previous investigators (5, 6, 7) have reported damping values as low as 10^{-4} at low stresses, and, as their technique permitted considerable latitude in the choice of specimen, this being a freely-suspended resonahe bar, development along these lines was selected. This is in contrast to the Föppl-Pertz technique, which needs a rather carefully prepared specimen of prescribed dimensions, which forms only part of the vibratory system. Moreover, this latter technique is not particularly well adapted for measurements in the region of very low stresses, which, as has been indicated, afford a damping value independent of stress.

The experimental method adopted consists in supporting a bar of the material in such a manner as to facilitate its executing vibrations of a "free-free" character. The bar is forced into oscillation at one of its natural frequencies by an electromagnetic method, and the existence of such oscillation detected similarly. The determinations are made by noting the time required for the vibration amplitude to fall to half of an arbitrary initial amplitude subsequent to cutting off the driving power.

Freely-suspended cylindrical or prismatic bars are capable of oscillation in three principal modes, *viz.*, flexural, longitudinal and torsional. For

reasons connected with the elimination of air damping and suspension losses, only torsional vibrations were found to be of practical use. A full discussion of all these and other matters connected with the apparatus and method will be found in an earlier paper by the present authors.[4]

The Apparatus.

The Suspension System.

The success of any damping measurement may be said to rest very largely upon the method of suspension of the specimen, consequently particular care was devoted to this matter in this work.

Fig. 1 illustrates the support stand and suspension system employed, and it will be shown to conform to two requirements: (*a*) That energy

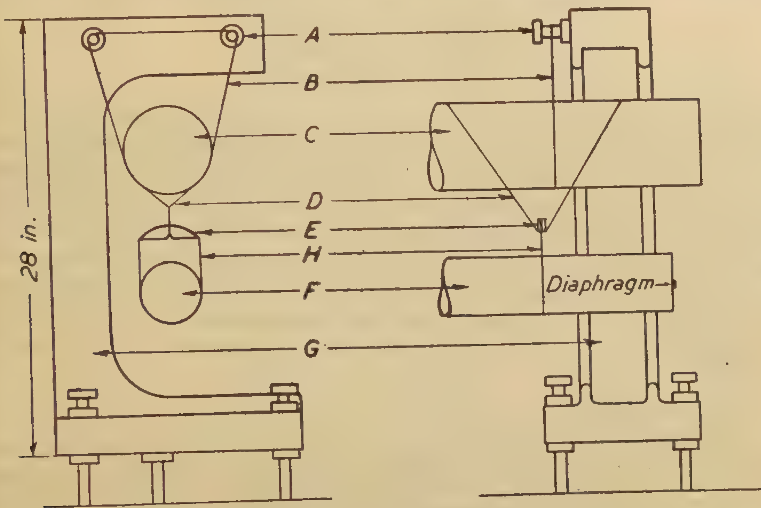

FIG. 1.—Support Stand and Suspension Details.

loss may be minimised by connecting the vibrator to a much larger mass through a "weak elastic coupling," and (*b*) to arrange for a minimum of restraint of the vibrator in its oscillatory motion. In the figure the specimen F rests in thin wire loops H which pass round the lower half of the specimen and over the balance beam E. This loop is made by interlinking two small loops formed at each end of the wire. This link and the associated "twists" nest in a groove in the top curved surface of the balance beam; the weight of the specimen thus holds the link against the beam and so prevents the twists from slipping undone.

The next stage is the V-suspension D, the apex of which passes under the balance beam, whilst the loops which form part of the sides of the V pass round the steel inertia bar C. Finally, the inertia bar C is suspended by 7-stranded wire cable made into loops B which pass over the two bolts A in the cast-iron stand G. Two of these stands are necessary for suspension purposes. The wire used for the loops H and the V-suspension D was 0·015-in. dia. steel piano wire.

A most valuable feature of this suspension system is the balance beam illustrated in Fig. 2. This component allows the specimen freedom to oscillate through a small angle during torsional oscillations. The balance beam oscillates about the apex of the V-suspension, which forms an axle for

the balance beam. The "stabilising" distance (Fig. 2) is an empirical value, experience showing that a greater distance tends to over-stabilise the

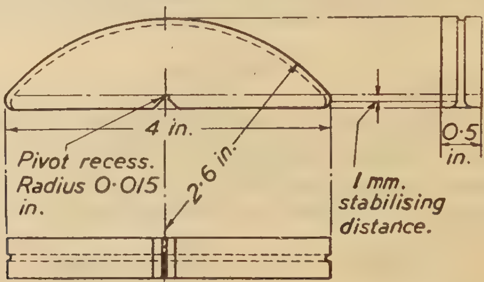

FIG. 2.—Balance-Beam Details.

system, thereby destroying the rotary freedom of the balance beam and specimen. The length of the balance beam is determined by the diameter of the specimen, a 4-in. beam serving for specimens from 3½ to 4½ in. in dia.

This feature eliminates the necessity for carefully positioning the suspension loops at the various nodes of vibration for each overtone of the specimen. At the fundamental tone a rigidly-fixed loop-suspension would be useless, whereas in this system little error is introduced, despite the fact that the loops are situated antinodally. The balance beam practically eliminates suspension loss for torsional vibrations, a fact which can be established by comparing damping measurements with the loops situated first at the nodes and then near the antinodes for a particular overtone.

The Electrical Equipment.

The equipment is designed to enable vibrations to be excited and detected in a wide range of specimens over the frequency range from 100–12,000 cycles per sec. The source of power is an audio-frequency oscillator of the "beat" frequency type in which two radio-frequency oscillators are heterodyned to provide a smoothly adjustable difference frequency between 60 and 6000 cycles per sec. The most important features of this instrument are the frequency stability and the fine control of frequency. The short-frequency drift is of the order of 0·002 cycles per sec. over a period of 8 hr. The fine frequency control is a dial graduated in 100 degrees, for which the frequency change is 0·6 cycle per sec., giving 1 division = 0·006 cycle per sec. These features are important owing to the narrow response curves of average specimens, *e.g.*, for a logarithmic decrement of 10^{-5} at 1000 cycles per sec. the resonance breadth at half-amplitude is 0·0055 cycles per sec.; this also represents a half-amplitude decay-time of 69·3 sec. An electronic frequency doubler has been incorporated to increase any oscillator output frequency by a factor of two. This feature is useful both for extending the working range and for identifying vibration modes in bars, *e.g.*, separating the harmonic torsional series from the inharmonic flexural. The audio-oscillator is connected to a power amplifier with an output of 12 watts, which feeds the electromagnetic-exciter unit. For steel specimens the pole tips of a polarised electromagnet are associated with, but not touching, a small block of low-hysteretic alloy (Radiometal) which is securely stuck to the end face of the cylindrical specimen by means of a shellac-based cement called Cenco-Sealstix, indicated by the word "diaphragm" in Fig. 1. When the

frequency of the current in the electromagnet is within the resonance range of one of the natural frequencies of the specimen, vibrations are produced therein by the alternating magnetic force on the Radiometal block.

The vibrations are detected by an instrument identical to the exciter and which is connected to a sensitive four-stage amplifier of conventional design; to the output of the latter is connected a robust copper-oxide rectifier voltmeter for indicating vibration amplitudes. A cathode-ray oscilloscope is connected across the output of the detector amplifier for observing whether the wave-form of vibration is monochromatically sinusoidal and to check that the vibration frequency of the specimen coincides with the drive frequency; sometimes a given driving frequency will excite its overtone in the specimen.

Method of Measurement.

The specimens were measured at all the torsional tones which were within the frequency range of the oscillator and which were capable of being excited to a sufficient amplitude by the available power.

After setting the oscillator to the resonance frequency of the specimen the fine tuning control is adjusted until a maximum vibration amplitude is secured, as indicated by the output voltmeter. The power is then cut off and the half-amplitude decay-time noted by a stop-watch; when this is done a further evaluation can be made by increasing the sensitivity of the detector amplifier so that full scale on the outer meter is again secured. In this way the constancy of the damping over the stress range may be assured by reason of the half-amplitude decay-times being equal for a given decay of the oscillation.

Specimens.

In this brief study six cylindrical bars, each 3 in. in dia. and 3 ft. long, were subjected to successive cycles of heat treatment, and damping measurements made at each stage in the cycle.

All six specimens were from the same cast with analysis as follows :

Carbon.	Manganese.	Silicon.	Sulphur.	Phosphorus.
0·59%	0·75%	0·23%	0·48%	0·044%

The bars were marked 1, 1*A*, 2, 2*A*, 3, 3*A*.

According to the manufacturer's report on these bars, magnetic etching at each end did not reveal the existence of hairline cracks, while examination by microscopic etching declared them free from micro-cracks. The examination by sulphur-printing did provide evidence of sulphide inclusions towards the centre of the bars, increasing in severity from No. 2 to No. 3*A*. The extent of this central segregation was not severe, but only bars No. 1 and No. 1*A* could be described as commercially perfect. The bars were received from the manufacturers in the normalised state, the normalising temperature being 840° C.

Experimental Results.

All specimens were measured first in the normalised condition as received, after which they were quenched from 830° C. in oil at temperatures between 23° and 33° C. Owing to the possibility that internal stresses resulting from the quenching might influence the damping, the specimens were given a mild temper at 250° C. in order that stresses might be relieved without essentially altering the structure. The damping values at the fundamental tone for three of the specimens are presented graphically in Fig. 3, the heat treatments subsequent to those mentioned above also being indicated.

Table I. shows all the damping measurements made on each specimen

TABLE I.—*The Effect of Heat Treatment on the Damping Capacity of 0·6%-Carbon Steel.*

Specimen No. :	1				1A				2			
Harmonic No. :	1	2	3	Brinell Hard-ness Number.	1	2	3	Brinell Hard-ness Number.	1	2	3	Brinell Hard-ness Number.
Frequency range. Cycles per sec. :	1725–2220	3450–4460	5175–6680		1675–2140	3350–4300	5020–6440		1650–2100	3300–4230	4950–6340	
Sequence of Treatments.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.		$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.		$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	
As received. Normalised at 840° C.	0·89	0·95	1·10	230	0·89	0·92	0·97	232	0·875	0·93	0·99	236
(a) Oil-quenched from 830° C. and tempered at 250° C., pickled .	0·70	0·77	0·77	301	0·78	...	1·05	281	0·64	0·73	0·80	300
(b) T. 650° C., pickled . . .	0·44	0·50	0·55	242	0·54	...	0·70	243	0·52	0·61	0·61	243
(c) Normalised at 800° C., pickled .	0·73	0·82	...	223	0·66	0·75	...	221	0·57	0·64	...	218
(d) Quenched from 830° C., pickled .	0·72	0·80	...	309	0·75	0·78	...	307	0·71	0·79	...	309
(e) T. 650° C., pickled . . .	0·50	0·64	0·65	222	0·49	0·54	0·60	222	0·49	0·55	...	223

Specimen No. :	2A				3				3A			
Harmonic No. :	1	2	3	Brinell Hard-ness Number.	1	2	3	Brinell Hard-ness Number.	1	2	3	Brinell Hard-ness Number.
Frequency range. Cycles per sec. :	1725–2200	3420–4410	5130–6630		1700–2190	3395–4380	4090–6565		1720–2220	3435–4435	5145–6650	
Sequence of Treatments.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.		$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.		$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	$\lambda \times 10^{-4}$.	
As received. Normalised at 840° C.	0·86	0·91	0·96	231	0·84	0·90	0·94	233	0·82	0·90	0·93	229
(a) Oil-quenched from 830° C. and tempered at 250° C., pickled .	0·71	0·72	0·77	301	0·72	0·80	0·84	300	0·72	0·81	0·77	294
(b) T. 650° C., pickled . . .	0·47	0·52	0·56	243	0·44	0·50	0·52	243	0·53	0·65	0·71	244
(c) Normalised at 800° C., pickled .	0·68	0·75	...	215	0·68	0·77	...	223	0·67	0·75	...	222
(d) Quenched from 830° C., pickled .	0·70	0·74	...	304	0·72	0·78	...	305	0·75	0·81	...	306
(e) T. 650° C., pickled . . .	0·46	0·53	...	221	0·50	0·65	...	223	0·51	0·60	...	222

and, wherever available, the second and third harmonic measurements. It should be noted that the frequency range quoted for each harmonic

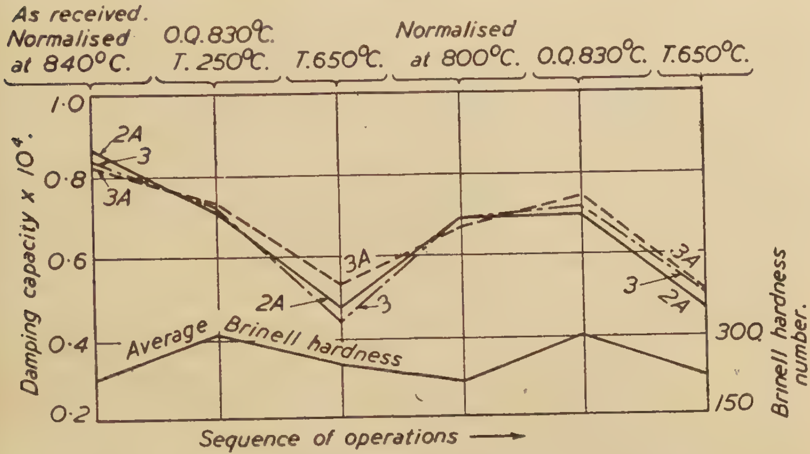

FIG. 3.—Influence of Heat Treatment on Damping Capacity.

arises from the change of natural frequency through successive shortening of the specimens after each heat treatment, due to taking a portion for micro-examination and Brinell hardness tests.

The Brinell hardness values are here regarded as quantitative checks on the metallurgical condition defined by the heat treatment. It may be mentioned that in all cases the microscopic examinations at each stage confirmed the structure normal to this material.

It will be noted that the dampings show an increase of up to 30% as a function of frequency. This is not regarded as evidence for a genuine effect of frequency but seems due to experimental difficulties which were somewhat increased by the technique then available. The principal difficulty arose in the measurement of the rather short decay-times afforded by the higher tones at such damping values. To illustrate this the measured half-amplitude decay-times for bar No. 3 after the first tempering at 650° C. are given in Table II.

TABLE II.—*Half-Amplitude Decay-Times, Bar No.* 3.

	Frequency. Cycles per sec.	Half-Amplitude Decay-Times. Sec.
First harmonic . . .	1980	7·9 8·0 8·0 7·9 7·7 8·1 8·1
Second harmonic . . .	3960	3 5 3·65 3·5 3·6
Third harmonic . . .	5940	2·25 2·25 2·25

These decay-times represent a damping of $0{\cdot}44 \times 10^{-4}$, $0{\cdot}5 \times 10^{-4}$ and $0{\cdot}52 \times 10^{-4}$, respectively. The decay-time measurements quoted in Table II. for the second and third harmonics were obtained by observing the time taken for the vibration amplitude to fall from full scale to quarter scale on the output meter. Thus, time values of 7·0 sec. and 4·5 sec. were actually observed for the second and third harmonics, respectively. This procedure

minimises observer errors, but is only justified where the damping is independent of the stress.

The half-amplitude decay-times quoted for the first harmonic have a special interest, as the first three values were obtained successively from a single decay of the oscillation, and similarly for the last four values quoted. Clearly, the damping as expressed by the logarithmic decrement is a constant for this range of amplitudes, which subsequent tests and deductions have indicated to correspond to a maximum surface-fibre stress of at least 100 lb. per sq. in. The decay-times quoted for the second and third harmonics are isolated values, each from a separate decay of the oscillation, thus it is not possible to say whether the damping values are affected by a higher initial damping due to stress. It is thought unlikely that a stress higher than that obtained at the first harmonic could have been obtained, as the detected voltage was much smaller, and, if the detector were considerably less efficient at the higher frequency, so also would be the exciter, and thus the same power would produce a smaller stress in the specimen.

An important factor in the reliability of the measurements concerns the surface condition of the specimens. Owing to the surface of the bar being the region of maximum stress in torsional vibrations, it is evident that the surface condition will be most favourably situated to modify the damping, thus the removal of surface scale or rust is most important. With the exception of the tests when received, in the normalised condition, the bars were always pickled in a 10% solution of sulphuric acid to remove scale after heat treatment, this being followed by a benzine wash. As a matter of interest the specimens were measured before pickling, as received from heat treatment. These tests showed that the heterogeneous scale due to the oil quench increased the measured damping to a value between two and three times that quoted in Table I. The influence of the pickling after the tempering and normalising treatment was less decisive, the decrease in damping varying from zero to 30%. It might be thought that the surface-condition problem would be solved by machining the surface of heat treatment; this, however, introduces the possibility of "cold-work" at the surface. Moreover, as a large number of heat treatments were contemplated, the successive reduction of the diameter would have increased the experimental difficulties, apart from the possibility of there being a genuine effect of diameter due to a significant radial distribution of metallurgical structure. The close agreement of the bars at the fundamental harmonic in each metallurgical state is considered to justify the procedure adopted.

Discussion of Results.

The most salient feature arising from this work is the reproducibility of the values for the quenched and tempered states, affording in each cycle an average damping of $0{\cdot}7 \times 10^{-4}$ for quenched, and $0{\cdot}5 \times 10^{-4}$ for tempered states. The Brinell hardness shows similar but less marked changes for these states The temper at 250° C. can hardly have influenced the damping due to the first quench, since the Brinell hardness, and therefore the structure, were not modified.

Turning to the tests in the normalised state it will be seen that the normalising at 840° C. afforded an average damping of $0{\cdot}84 \times 10^{-4}$, whilst the normalising at 800° C. gave an average damping of $0{\cdot}68 \times 10^{-4}$. It seems likely that the higher normalising temperature was responsible for this difference, which may be due, chiefly, to contrasting metallurgical conditions near the surface. It should be mentioned that the specimens were not pickled in sulphuric acid on being received as this did not seem necessary; the surface was merely wire-brushed and washed in benzine.

It has been suggested that the variation of damping with hardness may

be connected with the state of aggregation of the carbide resulting from the heat treatments given (*see* discussion on the paper in the *Journal of the Institute of Metals* [4]).

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The authors also wish to acknowledge the assistance of their colleagues, Messrs. F. Tiffen, B.Sc., and F. F. Edwards, in carrying out the experimental work and in the preparation of this paper.

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A NOTE ON THE RELATIONSHIP BETWEEN PRELIMINARY HEAT TREATMENT AND RESPONSE TO NITRIDING OF SOME NITRIDING STEELS.*

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(Figs. 11 to 14 = Plate I.)

Synopsis.

In the nitriding of some chromium-bearing steels it has been found that the hardness of the case is influenced considerably by the preliminary heat treatment to which the steel has been subjected. This is particularly true of the pre-nitriding tempering treatment. Steels of the chromium-molybdenum and chromium-molybdenum-vanadium types are considerably affected. A chromium-molybdenum steel containing a notable amount of aluminium is scarcely affected at all in this respect. The details of the experimental work are described and illustrated, and possible explanations of the results obtained are discussed.

Introduction.

The nitrogen hardening of special steels has now been in use for about twenty years, and during this period much attention has been given to the process. Some of the early investigators compared the results obtained when annealed or normalised steel was nitrided with the nitriding of quenched and tempered steel, but once it had been established that the quenched and tempered material produced a more satisfactory article, very little further attention seems to have been given to the influence of preliminary heat treatment on the results obtained during the nitriding operation. Harder, Gow and Willey,[1] using a steel containing 0·25% of carbon, 0·91% of chromium and 0·28% of molybdenum, reported that increased hardness of the case was obtained when the carbides were in a finely divided condition such as is obtained by quenching or quenching and tempering, but their results were really a comparison between such conditions on the one hand, with annealed, normalised or spheroidised on the other. No important difference in hardness was noted after nitriding water-quenched specimens and quenched specimens tempered at 870° and 1000° F. These tempering temperatures are lower than those usual in commercial practice, and it will be shown later that appreciable differences would have been detected had the tempering temperatures used by these workers been higher; they were also using an unsuitable method of hardness determination (*viz.*, the Rockwell *B* scale). Jones and Morgan[2] observed that a steel containing about 1½% of chromium, and a similar steel containing in addition ¼% of molybdenum, appeared to develop a slightly greater maximum hardness in the nitrided layer when in the normalised condition than they did in the quenched and tempered state. After the conclusion of their experimental work, the authors came across a brief reference by Genders[3] to the influence of initial quenching temperature on the hardness and penetration obtained when steels containing tantalum and niobium were nitrided.

* Received July 20, 1944.

Preliminary Experiments.

The attention of the authors was first drawn to the possible influence of preliminary heat treatment on the hardness obtained during nitriding when they were investigating some apparent anomalies in the maximum case hardness of nitrided Cr–Mo–V steel supplied to the following specification:

Carbon. %	0·30–0·37
Manganese. %	0·45–0·65
Nickel. %	0·30 maximum
Chromium. %	1·80–2·00
Molybdenum. %	0·25–0·35
Vanadium. %	0·10–0·20

Maximum diamond pyramid hardness numbers (D.P.H.N.), determined from depth/hardness curves, were found to vary from about 740 to 890 after nitriding for 72 hr. at a temperature of 500° C.; the examination of a considerable number of nitrided specimens seemed to show that there was a relationship between the maximum hardness of the case and the hardness of the core. To investigate this further, a number of specimens were prepared as follows: A slice $\frac{3}{4}$ in. thick was cut from a $4\frac{1}{2}$-in. square billet and the slice normalised by heating to 950° C., followed by cooling in still air. Specimens $\frac{1}{2}$ in. wide were cut from the normalised slice and hardened by quenching

Table I.—*Effect of Preliminary Heat Treatment on Maximum Hardness of Case Before and After Nitriding.*

Heat Treatment.	D.P.H.N. $H_D/5$.		
	Before Nitriding.	Core, After Nitriding.	Case, Maximum.
Normalised at 950° C.	358	381	957
Normalised at 950° C. and hardened at 930° C.	558	447	927
Normalised at 950° C., hardened at 930° C. and tempered at:			
500° C.	447	447	927
530° C.	447	447	891
580° C.	423	435	908
630° C.	381	391	825
680° C.	306	299	766
730° C.	241	241	739

in oil from 930° C. Some of these specimens were then tempered for 2 hr. at various temperatures from 500° to 730° C. and, after the preparation of surfaces suitable for subsequent hardness testing, they were nitrided for 72 hr. at 500° C. The results of these experiments are given in Table I. The figures for maximum hardness of the case were obtained from depth/hardness curves (diamond pyramid indentor, 5-kg. load).

These results confirmed the original conclusion that the hardness attained by nitriding was influenced considerably by the initial hardness, or rather, perhaps, by the initial condition. The high hardness attained by the normalised specimen was in keeping with the observations of Jones and Morgan;[(2)] this specimen showed considerable secondary hardening in the core. It is well known that steels containing vanadium exhibit the phenomenon of precipitation hardening when tempered after quenching or cooling quickly. According to Abram,[(4)] the hardening from this

cause is most pronounced when the rate of cooling has been sufficiently rapid to retain the vanadium carbide in solution in α-iron but too low to cause the formation of martensite. Depending on the composition of the steel and the conditions imposed by the preliminary cooling from high temperature before tempering, the hardness/tempering curve may show a slowing-up in the rate of softening, a definite arrest during which the hardness remains constant, or even an increase in hardness.

The hardness attained during precipitation-hardening depends not only on the reheating temperature but also on the time for which the material is held at this temperature. Burton and Russell,[5] experimenting with a steel containing 0·60% of molybdenum and 0·30% of vanadium, water-quenched from 900° C., found that the tensile strength increased from 56 tons per sq. in., after tempering for half an hour at 575° C., to 66·2 tons per sq. in., after 76 hr. at that temperature. In order to determine the influence of tempering time on the hardness attained after tempering, specimens 10 × 10 × 30 mm. in size were prepared from another consignment of 2% Cr–Mo–V steel. After normalising from 950° C., oil-hardening from 930° C. and tempering for varying times at 500° C., the specimens had the following hardness numbers :

Tempering Time. Hr.	D.P.H.N. $H_D/30$.
As quenched	626
½	475
2	454
12	454
48	454
120	449

There is no evidence here of any actual increase in hardness due to tempering, but the effect of the vanadium in holding the hardness constant over a considerable period of time is obvious; the normal softening that would have taken place as a result of prolonged tempering has been balanced by the hardening caused by precipitation.

In the absence of any reference in the literature to a relationship between variations in hardness of nitrided case and the preliminary heat treatment of the steel, it appeared possible that the effect might be peculiar to steels of this type and be connected with influence of nitriding on the particle size of the precipitated vanadium carbide or with the formation of vanadium nitride. To investigate this, experiments were made on a number of steels containing vanadium, chromium, chromium and molybdenum, and chromium and vanadium. The results of this work confirmed the effect of vanadium in holding up, or even increasing, the hardness of the core during nitriding. One vanadium steel containing 0·13% of carbon and 0·64% of vanadium, for example, increased in hardness by almost 100 D.P.H.N. units during nitriding, while the hardness of the nitrided case increased from 530 to 730 D.P.H.N. as the quenching temperature was raised from 850° to 1100° C. The influence of the quenching temperature could be detected in most of the steels, although in many it was only slight. The similarity between the nitrided hardness of as-normalised and as-quenched specimens was also fairly general in all the steels examined, despite the often very considerable differences in core hardness. The fall in hardness of the nitrided case with increasing tempering temperature was, however, by no means confined to steels containing vanadium.

Detailed Examination of Selected Steels.

As a result of the interest aroused by the preliminary experiments it was decided to check the results and extend the investigation. The

steels in which the authors were particularly interested were the 1% and 3% Cr–Mo and the 2% Cr–Mo–V steels, but, bearing in mind the special applications of the original Nitralloy, it was decided to include a Cr–Al–Mo steel.

Experimental Conditions.

All the results reported have been checked, many of them severa times; in some instances as many as five separate experiments have been made to establish the position of a single value, and the hardness figures recorded are the averages for any given set of experiments. The chances of obtaining small variations in hardness in heterogeneous material like a nitrided case are considerable, particularly when using such a light load as 5 kg. in hardness testing, and the authors are of the opinion that too much importance should not be attached to small differences in the hardness of nitrided specimens unless they have been established by a suitable and satisfactory checking.

Heat Treatment and Preparation of Specimens Before Nitriding.

Specimens of 2% chromium steel *M*2530 were prepared from 4½-in. square billet, and, at the time of heat treatment, these specimens usually had a cross-section of from ½ to ¾ in. square; the remainder of the steels were heat treated in the form of round bars on the full diameter as received (details are given in Table II.).

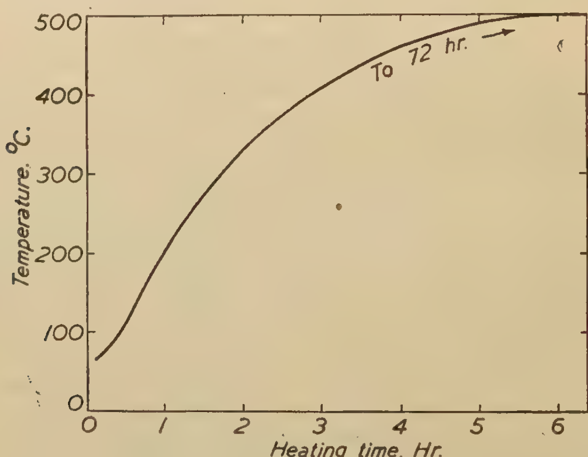

FIG. 1.—Average Rate of Heating of Nitriding Container.

Specimens were heated in an oil-fired furnace for normalising and hardening; tempering was done in an electric furnace. After heat treatment, scale was removed from the specimens by grinding; the surfaces on which hardness determinations were to be made were machined or ground sufficiently to remove any decarburised material, polished on emery cloth and emery paper, and finished on French 00 paper.

Nitriding Equipment and Procedure.

In connection with this investigation considerable use was made of a small laboratory furnace for exploratory work, but all the results reported

in the paper refer to test-pieces nitrided in commercial equipment. The capacity of the nitriding container was 70 cu. ft., and the test-pieces were placed at about the centre of this. A normal compliment of work always formed a part of the charge. The packed and sealed nitriding containers were purged for 12 hr., either with neat ammonia gas or with partially dissociated ammonia issuing from a nitriding furnace. The container, when ready for heating, was placed in a furnace already soaked at the operating temperature of 500° C. The total furnace time for nitriding was 72 hr.; a typical heating curve is shown in Fig. 1. Details of ammonia input and so-called "dissociation" figures varied slightly with working conditions, but the following are typical:

Time from Charging. Hr.	Rate of Flow of Ammonia. Cu. Ft. per Hr.	Gases, Insoluble in Water, present in Effluent Gas. %.
1	15	15
2	10	8
24	10	13
48	10	15
68	10	17

Hardness Determinations.

Hardness determinations were made with a Vickers machine using a diamond pyramid indentor. When depth/hardness curves were required, the specimen was taper-ground after nitriding, the ground surface polished with emery cloth and paper and finished on French 00 paper. The angle of taper was in the region of five degrees, but the exact angle was determined by means of a simple piece of apparatus made for the purpose. When using small loads it is not possible to make accurate hardness determinations on the matt nitrided surface. This surface was prepared by making the smallest number of passes on French 00 paper consistent with accurate measurement of the impressions. When only the maximum hardness was required, the nitrided surface was polished on worn 0 emery cloth, lubricated with paraffin until the grey surface layer was removed and then finished on dry French 00 paper.

Chemical Analysis of Steels Used.

The steels used for the second part of the investigation were of the composition shown in Table II.

TABLE II.—*Chemical Composition of Steels.*

Steel.	Carbon. %.	Ni. %.	Cr. %.	Mo. %.	V. %.	Al. %.	Cross-ction. In.
58915	0·37	0·22	1·23	0·69	...	...	$\frac{5}{8}$ (dia.)
*M*2530	0·38	0·08	1·90	0·30	0·18	...	$4\frac{1}{2}$ (sq.)
61630	0·27	0·30	2·96	0·52	...	...	$\frac{5}{8}$ (dia.)
65087	0·18	0·23	3·10	0·53	...	...	$\frac{5}{8}$ (dia.)
*M*2778	0·27	0·25	1·62	0·31	...	0·86	$\frac{3}{4}$ (dia.)

Normalising and Quenching Treatment.

In order to establish the influence of variations in the temperature used for normalising or quenching before nitriding, specimens were heated to temperatures of 850°, 950°, 1000° and 1100° C. and normalised by cooling in still air. Some of these normalised specimens were reheated to the same temperature as that used for normalising, and quenched in oil; some of the oil-quenched specimens were then tempered at 500° C. The core-hardness

figures are shown in Table III. and the maximum hardness after nitriding for 72 hr. at 500° C. in Fig. 2; all are the averages of three experiments.

TABLE III.—*Core Hardness Before Nitriding.*

Treatment.* ° C.	D.P.H.N. Hd/30.		
	1% Cr Steel 58915.	2% Cr Steel *M*2530.	3% Cr Steel 65087.
N. 850	407	406	397
N. 950	409	403	393
N. 1000	403	403	387
N. 1100	392	379	385
N. 850. O.Q. 850	...	557	444
N. 950. O.Q. 950	...	569	448
N. 1000. O.Q. 1000	...	572	453
N. 1100. O.Q. 1100	...	508	447
N. and O.Q. 850. T. 500	472	459	409
N. and O.Q. 950. T. 500	464	468	414
N. and O.Q. 1000. T. 500	474	473	416
N. and O.Q. 1100. T. 500	467	459	416

* N = normalised.

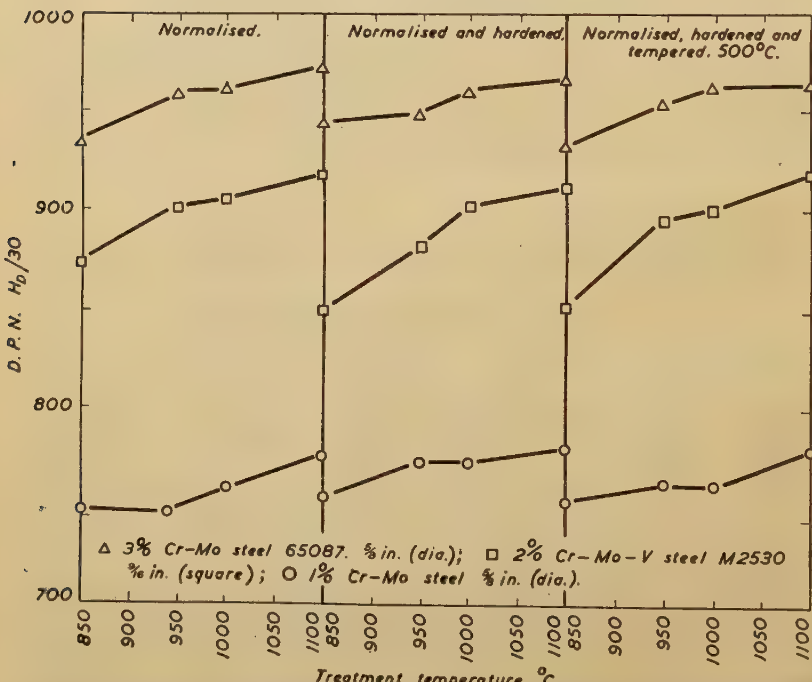

FIG. 2.—Influence of Preliminary Heat Treatment on Maximum Hardness of Case.

The maximum case hardness of all three steels after nitriding shows an upward trend as the temperature of normalising or quenching is increased, the effect being most marked in the instance of the 2% Cr–Mo–V steel. For any given normalising or quenching temperature, the steels, taken individually, show no appreciable difference in the hardness of the nitrided case, irrespective of the preliminary heat treatment (*viz.*, normalised, normalised and quenched, or normalised, quenched and tempered at 500° C.), except for the 2% Cr–Mo–V steel, which had a distinctly higher hardness after normalising from 850° C. and 950° C. than it had after quenching from the same temperatures. As the initial hardness of all the normalised specimens was lower than that of the corresponding quenched or tempered specimens, however, the total gain in hardness after nitriding was greatest for the steels in the normalised condition.

Relationship Between Tempering Temperature and Nitrided Hardness.

In the preceding experiments, normalising and quenching temperatures outside normal commercial ranges were sometimes used; in the present series, normalising and quenching temperatures were restricted to those usual in works practice. The minimum tempering temperature used was 500° C., *i.e.*, the same as would subsequently be used for nitriding. The following treatment temperatures were used :

	1% Cr–Mo Steel.	2% Cr–Mo–V Steel.	3% Cr–Mo Steel.
Normalised at . . .	950° C.	950° C.	950° C.
Oil-hardened at . . .	850° C.	930° C.	930° C.
Tempered at . . .	500° C. 600° C. 700° C.	500° C. 600° C. 700° C.	500° C. 600° C. 700° C.

The average maximum hardness of case obtained from five sets of experiments are shown plotted against tempering temperature in Fig. 3, and against core hardness before nitriding in Fig. 4. The case hardness is shown to be influenced to a marked extent by the pre-nitriding tempering temperature. The maximum hardness is obtained on nitriding untempered specimens, but the loss of hardness after tempering at 500° C. is so slight as to be negligible. The decrease in hardness after tempering at 600° C. is marked, but between 600° and 700° C. it is still more rapid. The presence of vanadium in steel *M*2530 did not appear to have any important influence on the maximum case hardness obtained as tempering temperature was increased, although there is some evidence that the vanadium present retards slightly the loss of hardness in specimens previously tempered at 600° C. As the maximum precipitation-hardening effect of vanadium might be expected to occur between 500° and 650° C., however, but not necessarily at 600° C., it was decided to make another series of experiments in which the tempering-temperature interval should be narrower.

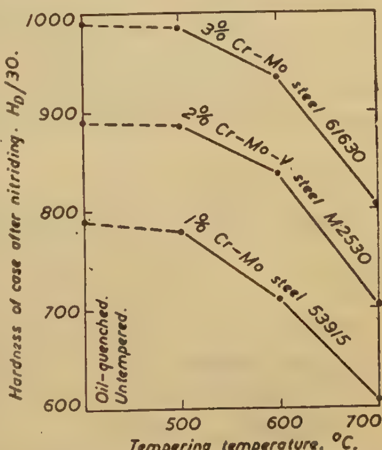

FIG. 3.—Relationship between Tempering Temperature and Hardness of Case after Nitriding.

Specimens of 2% Cr–Mo–V steel $M2530$ and 3% Cr–Mo steel 61630 were normalised and hardened, as before, and then tempered at temperatures increasing by increments of twenty degrees from 500° to 660° C.

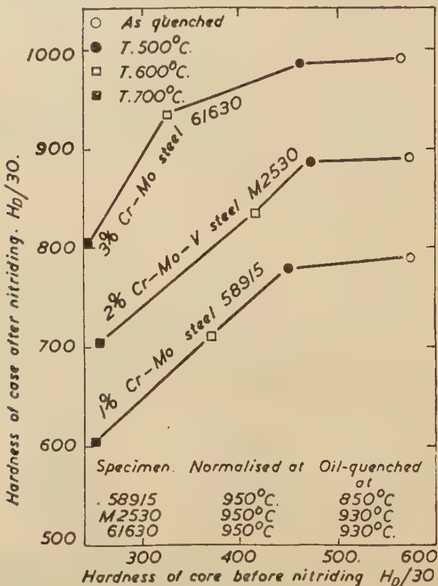

FIG. 4.—Relationship between Hardness of Steel before Nitriding and Maximum Hardness of Case after Nitriding.

Hardness/depth curves for these specimens after nitriding are shown in Figs. 5 and 6; they were plotted from the results of a single set of experiments. The maximum hardnesses of the 3% chromium steel after nitriding the as-quenched specimen, and that tempered at 500° C., are slightly lower than the values recorded previously in Figs. 3 and 4; otherwise the results of Figs. 5 and 6 agree well with those of Figs. 3 and 4. The case

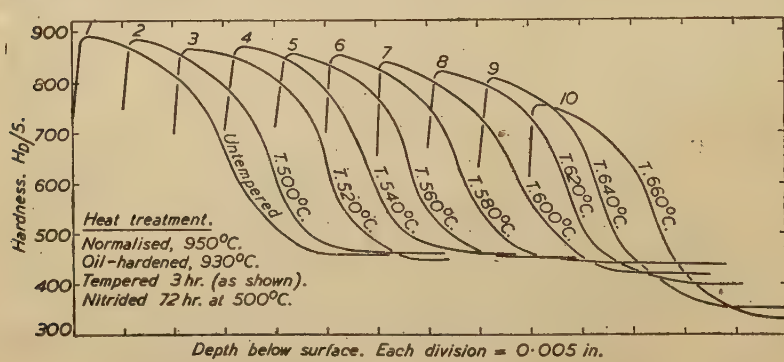

FIG. 5.—Hardness/Depth Curves for Cr–Mo–V Steel $M2530$.

produced in the vanadium-bearing steel is deeper than that of the vanadium-free steel, and the first portion of the hardness/depth curve does not fall away so steeply towards the depth co-ordinate; this, in the authors'

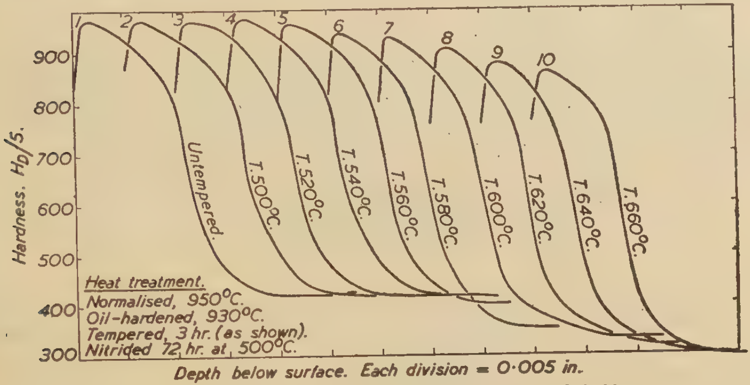

FIG. 6.—Hardness/Depth Curves for Cr–Mo Steel 61630.

experience, appears to be a characteristic of vanadium-bearing steels of the types capable of developing medium hardness on nitriding.

The experiments were repeated. The maximum hardness of the as-quenched specimens after nitriding did not agree as well as one could have wished with the first set of results, but otherwise they were satisfactory. The average maximum hardness results of the two sets of tempered and nitrided specimens are shown plotted in Fig. 7, the co-ordinates of which

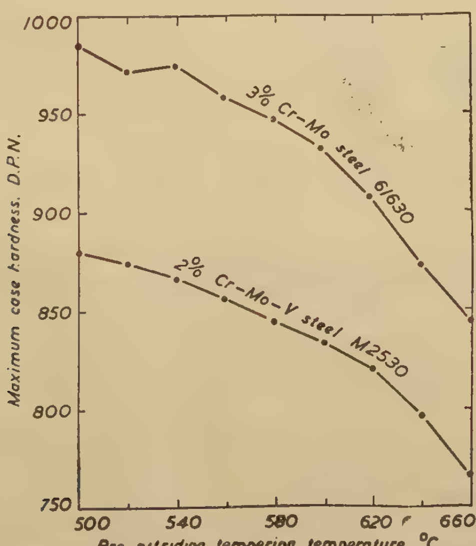

FIG. 7.—Relationship between Tempering Temperature and Hardness of Case after Nitriding.

are maximum case hardness and tempering temperatures. The slopes of the curves seem to show that the vanadium-bearing steel does not undergo so great a loss of case hardness as the 3% Cr–Mo steel, at any rate for specimens tempered in the interval 580–640° C.

Influence of Vanadium on Core Properties.

The carbide-forming elements chromium and molybdenum each give rise to precipitation-hardening effects on the tempering of quenched steel, but the amount of the added element needed to produce a given retardation

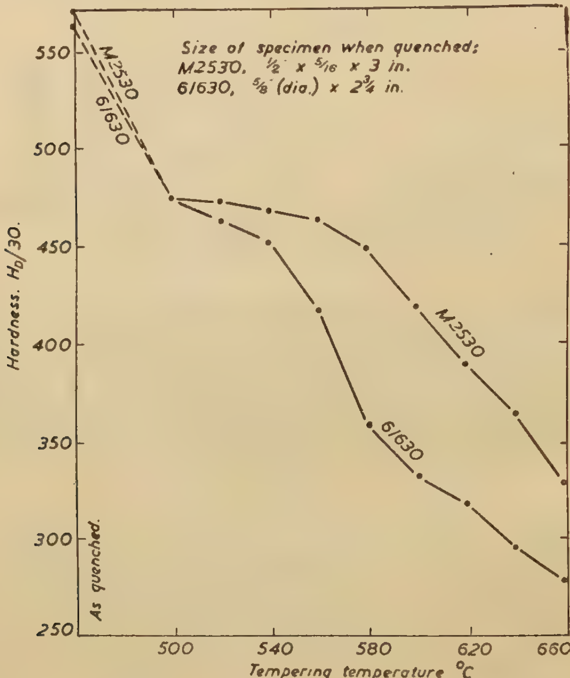

FIG. 8.—Influence of Tempering on Hardness of 2% Cr–Mo–V Steel *M*2530 and 3% Cr–Mo Steel 61630. Normalised at 950° C., O.Q. 930° C., tempered as shown. (Tempering time, 3 hr. in a furnace already soaked at the operating temperature.)

of softening, or a reversal of the softening curve, is much greater than the amount of vanadium needed to produce a comparable result. This is evident from a consideration of the results of Bain and collaborators [6] and of those of Abram.[4, 7]

A consideration of the points plotted in Fig. 4 will show that the rate of softening of the 2% Cr–Mo–V steel at 600° C. was less than that of either the 1% or 3% Cr–Mo steels; this is undoubtedly due to the influence of the vanadium present in steel *M*2530. The same effect is shown in greater detail in Fig. 8, which records the hardness results of another set of hardening and tempering experiments on two of these steels. The two steels have similar hardness values in the quenched condition, and after temper-

ing at 500° C. steel *M*2530 (with vanadium) softens more slowly than steel 61630 up to 580° C., and from this temperature the rate of hardness diminution increases, falling at a steady rate to 660° C. Steel 61630, on the contrary, softens rapidly in the temperature range 540–580° C., but in the final stage to 660° C., the effect of increasing tempering temperature is less marked than for the 2% chromium steel *M*2530. This relationship holds only for the imposed tempering conditions, and, as might be expected, the effect of increased tempering time is more marked in the instance of the vanadium-free steel; at least, so much can be concluded from core-hardness determinations made before and after nitriding (Table IV.). Nitriding the 3% Cr–Mo steel for 72 hr. at 500° C. causes appreciable additional softening of the core if the steel has been tempered at a temperature less than 580° C., whereas the core hardness of specimens made from the 2% Cr–Mo–V steel remains substantially the same as it was before nitriding throughout the tempering range 500–660° C.

Table IV.—*Core Hardness Before and After Nitriding for 72 hr. at 500° C., and Change in Hardness Caused by this Additional Tempering Effect.*

Tempering Temperature. ° C.	Core Hardness D.P.H.N. $H_D/30$.					
	2% Cr–Mo–V Steel *M*2530.			3% Cr–Mo Steel 61630.		
	Before Nitriding.	After Nitriding.	Loss or Gain in Hardness during Nitriding.	Before Nitriding.	After Nitriding.	Loss or Gain in Hardness during Nitriding.
O.Q. 930 (not tempered)	560	459	−101	564	434	−130
500	478	469	− 9	484	426	− 58
520	469	458	− 11	464	424	− 40
540	469	462	− 7	457	419	− 38
560	466	457	− 9	417	396	− 21
580	446	444	− 2	356	358	− 2
600	424	429	− 5	333	335	− 2
620	396	398	− 2	315	316	− 1
640	360	354	− 6	293	298	− 5
660	329	324	− 5	279	279	Nil

Influence of Load Applied to Indentor.

Before proceeding to a consideration of the results of the experiments it may be as well to state that the recorded variations in hardness of the case are not in any way connected with the collapse of the case due to yielding of the core during hardness determination.

Hardness/depth curves were constructed from determinations made both with loads of 5 kg. and 30 kg.; the specimens used were selected to cover a wide range of core hardness. The maxima of case-hardness/depth curves were substantially the same, irrespective of which of the two loads was used, and the slopes of the hardness/depth curves were similar for each specimen, irrespective of the load, the only important difference being that, with the heavier load, the curve was shifted slightly towards the hardness co-ordinate, a change which is the natural consequence of using a heavier indentor load on a specimen, the hardness of which is decreasing gradually from the surface inwards.

Relationship Between Core Hardness and Case Hardness.

The increase in hardness caused by nitriding hardened and tempered specimens of the three chromium steels studied here is, broadly speaking, additive; that is to say, a certain increase in hardness, characteristic of the steel, is developed by nitriding and added to the core hardness existing after nitriding. For the 1% Cr–Mo steel 58915, the hardness values of the cores of the specimens are not available, but if a loss of 55 D.P.H.N. units is assumed to have taken place during nitriding the specimens previously tempered at 500° C. and no loss for those tempered at 600° and 700° C., we arrive at the values of 350–385 D.P.H.N. units as being the range of hard-

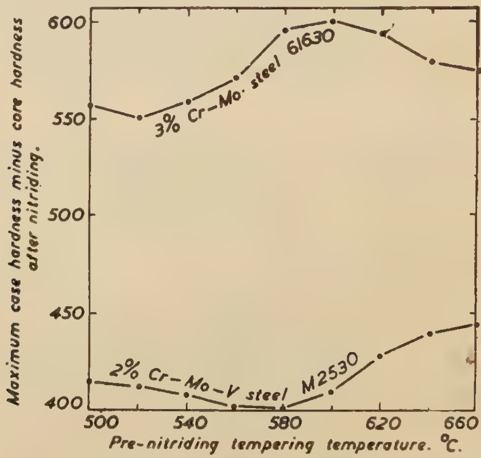

FIG. 9.—Net Increase in Hardness due to Nitriding.

ness by which the case of this steel is increased over the post-nitriding core hardness for specimens tempered in the interval 500–700° C. Similarly, the 2% Cr–Mo–V steel *M*2530, in all conditions within the same tempering range, has a case hardness which exceeds that of the core by some 400–450 D.P.H. units, the corresponding increase for the 3% Cr–Mo steel 61630 being 550–600 units. This, however, is only a generalisation, for, within these ranges, the difference between case hardness and post-nitriding core hardness does seem to vary in a regular manner, as can be seen from the curves in Fig. 9, which are plotted from the mean values of two experiments. The only unsatisfactory result is that for steel 61630 tempered at 500° C.; the spread here was 36 D.P.H. units, the actual values being:

Maximum case hardness	973	1002
Post-nitriding core hardness	426	419
Increase in hardness from nitriding	547	583

As there has been no opportunity of carrying out additional checks, however, the authors put forward these curves with reserve.

Relationship Between Core Hardness and Case Hardness of Aluminium-Bearing Steel.

In order to establish the effect of preliminary heat treatment on aluminium-bearing steel, a number of specimens were prepared from

$\frac{3}{4}$-in. dia. Al–Cr–Mo nitriding steel, the chemical composition of which is given against reference *M*2778 in Table II. Nine specimens were oil-quenched from 900° C., three of these were nitrided without tempering, and, of the remaining six specimens, three were tempered at 500° C. and three at 700° C., before nitriding. Core-hardness figures determined both before and after nitriding are shown in Table V.

TABLE V.—*Hardness Before and After Nitriding Al–Cr–Mo Steel* M2778.

Heat Treatment Before Nitriding.	D.P.H.N. Hd/30.	
	Before Nitriding.	After Nitriding.
O.Q. 900° C.	419	319
O.Q. 900° C. T. 500° C.	355	322
O.Q. 900° C. T. 700° C.	209	213

Hardness/depth relationships for the three sets of specimens, all of which had been nitrided for 72 hr. at 500° C., were determined, and the smoothed-out curves obtained from average figures are shown in Fig. 10. The maximum hardness attained does not appear to be greatly affected by considerable modifications in preliminary heat treatment or by variations in core hardness.

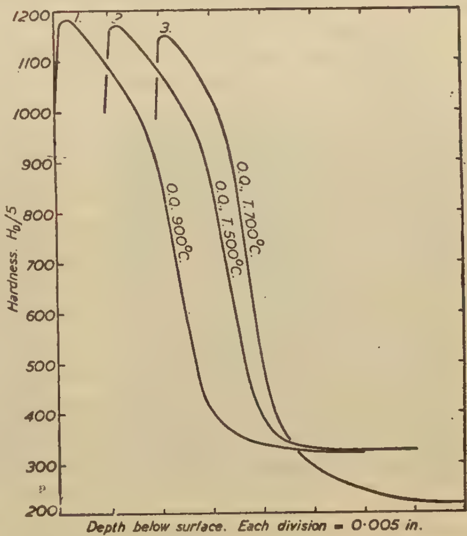

FIG. 10.—Hardness/Depth Curves for Cr–Al–Mo Steel *M*2778.

Summary and Conclusions.

The results of this investigation show that the preliminary heat treatment to which a steel has been submitted may influence considerably the maximum hardness attained during nitriding.

Two Cr–Mo steels, in which chromium was the principal element contributing to nitrogen hardening, possessed a maximum case hardness in the normalised and nitrided condition which was not substantially different from that of the same steels if quenching was substituted for normalising; tempering the quenched specimens at 500° C. did not influence materially the results obtained after nitriding. The presence of a small amount of vanadium caused the maximum case hardness of specimens normalised at 850–950° C. to be distinctly higher than in the corresponding quenched state, or quenched state followed by tempering at 500° C.

Raising the normalising or quenching temperature brought about an increase in the maximum hardness attained during subsequent nitriding.

The increase in hardness, although not very marked, was distinct for 1% and 3% Cr–Mo steels, but on nitriding specimens of a 2% Cr–Mo steel containing 0·18% of vanadium, an increase of 65 D.P.H. units resulted when the treatment temperature was increased from 850° to 1100° C.

The most marked change in the maximum case hardness of the Cr–Mo steels studied was found when the steels had been quenched and tempered; a continuous decrease in hardness occurred as the tempering temperature was raised from 500° to 700° C. The loss in hardness suffered after nitriding amounted to some 200 D.P.H. units over the tempering temperature range. Stated in another way, there is a direct relationship between the case hardness and the post-nitriding core hardness of the quenched and tempered chromium steels studied. It has been reported of chromium steels, similar to those used by the authors, that wide variations in case hardness are possible even when the imposed conditions remain constant; [8] the authors' results do not confirm this. It was found in the present work that the change in hardness follows a regular curve throughout the range of pre-nitriding tempering temperatures used, and, although it is usual to represent the hardness of the 1%, 2% and 3% chromium steels after nitriding at 500° C. as increasing in case hardness by increments of roughly 100 D.P.H. units, it is actually possible, by suitably controlling the pre-nitriding tempering temperature, to cause all three steels to develop the same hardness of nitrided case, *i.e.*, about 800 D.P.H. units. It would, therefore, be more accurate to say that the hardness of a nitrided case exceeds the core hardness by a certain value. These values for the three steels studied (in ascending order of chromium content) are approximately 375, 425 and 575 D.P.H. units under the imposed nitriding conditions.

The addition of a small amount of vanadium did not exert any marked influence on case hardness, but even in such a small amount as 0·18% it prevented appreciable softening when an oil-hardened 2% Cr–Mo steel was tempered in the range 500–560° C., and, when compared with the vanadium-free 3% Cr–Mo steel, had an even more pronounced effect on the additional softening of the core which accompanied the nitriding of the specimens previously tempered at a temperature of 560° C. or less.

The maximum case hardness of a Cr–Al–Mo steel which was examined was not appreciably less for specimens previously tempered at 700° C. than it was for specimens tempered at 500° C.

Nitriding steels are not used in the normalised condition and consequently the relationship between normalising and nitriding is not of practical importance. Similarly, practical considerations rule out the use of very high quenching temperatures during preliminary heat treatment; but, when using a nitriding steel containing vanadium, a fairly high quenching temperature assists in obtaining the best hardness from the case. The influence that the tempering temperature has on the hardness attained during the nitriding of steels, in which chromium is the principal nitride-forming element, may be used to advantage in certain practical applications. When assessing the value of such treatments, due consideration should be given to the rather poor Izod-impact figures that are obtained from vanadium steels when these have been treated in a manner intended to produce the maximum effect of precipitation-hardening; some Cr–Mo steels that have been tempered at fairly low temperatures (around 500° C.) have notched-bar impact-values that are not very impressive. There are, however, some applications for which a high notched-bar value is not of paramount importance, although a high surface hardness may be desirable; in such instances the use of a lower tempering temperature with its concomitant improvement in case hardness may be indicated. Means of economising in the use of alloys also suggest themselves.

The hardness of a nitrided case is generally assumed to be caused by

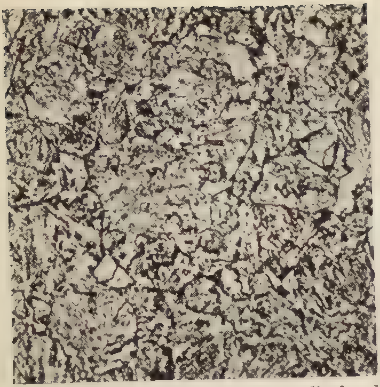

FIG. 11.—Steel *M*2530, core. Normalised and nitrided. × 500.

FIG. 12.—Steel *M*2530, core. Hardened, tempered and nitrided. × 500.

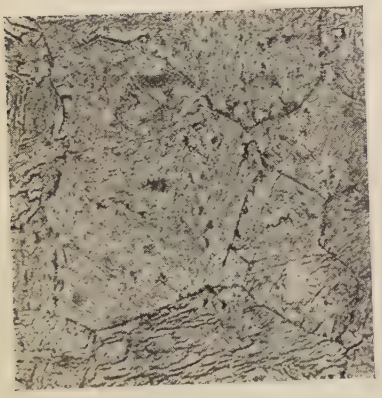

FIG. 13.—Steel 61630, core. Normalised and nitrided. × 500.

FIG. 14.—Steel 61630, core. Hardened, tempered and nitrided. × 500.

[*Hodgson and Waring*
[*To face p.* 68 P.

slip-interference, the hardest condition being attained when the combined effect of the precipitated constituents most nearly approaches that of the optimum number of particles of critical size. It is known that the carbide particles affect, to a considerable extent, the hardness that can be attained during nitriding; for example, Jones[(8)] found that a carbon-free iron containing 0·8% of aluminium did not nitride satisfactorily, as the nitrides separated in plates and the hardness was only 310 D.P.H. units. After normal nitriding the nitrogen content of the outer portion of the case of any of the usual nitriding steels is much greater than that needed to combine with the total special alloys present, even if the whole of these were combined with nitrogen. The excess nitrogen must, therefore, be present as iron nitride, but it exists in a form (*i.e.*, particle shape) different from that in carbon-free iron, either pure or alloyed, or in annealed carbon or low-alloy steel. The hardness of a nitrided case is, then, the resultant of the combined influences of the particles of iron carbide, special alloy carbides and iron and alloy nitrides, and is influenced to a considerable extent by the structure of the steel.

When normalised steel is nitrided considerable variation in hardness can be expected according to the mass of the part at the time of air-cooling for normalising. In the normalised condition the 1% and 3% Cr–Mo steels and the 2% Cr–Mo–V steel used in this investigation (in the small sections that were used for normalising) had structures that were typical of alloy steels that had been cooled at a rate slower than the critical, but too fast for the formation of pearlite—the so-called "intermediate structures."

A few of the specimens of steels *M*2530 and 61630 have been examined microscopically. In the normalised condition the microstructure of the former consists mainly of an indefinite structure, etching rather readily, with smaller areas (associated with regions of micro-segregation) that have a marked acicular structure. The structure of normalised 3% Cr–Mo steel was mainly acicular; after nitriding, the grain boundaries in the core of this steel were more clearly outlined than in the as-normalised state. Massive ferrite was not observed to be present in the small normalised specimens of either steel. In the quenched condition the specimens were martensitic. Some typical microstructures are shown in Figs. 11 to 14 at a magnification of 500 diameters:

Fig.	Steel.	Condition of Steel.
11	*M*2530	Normalised from 950° C. and nitrided for 72 hr. at 500° C.
12	*M*2530	Oil-quenched from 930° C. and tempered at 660° C., followed by nitriding for 72 hr. at 500° C.
13	61630	Normalised from 950° C. and nitrided for 72 hr. at 500° C.
14	61630	Oil-quenched at 930° C. and tempered at 660° C., followed by nitriding for 72 hr. at 500° C.

The structure of a normalised steel can be expected to influence considerably the case hardness after nitriding, depending upon how nearly this structure approaches in fineness and homogeneity that of a quenched specimen. In the present instance, the time allowed during cooling has not been sufficient to permit appreciable diffusion of the constituents through the austenite, nor for the separation of ferrite. The structure of the steel is such that it might be expected to behave during nitriding in a manner similar to that of the fully-quenched material.

The quenched specimens which were nitrided without intermediate tempering really fall into the same class as the quenched and tempered specimens, for the rate of heating used in the nitriding experiments would permit considerable tempering to take place before reaction with ammonia commenced. The slight improvement in hardness that was found to

accompany the use of high preliminary quenching temperatures may be explained by the more complete solution of the alloy carbides and the better diffusion of the alloy metals throughout the austenite. The better diffusion of the alloys should lead to a more uniform dispersion of the particles of alloy nitride. The low tempering temperature used may not have been without effect, as it would lead to a small proportion of the alloys being converted into alloy carbides, thus leaving more alloy available for nitride formation; this will be referred to later. At high tempering temperatures the influence of preliminary hardening temperature may not have been as marked, but the authors have no evidence of this.

The influence of tempering temperature was the most interesting effect of preliminary treatment disclosed during the investigation, and, practically, the most important one. The difference between the behaviour of the Cr–Mo and Cr–Mo–V steels on the one hand, and the Cr–Al–Mo steel on the other, when nitrided after hardening and tempering at high and at low temperatures, possibly depends upon whether or not the special nitride-forming elements are also carbide-forming elements. In the Cr–Mo and Cr–Mo–V group, more and more of the special elements become locked up as carbides as the tempering temperature is increased, leaving less available to form nitrides during nitriding. The aluminium of the aluminium-bearing steel is not affected in this way; all of it remains available for nitride formation. The Cr–Al–Mo steel contains carbide-forming elements also, but a greater proportion of the total nitride-forming elements remains free to form nitrides in this steel than in those of the other group.

An argument that might be put forward against this hypothesis is that, if the content of nitride-forming elements is depleted as the tempering temperature increases, then one might expect the difference between core hardness and case hardness to decrease as the tempering temperature increases, and it has been shown in this paper that this does not happen. Considerable doubt is cast on the validity of such an objection when the behaviour of the aluminium-bearing steel is recalled. Tempering this steel so as to produce considerable differences in core hardness did not affect the case hardness. In other words, in the presence of an approximately constant content of (available) nitride-forming elements, the maximum hardness developed during nitriding remains practically unaffected by variations in the core hardness of the hardened and tempered steel.

Acknowledgments.

The authors wish to express their thanks to the Directors of Leyland Motors, Ltd., for permission to publish this note, and to their colleagues Mr. A. Roskell and Mr. W. Hardcastle for painstaking work in making many hardness determinations.

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CORROSION OF STEEL SALT-BATH POTS BY MOLTEN ALKALI NITRATES.*

By R. J. BOX, B.Sc.Eng.(Met.), A.R.I.C., and B. A. MIDDLETON, B.Sc., A.R.I.C.

Synopsis.

The corrosion of steel containers of salt-bath furnaces by molten alkali nitrates was kept under observation simultaneously in several factories over two to three years. The furnaces were used for heat treatment of aluminium alloys at about 380° and 495° C., with variations in (1) the quality of steel used for the pot, (2) the shape of the pot and (3) the type of salt employed. Details of analytical control found useful are given. Results obtained indicated that with commercially available salt of guaranteed purity (sodium chloride 0·3% max.) corrosion was negligible at 380° C., and proceeded only slowly at 495° C. Where rapid attack took place the furnace design appeared to be a fairly prominent factor. For all practical purposes the use of ingot iron for the pot was a complete insurance against corrosion under the conditions described. The rate of increase of alkalinity of the melt was an index of the rapidity with which corrosion was taking place.

The work described in this paper follows on an industrial scale certain aspects of the experimental work of Lloyd and Chamberlain,† without which it would not have been possible to make serious progress. It is based on observations during several years on salt-bath furnaces in a group of factories. These salt baths are used for the heat treatment of Duralumin-type alloys operating at the following temperature ranges :

(1) Solution treatment or normalising, 490–500° C.
(2) Annealing, 360–400° C. (generally held at about 380° C.).

Equipment and Plant.

Temperature-recording equipment employed on such baths is required to have an accuracy within $\pm$ 1% of the scale reading at the temperature of operation, and this is rigidly adhered to.

All furnaces are gas-fired with certain main features of design common to all. They are fitted with welded or bolted steel casings lined with firebrick, and are fired the complete length of either side with burners at intervals of 1 ft. Between the burners and the pot a firebrick baffle is placed so as to direct the flames and hot gases upwards to a sloping throat and so down past the pot to a longitudinal ventral flue communicating at one or both ends *via* a vertical pipe to atmosphere. This design has the effect of reducing to a minimum the possibility of local overheating. Lids made of steel sheet, with lagging and designed for manual operation, are fitted in all cases.

Throughout the factories under review mild-steel or boiler-steel pots were in use. These were made from sheet with arc-welded seams and fall into three broad categories :

Type I.—Large rectangular pots made of flat sheets welded together. Ratio of depth to width slightly in excess of unity.

* Received July 7, 1944.
† *Journal of The Iron and Steel Institute*, 1940, No. II., pp. 141 P–164 P.

Type II.—Medium-size pots about 9 ft. long, 3 ft. deep and 21 in. wide. Ratio of depth to width high. Sides and bottom made from one formed sheet, with flat ends welded on.

Type III.—Medium-size pots about 8 ft. long, 21 in. deep and 2 ft. 9 in. wide. Ratio of depth to width less than unity. Construction as type II.

It is regrettable that analytical details of the metal employed were not obtained, but while the experimental work indicated that local impurities in the steel were probably responsible for a given spot being attacked by the salts, a practical solution of this aspect has already been indicated by Lloyd and Chamberlain (*loc. cit.*), *viz.*, the use of ingot iron, and since some baths were fitted with ingot-iron pots observations as to their behaviour were possible.

The salts used to fill the pots were basically of the same type, *i.e.*, equimolecular sodium-nitrate/potassium-nitrate eutectic mixtures, with one variant containing up to 14% of sodium nitrite in place of sodium nitrate. All were commercially available and, being relatively low in impurities, particularly chlorides, were suitable for use in the furnaces. It was always possible to purchase under a guarantee that the chlorine ion would not exceed 0·18%, and certain suppliers were able to improve on this at a higher cost.

No special advantage is claimed for the eutectic mixture, except that mechanical strain on the pot is reduced by the low freezing point so obtained. As already mentioned, the commercial supply contained 14% of sodium nitrite, which lowered the freezing point still further and at the same time enabled observations to be made on the effect of deliberate additions of nitrite. Lloyd and Chamberlain's experiments (*loc. cit.*) had indicated that such an addition might be advantageous in preventing pot corrosion.

Analytical Control.

Lloyd and Chamberlain's work suggested that daily analyses for the following constituents would be of assistance : Chlorine ion, nitrite (expressed as a percentage of $NaNO_2$) and alkalinity (expressed as a percentage of NaOH). It soon became apparent, however, that the nitrite figure was not essential and it was dropped; the chloride was checked only once a week, since it did not change very much. Apart from deliberate additions, nitrite salts are formed from molten nitrates, and up to a point were found to move in sympathy with the alkalinity, but the rate of formation fell off rapidly and the test was therefore dropped. Checks on chloride were, however, maintained, mainly with the view to covering the possibility of salts with high chloride content coming into use. Tests on bath samples were used rather than on the salts as delivered, owing to the formidable problems involved in sampling 1-cwt. bags of salt received mixed ready for use.

The analytical results obtained are of an irregular nature and rarely form a smooth curve if plotted graphically, which is the most informative way of recording them. Amongst possible causes of this the following are thought worthy of note :

(1) Analytical errors, which are significant because of the small amounts sought.

(2) Regular topping-up with fresh salts to make good drag-out losses.

(3) Failure to obtain a representative sample.

No attempt was made to deal with cause (1), since the standards employed were those commonly maintained, and more refined methods were out of

the question during war-time; besides, it was not considered to be as serious as causes (2) and (3). Topping-up has a considerable effect, particularly where shift work is in force. In one case additions of 1 cwt. of salt were made per 24 hr. on a bath holding 7 tons of salt, and these heavy additions effectively masked a rapid rise in alkalinity. Taking a representative sample from a molten bath presents no problems, but evidence was obtained that segregation occurred on solidification and the whole of the sample had to be used for analysis. As a result it was necessary to reduce the weight of the sample to about 40 g., dissolve it in 500 ml. of distilled water and take suitable volumes for each determination.

The methods employed for analysis were those in general use, *viz.*:

(1) Nitrite: Lunge's method as given by Treadwell and Hall * and modified by Lloyd and Chamberlain.

(2) Chloride: Titration with silver nitrate solution.

(3) Alkalinity: Titration with $N/20$ HCl, using B.D.H. Universal indicator.

Before leaving the general discussion of salts it should be noted that restrictions in the supply of potassium nitrate lead to all baths going over to sodium nitrate with or without additions of up to 7% of sodium nitrite. The standards for maximum impurities have been maintained, however.

Observations.

In the record of observations which follows, each factory has been dealt with separately and, apart from details of design peculiar to the plant under discussion, analytical results obtained by the above methods and pot corrosion are considered.

Factory A.

Salt Baths.—Manufactured and installed by supplier X; one annealing and one normalising bath.

Pots.—Type I., made of boiler plate.

Tempertaure Recorder.—Base-metal couple and sensitive galvanometer graduated in ° C.

Thermostat.—Base-metal couple operating motorised gas valve through relay.

Basket.—Wire mesh on angle-iron frame.

Salts.—Initially a eutectic mixture of sodium and potassium nitrates, but modified after a short time by an addition of sodium nitrite. The salts were purchased under a guarantee that the chlorine ion content could not exceed 0·18%. Analyses some months after starting up:

	Chloride. %	Sodium Nitrite. %	Alkalinity as NaOH. %
Normalising bath .	0·13	5·86	0·17
Annealing bath .	0·14	6·27	0·30

Bath History.

Normalising Bath.—During the first three months (*see* Fig. 1, covering this period of a total life of thirteen months) the alkalinity rose at a uniform rate from 0·1% to about 0·5% and then levelled off for about three months. Thereafter it fell gradually to about 0·35% during a further four months, when a new basket was installed. The angle-iron framework of the old basket was heavily corroded, but the wire mesh had not been attacked at all. Apart from one or two points in the sides, the pot had not been attacked. These points were located about half-way up the sides and near one end of the pot. The depth of penetration was

* "Analytical Chemistry," 8th edition, p. 570. London, 1935: Chapman and Hall, Ltd.

small compared with the thickness of the pot wall (1 in.), and it was put back into use again.

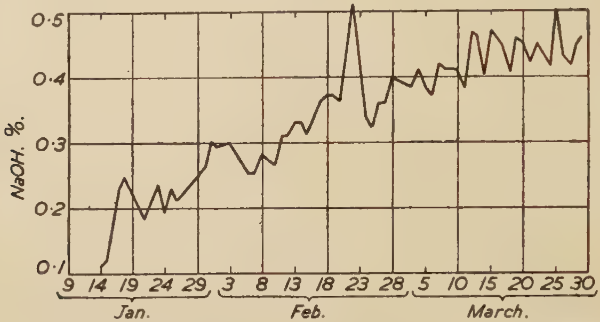

FIG. 1.—Variation of the Alkalinity of the Normalising Bath at Factory *A*.

After further service amounting to thirteen months in all the pot was replaced by a new one, as it had been in use continuously on a 24-hr. basis. The corrosion already noted had progressed further, penetrating to about one-third of the wall thickness.

The new pot perforated in three months, although there was no significant change in alkalinity. Investigation suggested that the increase was probably masked by heavy topping-up. Quantities of 1–2 cwt. were being added daily to a bath holding 7 tons of salt.

The old pot was descaled inside and out and re-installed. After a second term of six months the salt broke through, thus ending a working life of about nineteen months.

In both pots perforation occurred about half-way up the side of the pot, roughly midway between the centre and the end.

At the time of the second failure a second normalising bath was started, of identical design to the first one, but made of ingot iron, filled with sodium nitrate + 7% sodium nitrite. Analytical details of essentials at the commencement were : Chloride, 0·09%; sodium nitrite, 6·2%; alkalinity as NaOH, 0·03%.

During a period of eight months the alkalinity rose gradually from 0·01% to 0·08%; an inspection of the interior after that time afforded no evidence of heavy corrosion such as would lead to perforation, but only a loosely adherent fine red oxide of iron was observed. To date this pot has been in use day and night for twenty months without failure or significant rise in alkalinity.

Annealing Bath.—The alkalinity rose gradually from 0·01% to about 0·35% in 2½ years, and at no time was serious corrosion leading to perforation observed.

Factory B.

Salt Baths.—Manufactured by supplier *Y* and installed by factory staff, one normalising (No. 1) and one annealing (No. 2) bath.

Pots.—Type II., made of mild steel.

Temperature Recorder.—Mercury-in-steel thermometer.

Thermostat.—Liquid in metal bulb operating gravity gas-pressure valve.

Basket.—Wire mesh on angle-iron frame.

Salts.—Low-impurity eutectic mixture. Analyses at an early stage :

	Chloride. %.	Sodium Nitrite. %.	Alkalinity as NaOH. %.
Normalising bath .	0·13	4·24	0·33
Annealing bath .	0·13	3·66	0·35

Bath History.

As the baths were repeatedly changed over from normalising to annealing and *vice versa*, they are dealt with together. In both the alkalinity rose fairly rapidly to 0·3–0·4%, with the normalising bath (No. 1) slightly in the lead. From this figure there was no significant move up or down, except that when the functions of the two baths were changed over after about six months their relative positions with respect to alkalinity became transposed during the next three months. Fig. 2

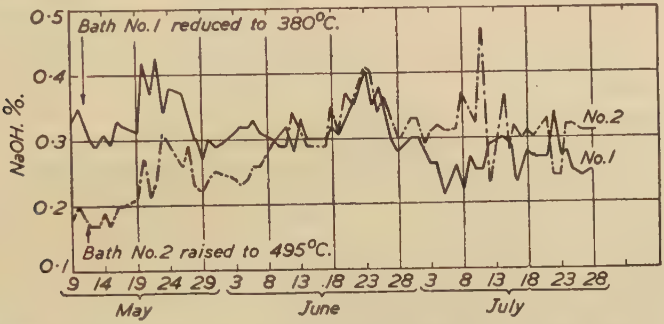

FIG. 2.—Variation of the Alkalinity of Baths Nos. 1 and 2 at Factory *B*.

shows this latter period. The somewhat abrupt rise noted at the commencement led to an inspection of the normalising pot. It was found to be free from corrosion, but the iron framework of the basket had been attacked.

After two years' service the then normalising pot perforated without any change in alkalinity. Corrosion was found to have taken place over a fairly large area about half-way up the rear wall of the pot near one end.

Factory C.

Salt Baths.—By supplier *Y* and installed by factory staff; two normalising baths.

Pots.—Type II., made of mild steel.

Temperature Recorder.—Base-metal thermocouple operating sensitive galvanometer, graduated in ° C.

Thermostat.—As at Factory *B*.

Basket.—As at factory *B*.

Salts.—As at factory *B*. Analyses at an early stage:

	Chloride. %	Sodium Nitrite. %	Alkalinity as NaOH. %
Normalising bath No. 1	0·09	1·25	0·25
Normalising bath No. 2	0·13	1·50	0·22

Bath History.

With both baths working at normalising temperatures the loading was not always heavy enough to justify more than one in action for considerable periods. However, their behaviour was similar, the alkalinity in both rising gradually to about 0·25% during nine months, from which point it levelled out for about a further twelve months. No. 2 bath was then emptied, cleaned out and filled up with fresh salt. Following this the alkalinity did not rise above 0·05% in three to four months.

Routine overhauls at intervals of three months were instituted from the beginning, involving removal of salt and descaling inside and outside. At no time was heavy corrosion of the pots observed, although the iron frame of the basket was attacked, necessitating various lengths being replaced.

Factory D.

Salt Baths.—By supplier *Y* and installed by factory staff; one normalising and one annealing bath.

Pots.—Type II., made of ingot iron.

Temperature Recorder.—As at factory *B.*

Thermostat.—As at factory *B.*

Basket.—Wire mesh on stout wire frame.

Salts.—As at factory *B.* Analyses at an early stage:

	Chloride. %	Sodium Nitrite. %	Alkalinity as NaOH. %
Normalising bath .	0·11	1·89	0·07
Annealing bath .	0·10	0·94	0·03

Bath History.

The alkalinity of the normalising bath rose slowly over two months to about 0·25%, from which point it gradually fell during two years to 0·05%. Charts were not kept for the annealing bath, as neither the magnitude nor the change of the alkalinity were significant. During the life of both pots to date no signs of corrosion were observed; the baskets were not attacked either.

Factory E.

Salt Baths.—By supplier *Z* and installed by factory staff; one normalising and one annealing bath.

Pots.—Type II., made of mild steel.

Temperature Recorder and Thermostat.—Base-metal thermocouples operating potentiometric instruments.

Baskets.—Wire mesh on angle-iron frame.

Salts.—As at factory *B.* Analyses at an early stage:

	Chloride. %	Sodium Nitrite. %	Alkalinity as NaOH. %
Normalising bath .	0·17	6·38	0·39
Annealing bath .	0·14	1·64	0·10

Bath History.

Normalising Bath.—The alkalinity rose gradually during six months to about 0·4% and corrosion of the angle-iron basket frame was observed but not of the pot. About twelve months after starting up, however, an accelerated increase in alkalinity was noted and inspection of the pot confirmed advanced corrosion at one point low down in one side. The

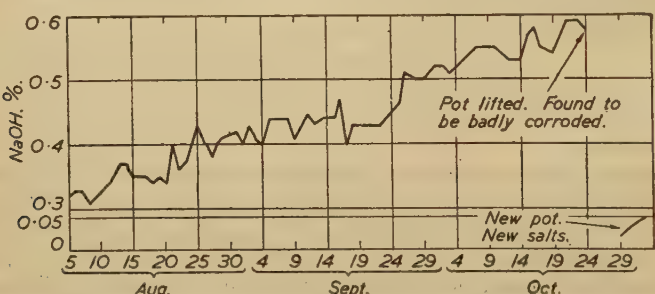

FIG. 3.—Variation of the Alkalinity of the Normalising Baths at Factory *E.*

pot was therefore changed after a life of eighteen months. Fig. 3 shows the last three months of its life and the commencement of the second pot. This showed a fairly steep rise in alkalinity during the first two months, which then flattened out somewhat in the neighbourhood of 0·4–0·5%.

Later another rise set in, but no evidence of corrosion was apparent on probing without emptying the salt.

Annealing Bath.—In this bath the alkalinity rose gradually to 0·2% in three months and remained at that value for two years.

Factory F.

Salt Baths.—By supplier *Y*, installed by factory staff; one normalising and one annealing bath.

Pots.—Type III., made of mild steel.

Temperature Recorder and Thermostat.—As at factory *C*.

Basket.—Expanded-metal mesh with angle-iron frame.

Salts.—Eutectic mixture; in normalising bath 14% of sodium nitrate replaced by sodium nitrite, in annealing bath sodium nitrate with 5% of sodium nitrite. Analyses at an early stage:

	Chloride.	Sodium Nitrite.	Alkalinity as NaOH.
	%.	%.	%.
Normalising bath .	0·17	8·73	0·92
Annealing bath .	0·16	3·00	0·31

Bath History.

Normalising Bath.—The first pot was heavily corroded after about two weeks and was replaced although perforation had not actually taken place. Extensive corrosion had occurred, but it was confined to the two sides, extending longitudinally at two levels, one just below the surface of the molten salt and the second near the bottom bend. No rapid rise in alkalinity was noted, but the chloride figure was rather higher than usual at 0·2%.

During the life of the second pot the alkalinity increased gradually from 0·01% to 0·3% over five months, when it levelled off. Nine months after starting up corrosion was observed visually, and the bath was shut down for replacement.

A third pot was installed which had been made by welding-up ½-in. constructional-quality steel sheet, and in two months the alkalinity shot up from 0·01% to 0·48%, whereupon the pot was replaced owing to the imminent danger of perforation.

The next and to-date final replacement was made of ingot iron, and, although the alkalinity rose steeply from 0·05% to 1·0% in six months and then levelled out, no corrosion was observed. The iron framework of the basket, however, was severely attacked and had to be replaced. The ingot-iron pot has been in continuous use day and night since installation without any signs of corrosion and latterly the alkalinity has fallen slowly to about 0·6%.

Annealing Bath.—While normally operating at annealing temperatures, this bath was brought up to normalising heat when shut-downs occurred on the normalising bath. The alkalinity commenced at 0·02%, but within a few days the bath was raised to normalising heat with the result that the alkalinity moved sharply upwards to 0·1% during eleven days. On reverting to annealing temperatures it remained at that value for about seven weeks, after which it was again worked at normalising temperature. At once the alkalinity commenced to rise, reaching 0·18% in rather more than a week; Fig. 4 shows the two sharp rises with the quiescent period intervening. By this time signs of heavy corrosion were seen, and during a further nine months at annealing heat the alkalinity rose gradually to 0·38%.

At this time an ingot-iron pot was installed with a charge of new salts, of the same composition, *i.e.*, sodium nitrate with an addition of 5% of sodium nitrite. The alkalinity at the commencement was 0·2% and over four months it gradually rose to 0·33%, at which figure no further increase was noted for three months. From then on a slow fall set in to 0·28% in another three months.

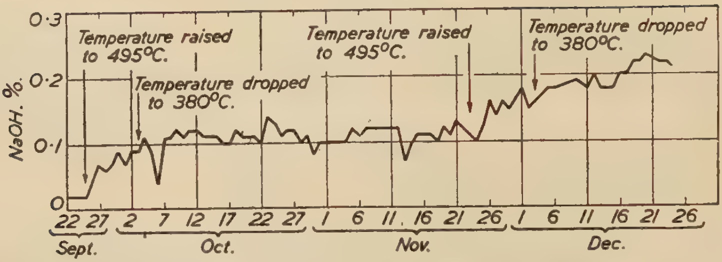

FIG. 4.—Variation of the Alkalinity of the Annealing Bath at Factory *F*.

CONCLUSIONS.

From the observations recorded above it is felt that certain conclusions may be drawn, but they are put forward in a tentative way, since a large number of interfering factors are at work which under different circumstances might have a material effect. The following, however, do appear worthy of note so far as the installations under discussion are concerned :

(1) Boiler-plate or mild-steel pots are liable to corrosion with any commercially available supplies of salts, although a high-quality salt may postpone perforation, giving a life of one to two years.

(2) Welded seams are not attacked preferentially when the design of the pot places them at the ends only.

(3) Slight local hot spots due to impingement of flame and/or hot gases appear to be a secondary factor in promoting corrosion. In this respect the shallow type of pot represents a weakness in design, conducive to a more marked tendency to hot spots than the deeper type.

(4) Annealing temperatures are not likely to lead to corrosion of any pot material in general use, and where this has commenced at normalising temperature it does not proceed to any significant extent upon dropping to annealing temperatures.

(5) There is no evidence to support the view that additions of sodium nitrite postpone or prevent corrosion by affecting the equilibrium of the reversible reaction :

$$2NaNO_3 \rightleftharpoons 2NaNO_2 + O_2.$$

(6) The rate of rise of the alkalinity is worth observing, since it frequently gives prior warning of corrosion. It can be masked by heavy topping-up and is not always obvious where corrosion is very slow. Persistent insertion of greasy parts will also cause a rise in alkalinity.

(7) Periodic overhauls at intervals of, say, three months are advisable but do not guarantee freedom from trouble. The salt should be baled out, but all except the discoloured portion may be used again. The pot should be lifted and cleaned, but descaling of the interior is not recommended unless the corrosion is heavy, when the scale should be chipped away to ascertain the extent of the damage.

(8) Pots made of ingot iron would appear to be a worth-while insurance against pot failures with commercially available salts containing less than 0·18% of chlorine ion.

(9) Where parts of the basket are attacked it is probable that the pot is protected by a negative polarity. The results of deliberately making the pot and basket negative with respect to the melt would appear to be worth some attention.

THE EFFECT OF OVERHEATING ON THE TRANSFORMATION CHARACTERISTICS OF A NICKEL-CHROMIUM-MOLYBDENUM STEEL.*

By K. WINTERTON, B.Sc., Ph.D. (Birmingham University).

(Figs. 6 to 10 = Plates II. and III.)

Synopsis.

The S-curve for a nickel-chromium-molybdenum steel has been determined by a dilatometric method, using an initial treatment of 10 min. at 850° C. By the same means the S-curve was redetermined after 10 min. treatment at 1200° C. (before treatment at 850° C.). The isothermal characteristics were also found for a prior treatment of only 1 min. at 1200° C.

With treatment at 850° C., transformation was sluggish and incomplete. Prior treatment at 1200° C. accelerated the intermediate transformation and retarded transformation to pearlite. Prior treatment at 1200° C. had no effect on the transformation to martensite by continuous cooling.

Introduction.

When alloy-steel plate is welded a martensitic zone is formed on either side of the weld, and if the plates are restrained the stresses often cause severe cracking in this zone. Previous work [1] has shown that these stresses are very low (about 20 tons per sq. in.), whereas the normal strength of the martensite is more than 100 tons per sq. in. It was considered that weakening of the martensite might be caused by overheating in welding, and that this might admit of dilatometric investigation.

The present work is less specific, its object being to ascertain the effect of overheating on the S-curve of transformation for a nickel-chromium-molybdenum steel which is subject to typical base-plate cracking. It was decided to make the investigation on the steel under the conditions in which it is used in welding. No attempt was made therefore to remove the heavily-banded structure typical of this steel in the rolled condition.

The results of the work, though negative from the point of view of welding, are of some interest on their own account. In the present case, overheating caused an acceleration of the intermediate transformation, whereas other investigators have found a retarding,[2] or no effect.[3, 4]

Dilatometric Apparatus and Method.

It was decided to find the isothermal transformation characteristics for an initial treatment at 850° C. with and without a prior treatment at 1200° C.

The composition of the steel used in this work is as follows :

C. %.	Si. %.	Mn. %.	P. %.	S. %.	Ni. %.	Cr. %.	Mo. %.
0·31	0·20	0·44	0·036	0·033	3·40	1·26	0·47

The steel was softened at 600° C. for machining.

Dilatometer specimens (1·750 in. in length and 0·20 in. in diam.) were flat-ended, and, as a precaution against decarburisation, were nickel-

* Received February 12, 1944.

plated to a thickness of 0·7–1·0 × 10^{-3} in. A fresh specimen was used for each test except in one particular case, mentioned later. The dilatometer was similar to that used by Allen, Pfeil and Griffiths,[(5)] and the dial gauge could be read to 0·001 mm. The metal bath could be stabilised at any given temperature to within 1½° C.

The induction period was assumed to start from the time when the specimen was within 10° C. of the temperature under investigation. This time was found sufficiently accurately from a graph of the time to reach a temperature of $(T + 10)$° C. against T, where T is the bath temperature. The graph was constructed from cooling curves for specimens quenched in the dilatometer to various bath temperatures.

To find the expansion associated with 100% transformation at any given temperature, recourse was made to a device employed by Allen, Pfeil and Griffiths,[(5)] where comparison is made with the transformation data for a number of similar alloy steels containing 0·3–0·4% carbon, which transform completely in the temperature range considered. The beginning line of the S-curves represents 1% of the full transformation at any particular temperature. Transformation stopped in almost all cases well below 100% completion, and the ending line of the S-curves represents 99% completion of the amount of transformation which occurred in practice at any given temperature.

Isothermal Transformation after Initial Heating at 850° *C.*

A 10-min. treatment was given at 850° C., excluding 3 min. heating time.

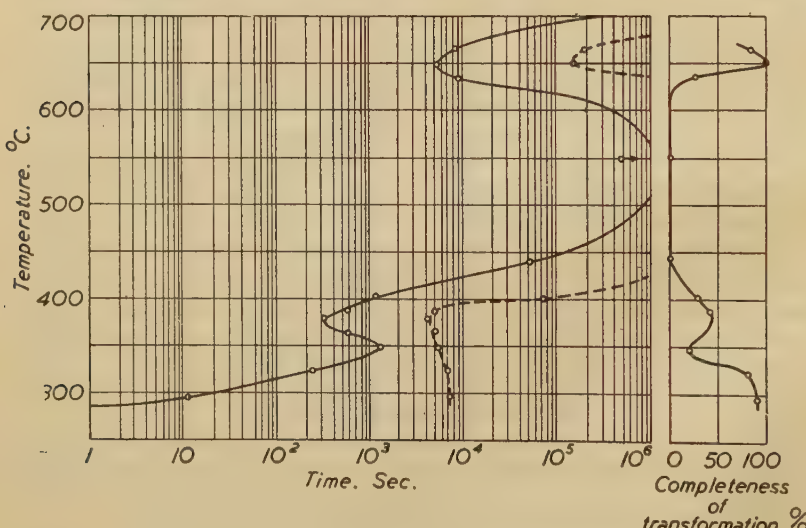

Fig. 1.—S-Curve (treatment at 850° C.).

The S-curve (Fig. 1) shows the usual characteristic shape for this alloy-steel type, though the rather high alloy content causes:

(*a*) Rather sluggish transformation at all temperatures, and
(*b*) low temperatures of transformation in the intermediate range.

Effect of Prior Treatment at 1200° *C. on the Isothermal Transformation.*

To examine the effect of the time-period at 1200° C., two categories of test were used as follows :

(a) *One Minute at* 1200° *C.*—Treatment was given for 1 min. at 1200° C. in nitrogen (3 min. heating time). Without cooling below 850° C. a second treatment was given for 10 min. at 850° C., testing then being as before.

(b) *Ten Minutes at* 1200° *C.* Treatment was given at 1200° C. (9 min. heating time) in a packing of alumina powder. Treatment at 850° C. and testing were then as before.

TABLE I.—*Test Results.*

Test.	Bath Temp. ° C.	Time for 1% of Complete Transformation. Sec.	Time for 99% of Transformation Occurring in Practice. Sec.	Completeness of Transformation. %.
		10 *Min. at* 850° *C.*		
A	664	8,080	199,000	84
B	650	5,340	150,000	100
C	633	9,000	...	26
D	550	>480,000	...	0
E	440	53,200	...	1·5
F	402	1,120	73,900	30
G	385	584	5,030	42
H	377	324	4,180	43
I	361	587	5,010	30
J	347	1,320	5,210	18
K	322	233	6,840	80
L	295	11	7,300	90
		10 *Min. at* 1200° *C.*		
M	659	17,600	461,000	21
N	650	19,800	242,000	38
O	405	121	2,800	58
P	387	159	2,400	70
Q	368	229	2,850	84
R	347	170	4,210	100
S	339	156	4,320	95
T	323	80	5,890	92
U	289	2	7,140	98
		1 *Min. at* 1200° *C.*		
V	654	18,300	302,000	25
W	374	264	3,110	84
X	342	174	4,280	99·5
Y	290	4	7,100	99
		*Repeated Treatment.**		
Z	341	235	4,300	84

* *See text.*

For purposes of reference the complete results are given in Table I., and the S-curve after 10 min. prior treatment at 1200° C. is presented in Fig. 2. Test *S* (Table I.) was treated for 10 min. at 1200° C. in nitrogen as a check, and the results are incorporated in Fig. 2. For comparison, the S-curve after an initial treatment at 850° C. is included in Fig. 2 in thin line.

For the intermediate transformation the beginning and ending curves are moved upwards and to the left, overheating having accelerated the changes.

In the pearlite range of transformation, times for the beginning and

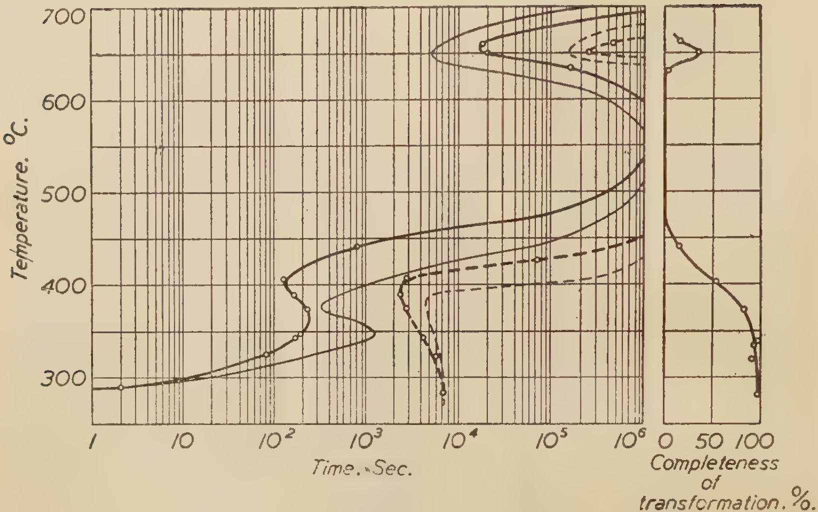

FIG. 2.—S-Curve (treatments at 1200° and 850 °C.).

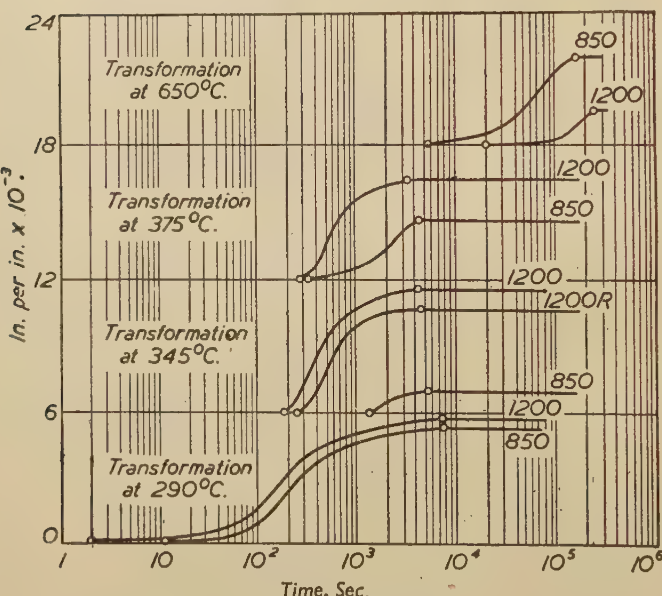

FIG. 3.—Dilatation Curves.

the end of the transformation were considerably retarded and rendered much less complete (38% compared with 100%). The same effects have been found (in the few tests available) after only a 1-min. treatment at 1200° C. as after the 10-min. treatment.

Treatment at 1200° C. for 1 or 10 min. caused an increase in grain

size from 11 to 2 (American Society for Testing Materials (McQuaid-Ehn) Standard).

Dilatation Curves.

These curves are shown in Fig. 3 for tests *B* and *N* (top panel), *H* and *W* (second panel), *J* and *R* (third panel) and *L* and *U* (bottom panel).

Transformation characteristics for test *W* (1 min.) agreed, except for starting time, with those for 10-min. treatment at 1200° C., and differed from those of test *H*. Similarly, test *X* (1 min.) gave a curve almost coincident with that shown for test *R*. The third curve (1200*R*) is for test 2, in which the specimen from test *R* was reheated to 850° C. and tested again. Its transformation characteristics show a partial reversion to those of test *J*. The curve for test *Y* lay close to that given for test *U*, and it is possible that the small differences in this case can be in part ascribed to the slight differences in bath temperature.

Transformation to Martensite on Continuous Cooling.

The S-curve is less accurate at low temperatures and, as a check, the effect has been found of overheating on the transformation of this steel to martensite on continuous cooling. Treatment was given for 10 min. at 850° C. with and without a prior treatment of 1 min. at 1200° C., the resultant transformation curves being the upper and lower curves (full line), respectively, shown in Fig. 4. They are almost identical and are therefore shown separated by an arbitrary amount. Correction has been made to these curves for the normal thermal contraction during the change, and this is shown in broken line for the upper curve only, the other being almost identical. In both cases transformation was about 90% complete at 100° C. and continued increasingly slowly to room temperature. A small part of the transformation (5%) occurred at constant temperature (17° C.); this proceeded increasingly slowly with time and, in fact, was found to be proportional to log.-time, being completed in about 200,000 sec. Transformation, as a whole, was then probably very near completion.

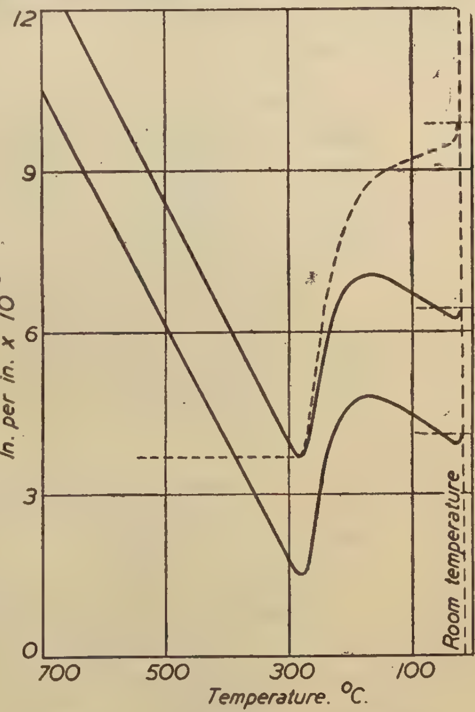

Fig. 4.—Transformation to Martensite by continuous cooling.

Microscopic Examination of Transformed Specimens.

Fig. 6 illustrates the marked banding noticed with this steel. With this specimen (test *J*) the structure is fine and is largely martensitic (18%

transformation to lower-bainite). Fig. 7 shows the transformation to lower-bainite after overheating (test S).

The transformation product at the most rapid rate in the intermediate range (test H) is seen in Fig. 8, the martensitic matrix remaining unetched.

The effect of overheating on the transformation to pearlite is shown in Figs. 9 and 10 from tests C and N, respectively.

Micrographs have not been included from specimens treated for 1 min. at 1200° C., but were similar in grain size and structure to those treated for 10 min.

Discussion.

Acceleration of the intermediate transformation due to overheating was not found by previous investigators,(2, 3, 4) in one case (2) the overheating causing a retarding, and in others (3, 4) having no effect. In these cases steels were used giving much faster transformation, and lower overheating temperatures were employed. This acceleration of the intermediate transformation was partly reversed on treating again at 850° C. (Fig. 3, third panel), and this reversible part of the change may be due to a change in grain size. In the main, however, overheating produced a permanent change in the characteristics of the intermediate transformation.

This permanent change may be due to the redistribution of carbon or alloying elements on overheating. Treatment at 1200° C. was observed

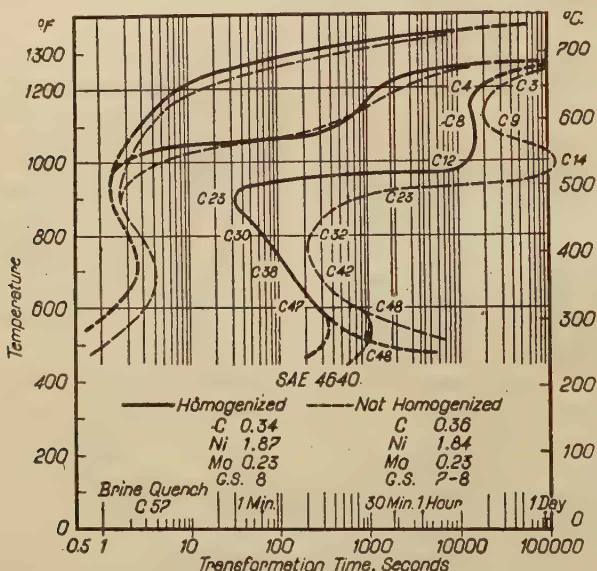

FIG. 5.—Effect of Homogenising on S-Curves (Davenport (6)).

to lessen the banding, without completely removing it. Fig. 5 is a diagram reproduced from a paper by Davenport (6) showing the effect on the S-curve, for an alloy steel, of homogenising for several hours at 1200° C., and it may be seen that permanent changes were produced very similar to those found in the present case.

It should not be assumed that the sole effect of overheating is to increase the grain size. For instance, Steinberg, Skluiev and Nossyreva (2) have

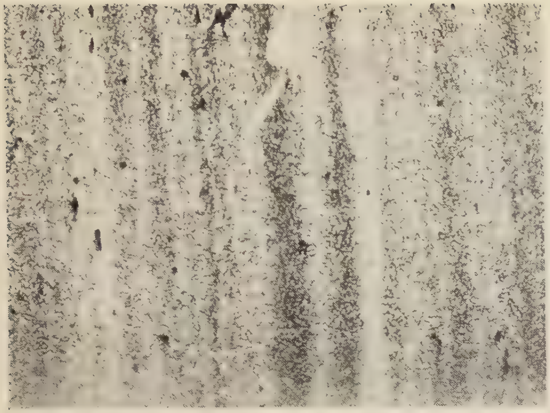

FIG. 6.—Banding in 4S11 Steel. × 150.

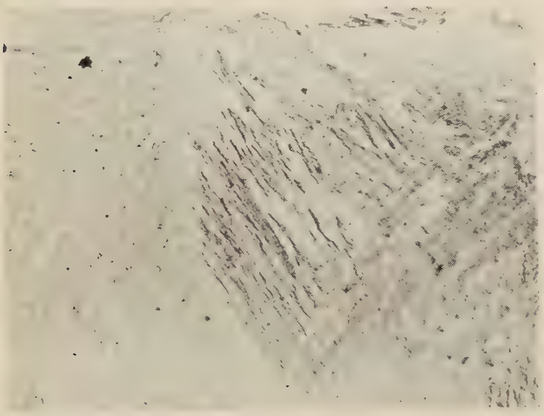

FIG. 7.—Specimen Treated at 1200° C. and Transformed at 339° C. × 600.

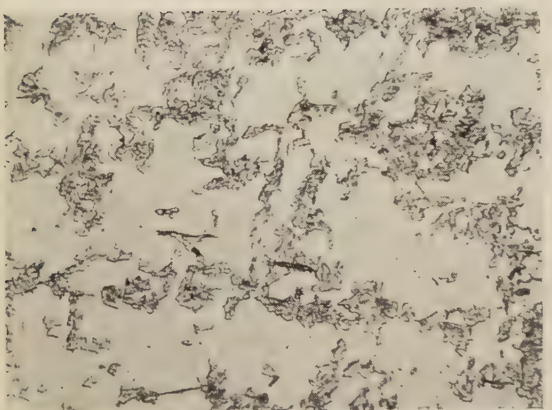

FIG. 8.—Specimen Treated at 850° C. and Transformed at 377° C. × 600.

Specimens etched in 2% nitric acid.
(Micrographs reduced to four-fifths linear in reproduction.)

[*Winterton.*
[*To face p.* 84 P.

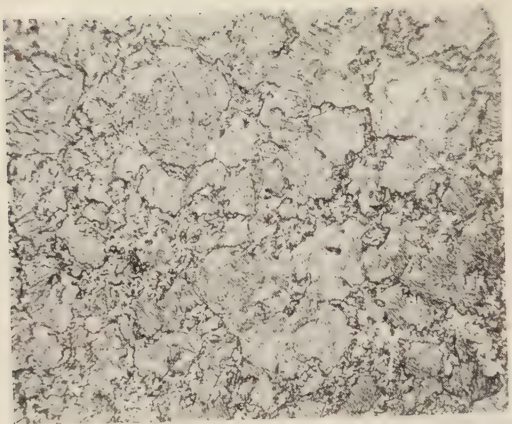

FIG. 9.—Specimen Treated at 850° C. and Transformed at 633° C. × 1200.

FIG. 10.—Specimen Treated at 1200° C. and Transformed at 650° C. × 600.

Specimens etched in 2% nitric acid.
Micrographs reduced to four-fifths linear in reproduction.)

[*Winterton.*
[*To face p.* 85 P.

explained some of their results on the assumption that overheating causes an "increased inherent stability of the austenite, possibly by producing a more orderly lattice." Again, in the present case, overheating has a homogenising influence.

Analysis showed that the general loss of carbon due to overheating did not exceed 0·03%, which is not significant. In addition, the similarity in transformation characteristics after 1 min. and 10 min. treatment at 1200° C. militates against significant decarburisation.

The results found here for the effect of overheating in retarding the transformation to pearlite confirm the work of previous investigators,[2, 3, 4, 7, 8] and the absence of effect of overheating on the transformation to martensite[3] has also been confirmed.

As far as welding is concerned it may be stated that any weakening of the steel due to overheating in welding (comparable with 1 min. at 1200° C.) is not accompanied by appreciable dilatation changes.

Conclusions.

(1) Treatment for 10 min. or 1 min. at 1200° C. (prior to treatment at 850° C.) caused an acceleration of the intermediate transformation and a retarding of the pearlite transformation.

(2) Acceleration of the intermediate transformation was, in the main, a permanent change, and was attributed to a redistribution of carbon or alloying elements. This acceleration may be augmented by a reversible change with grain size.

(3) Prior treatment at 1200° C. did not affect the transformation to martensite on continuous cooling.

(4) Any weakening of the steel due to overheating in welding is not accompanied by dilatation changes.

Acknowledgments.

The author wishes to acknowledge the considerable help and encouragement given by Professor D. Hanson, in whose laboratories this work was carried out. Thanks must be expressed to Mr. Harman for help in the supply of alloy-steel plate.

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A NOTE ON THE PHYSICAL PROPERTIES OF AN AUSTENITIC WELD-METAL AND ITS STRUCTURAL TRANSFORMATION ON STRAINING.*

By K. WINTERTON, B.Sc., Ph.D. (Birmingham University).

(Figs. 4 to 6 = Plate IV.)

Synopsis.

Mechanical tests at elevated temperatures on composite 18/8 weld-plate tensile specimens, showed that the tensile strength, yield strength and hardness declined with increased testing temperatures. The elongation was at a maximum when testing at 250° C.

The effect of prior heat treatment at 850° C. in causing increased tensile strength and decreased yield strength decreased with testing temperature and was not apparent above 150° C. Its effect in reducing elongation persisted for testing temperatures up to 500° C.

Microscopic examination showed a breakdown of dendritic regions to a light-etching α-constituent, and the formation of lines and blocks of a deep-etching α-constituent, probably due to uneven straining. After treatment at 850° C. the deep-etching α-constituent was found at the grain boundaries, particularly after testing in the range of 200°–400° C. (This reduced the elongation values.)

Introduction.

Previous work,[1] of which this is an extension, had shown that a short treatment in the range 700°–975° C. caused a certain embrittlement of austenitic (18/8 type) welds, *i.e.*, an increased tensile strength, slightly decreased yield strength and much reduced ductility. Low ductility is undesirable since it may cause high stresses to be built up across the weld.

An equivalent heat treatment obtains in multi-run welds and it was important to find whether the embrittlement persisted at elevated temperatures, since it is at elevated temperatures that failure of welded joints occurs, and weld properties affect the restraint stresses built up.

Testing Method.

The analyses of the materials used were as follows :

	C. %	Mn. %	Si. %	S. %	P. %	Ni. %	Cr. %	Mo. %
Plate composition . .	0·32	0·57	0·22	0·016	0·016	3·38	0·65	0·26
Electrode composition .	0·06	...	...	0·03	0·03	9·5	20·0	3·0
Weld deposit composition	0·142	0·96	0·35	0·039	0·021	7·84	{14·22, 14·26}	2·51

Small composite tensile specimens (1·5 in. long × 0·15 in. in dia.), were made from single-run 60° V butt-welds on the ½-in. thick alloy steel plate. Specimens were taken at right angles to the weld, to include a small length of weld metal at the centre, and two series were tested :

* Received February 12, 1944.

Series I.—Before cutting specimens, the butt-welded plate was softened at 400° C. This facilitated machining without significantly affecting the properties.

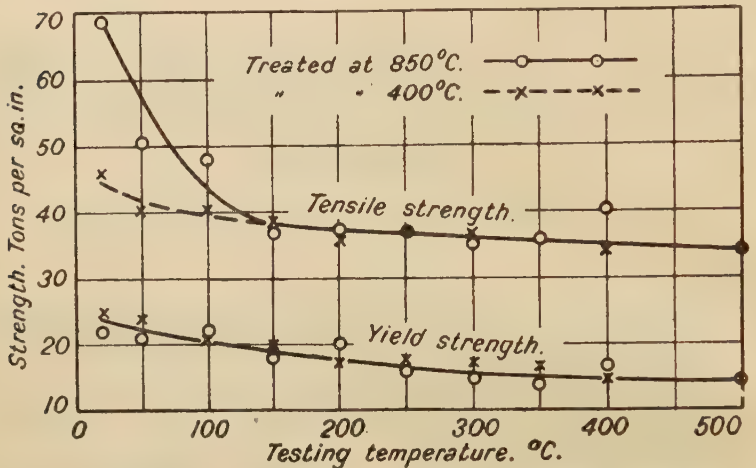

Fig. 1.—The Strength of Weld-Metal Test-Pieces at Elevated Temperatures.

Series II.—The butt-welded plate was softened at 400° C., and the specimens subsequently treated at 850° C. for 10 min. (A central temperature from the brittle range 700–975° C.)

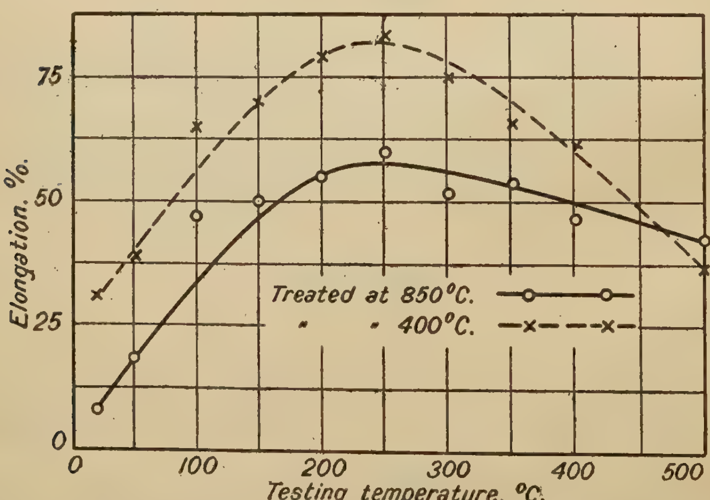

Fig. 2.—The Elongation of Weld-Metal Test-Pieces at Elevated Temperatures.

Testing was carried out in a Hounsfield tensometer, the specimen being held in long steel grips inside a small electric furnace. The length of weld-

metal included in any specimen varied over the section because of the V-shape of the weld. The percentage elongation was calculated using the narrowest part of the weld as gauge length, this being 0·2–0·3 in. The rate of straining was approximately 0·0025 in. per sec.

Results of Physical Testing.

Tensile and yield strength decline with increased testing temperature (Fig. 1), and the elongation shows a maximum at 250° C. (Fig. 2).

The effect of prior treatment at 850° C. in causing increased tensile strength persists for testing temperatures up to 150° C., and the loss of ductility persists throughout the range employed (up to 500° C.).

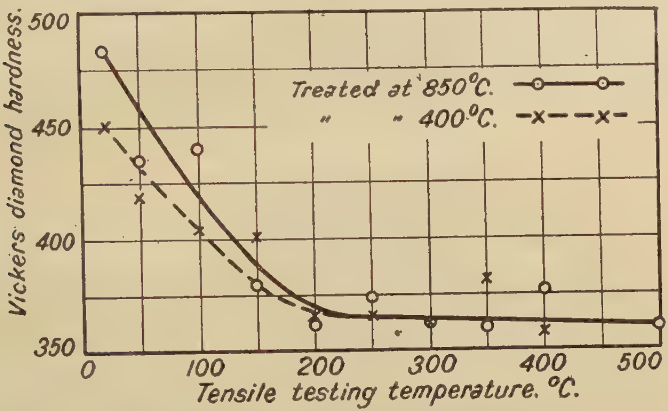

FIG. 3.—The Vickers Diamond Hardness of Weld Metal Measured at Room Temperature after Tensile Testing at Elevated Temperatures.

Average Vickers diamond hardness values (V.D.H.—10 kg.m., ⅔-in. objective) were taken on the weld metal near the fracture, at atmospheric temperature. The high hardness after tensile testing (450–500 V.D.H.), decreased with the testing temperature (Fig. 3) to about 360 V.D.H. at 200° C., thereafter remaining fairly constant.

Nature and Position of the Fracture.

A transcrystalline fracture occurred in the centre of the weld with two exceptions :

(1) The specimen from series II. tested at atmospheric temperature gave an intercrystalline fracture along the weld/plate interface.

(2) The specimen from series II. tested at 50° C. gave a partly intercrystalline fracture which started at the interface and ran through the weld.

Intercrystalline cracking at or near the interface under these conditions has been attributed to the precipitation of fine carbides along the interface.[1]

Microscopic Examination of the Strained Austenite.

Each specimen was ground in longitudinal section, polished and etched in Kalling's reagent (5 g. of cupric chloride, 100 c.c. of ethyl alcohol, 100 c.c. of concentrated hydrochloric acid, 100 c.c. of water).

In all cases, a light-etching α-constituent had formed in the dendritic regions throughout the weld (Fig. 4). This constituent increased in depth of etching, both near the fracture and with increased tensile testing temperature to a maximum value at 400° C.

A different α-constituent (much deeper etching) was also found in blocks (Fig. 5, left), and in a pattern of crossed lines (Fig. 5, top right) with specimens from both series. This was also found on the grain boundaries of specimens from series II., increasing in quantity with the testing temperature to a maximum at 400° C. This transformation at the boundaries can be seen in Fig. 5 (specimen from series II., tested at 200° C.), and at high magnification in Fig. 6. Experiment has shown that heat treatment in the range of 200–500° C. on specimens previously treated at 850° C. has no effect; and it follows that straining in this temperature range is necessary to produce this transformation at the boundaries.

The deep-etching structures may be compared with the very faint dendritic α-constituent within the grain just discernible at the centre of Fig. 5 and shown at higher magnification in Fig. 4.

Discussion of the Results.

The two important factors governing the $\gamma \longrightarrow \alpha$ breakdown in this steel are as follows:

(1) Steels of this composition are usually far from equilibrium, but equilibrium conditions on straining become of importance. For 18/8 steels the α-phase is stable below 400° C. and the γ-phase above 600° C.[2] In the present case specimens revealed a greater tendency to remain austenitic when strained at 500° C.

(2) Straining causes an approach to equilibrium conditions, this tendency appearing to increase with increased straining temperatures.

The occurrence of the deep-etching α-constituent in the form of blocks and lines is probably due to the uneven straining which is characteristic of 18/8 steel. The lines are presumably related to the stressing direction, but in any event are not connected with the crystal orientation, since they do not change their direction across grain boundaries.

Treatment at 850° C. will cause precipitation of carbides at the boundaries, and it is well known that where such precipitation occurs, there is a predisposing tendency for the $\gamma \longrightarrow \alpha$ change, thought to be due to chromium impoverishment in the vicinity. The presence of the α-constituent at the boundaries in specimens from series II. in the form of a rigid network accounts for the lower ductility found in this case. As the fracture is transcrystalline in both series for testing temperatures over 150° C., it is to be expected that the boundary constituent will not cause much difference in tensile strength at the higher temperatures.

The high final hardness (about 500 V.D.H.) attained after testing 18/8 austenitic steel to fracture may be due to the considerable lattice strain imposed by a close intermixing of the α and γ lattices, rather than to the retention of carbon in supersaturated solution, for the following reasons:

(1) The hardness of martensite increases with the percentage of carbon, but Pfeil and Jones[3] found a maximum hardness for 0·1% carbon with strained 18/8 austenitic steel.

(2) The hardness falls off with the straining temperature to about 370 V.D.H. at 150° C. Martensite usually maintains its hardness in this temperature range.

PLATE IV.

FIG. 4.—The α-Constituent formed in the Dendritic Regions. Etched with Kalling's reagent. $\times$ 600.

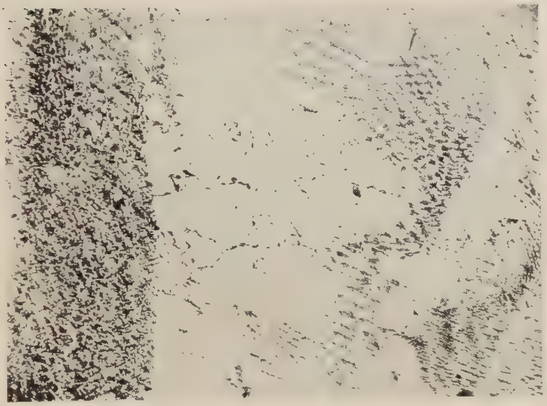

FIG. 5.—Various Forms of Deep-Etching α-Constituent. Etched with Kalling's reagent. $\times$ 60.

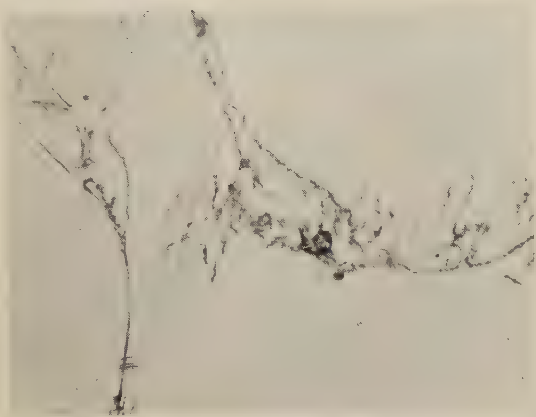

FIG. 6.—The α-Constituent at Grain Boundaries. Etched with Kalling's reagent. $\times$ 2000.

(Micrographs reduced to four-fifths linear in reproduction.)

[*Winterton.*

[*To face p.* 90 P.

Acknowledgments.

The author is indebted to Professor D. Hanson for his considerable help and encouragement, the work being carried out in the Department of Metallurgy, University of Birmingham. Thanks are due to Mr. H. H. Burton for the supply of alloy steel plate.

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TENSILE PROPERTIES OF UNSTABLE AUSTENITE AND ITS LOW-TEMPERATURE DECOMPOSITION PRODUCTS.*

BY A. H. COTTRELL, B.Sc., Ph.D. (UNIVERSITY OF BIRMINGHAM).

(Figs. 12 to 18 = Plates V. and VI.)

SYNOPSIS.

The mechanical properties and transformation behaviour of an air-quenched nickel-chromium-molybdenum steel were studied at various stages before, during and after the austenite-martensite change. Part of the S-curve was determined as a preliminary to the investigation.

Tensile tests on specimens in the unstable austenitic condition show that (a) the metal has a low elastic limit and is ductile, (b) transformation is induced by plastic strain, the extent of the effect depending upon the sluggishness of the $\gamma \longrightarrow \alpha$ change, as indicated by the S-curve, and (c) the products of transformation (bainite and martensite) are more elastic and less ductile than the austenite from which they are derived.

The tensile properties of martensite obtained by air-quenching were examined during cooling and show that the material possesses high strength and appreciable ductility. Minimum ductility in the metal is obtained immediately after the completion of the change to martensite.

These results are discussed briefly in relation to the problem of cracking encountered in this type of steel after welding.

Introduction.

SINCE the work of Davenport and Bain,[1] the decomposition of unstable austenite under isothermal conditions has been much studied. In all investigations specimens have been used in such a way that transformation has occurred in a stress-free material, save only for those internal stresses created by the action of the transformation. Under certain industrial conditions, such as welding for example, transformation frequently occurs in material under considerable externally applied stress, and there are good reasons for expecting the transformation to be accelerated by its presence. That plastic deformation increases the rate of approach to equilibrium is well known and has been demonstrated [2, 3, 4] in those high-alloy steels possessing a metastable austenitic structure at room temperature, cold-working of which causes a partial breakdown to the α structure.

There is a practical need for some knowledge of the mechanical properties of quenched austenite before and during its change to martensite. The fact that the atoms in the austenitic structure are unstably situated and can readily move to new positions suggests that plastic deformation might occur with considerable ease while disintegration of the face-centred cubic lattice is taking place. This has been demonstrated by Wassermann [5] for the $\gamma \longrightarrow \alpha$ change in an iron-nickel alloy at low temperatures.

A knowledge of the properties and behaviour discussed above is of interest in connection with problems such as distortion and cracking associated with the hardening of steel. Of importance in this respect are the properties of the freshly formed martensite while cooling down to room temperature and before tempering effects have been introduced. The

* Received February 12, 1944.

cracks which occur after welding in air-hardening alloy-steels are frequently situated in a martensitic zone produced adjacent to the weld by the heat of welding and the rapid cooling of the welded region, and under the action of shrinkage stresses in the joint. Cracking usually occurs before room temperature is reached,[6] and one is led to suspect some unusual form of weakness in this zone either before, during or after transformation to martensite. In explaining and overcoming this problem it is important to know the mechanical properties of the zone during the later stages of the cooling period, and the effects of varying heat treatment conditions and composition upon them. The work described below is limited to a general survey of mechanical properties and transformation behaviour, using a fixed initial heating temperature, a fixed rate of cooling and a single steel of the following composition :

C. %.	Mn. %.	Si. %.	S. %.	P. %.	Ni. %.	Cr. %.	Mo. %.
0·32	0·57	0·22	0·016	0·016	3·38	0·65	0·26

Apparatus and Technique.

(a) *Mechanical Tests.*

Small specimens, first rendered austenitic by heating, were cooled to predetermined temperatures at a cooling rate fast enough to suppress all changes except the austenite-martensite reaction, and tested in tension at these temperatures.

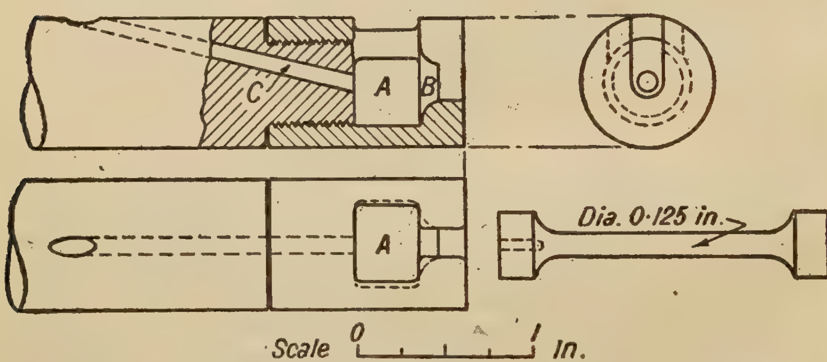

FIG. 1.—Testing Grip and Specimen.

The form of a specimen is shown in Fig. 1, and was designed for use in a tensometer and for quick mounting in the grips. A thermocouple hole was provided in one end for temperature measurement. Each specimen was heated to 850±5° C. in a small electrical-resistance furnace, 15 min. heat treatment being given, 5 min. of which were required to reach the soaking temperature. An atmosphere of dried nitrogen was used successfully for preventing decarburisation and scaling. After this treatment the specimen was withdrawn and cooled in air to the testing temperature.

The very rapid metal-bath quenching so frequently used in S-curve determinations was not adopted, because (*a*) steels of this composition transform rather slowly at temperatures above that at which martensite is formed, and (*b*) air-quenching on small specimens gives a better approximation than metal-bath quenching to the rates of cooling obtained in welds. Fig. 2 gives a comparison of the cooling curve of an air-quenched tensile

specimen A with those obtained on two butt-welds on $\frac{1}{2}$-in. plate—B representing one on unheated plate and C one on plate preheated to 150° C. It was found by dilatometry that air-cooling was sufficient to prevent all change until the temperature of the start of the martensite change was reached. This was about 310–300° C., and could not be lowered appreciably by increasing the rate of cooling.

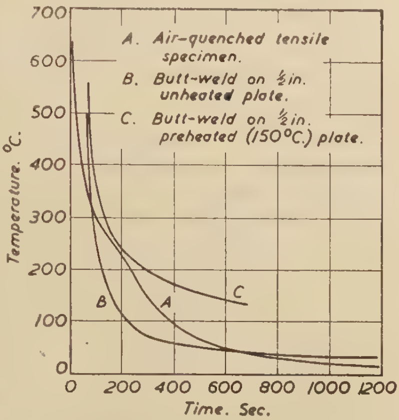

Fig. 2.—Cooling Curves.

When the specimen had cooled to the chosen testing temperature it was mounted in testing grips in a low-temperature furnace maintained at that temperature. Mounting was effected by inserting the ends of the specimen into the recesses A (*see* Fig. 1); the inner surface B of the shoulder of the grips was shaped to fit the shoulder of the test-piece, the load being applied through these two contacting surfaces.

The low-temperature furnace (Fig. 3) was designed to allow change of

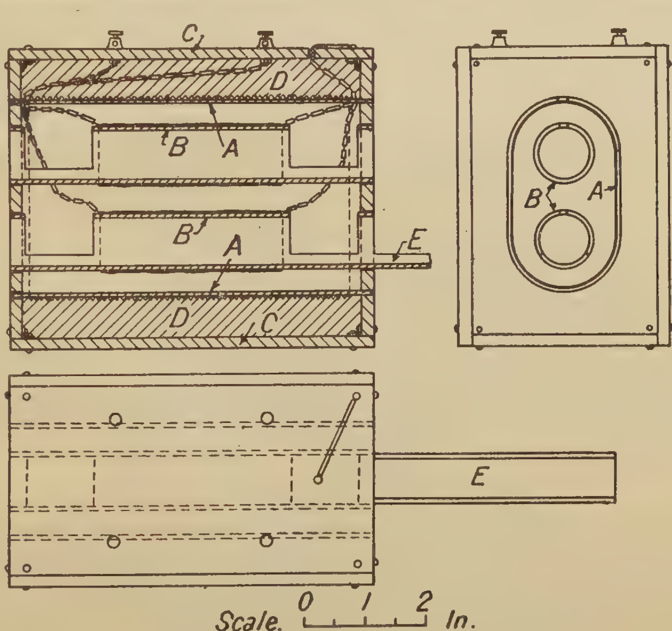

Fig. 3.—Low-Temperature Furnace.

the magnetic permeability of the specimen to be observed, this being used to indicate the progress of the transformation in the specimen. The

transformer principle was used, the specimen being the core and the furnace winding the primary circuit. A secondary winding allowed detection of the magnetic-flux changes consequent upon the transformation in the specimen and was connected to a similar compensating secondary coil in order to balance all induced e.m.f.'s except that due to the magnetic properties of the specimen. The primary and secondary windings were made of Nichrome wire wound on split tubes of non-magnetic nickel-chromium alloy. In Fig. 3, *A* represents the primary (or furnace) tube and *B* the secondary tubes. The furnace was enclosed in asbestos board *C*, and lagged with asbestos *D*. Brackets and a screw adjustment were attached to allow the furnace to be mounted and aligned on the tensometer.

The testing grips were heated and supported in the lower secondary tube, and a couple hole *C* (*see* Fig. 1) allowed temperature measurements to be made. Temperature control varied with the testing temperature, but was within $\pm 2°$ C. at the highest temperature used. An extension-piece *E*, shown in Fig. 3, was fitted on the secondary tube to support the grips while mounting a specimen. An auto-transformer provided the heating current, and the secondary current, after rectification, was measured on a suitable galvanometer.

It was found impossible to record the progress of transformation *during* the tensile test by this measurement, owing to large permeability changes produced by the straining action. The method was therefore confined to the determination of the progress of transformation for the construction of the S-curve.

Each specimen was tested in the tensometer using a constant rate of strain (0·0025 in. per sec.), and a load-elongation curve was determined. The elongation obtained by this method was a combination of the total strain of the specimen and the elastic strain of the grips, and correction for the latter was not applied. Properties measured were tensile strength, elastic limit, percentage reduction of area and total elongation. Where required, the true breaking stress (breaking-load/final-area) was also measured. When an examination of the microstructure was necessary, the specimen was quenched in 5% NaOH solution immediately after testing.

(b) *Isothermal Transformation.*

To determine the S-curve, dilatometer specimens (1·75 in. × 0·2 in. in dia., axial couple hole $\frac{3}{4}$ in. long) were used in a simple silica-tube type of dilatometer, rather similar to that used by Allen, Pfeil and Griffiths.[7] A dial gauge, sensitive to 0·001 mm., measured the dilatation. The specimens were heat-treated as described above, and when air-cooled to the selected temperature the low-temperature furnace was used to obtain an isothermal transformation. Simultaneous readings of the dial gauge and the galvanometer provided dilatometric and magnetometric records of the transformation.

The Isothermal Decomposition Curve.

Records of the transformation were made at various temperatures between 525° and 300° C. and some examples of the results obtained are seen in Fig. 4, in which the progress of the transformation, as indicated by the dial gauge and galvanometer deflections, is recorded. The agreement between the two methods is fairly good. From curves of this type the time was determined for 1%, 50% and 99% transformation to occur, these points being used to construct the S-curves (Fig. 5) representing the start, middle and end of transformation, respectively.

No recordings were made above 525° C., so that the part of the curve giving the austenite-to-pearlite change is missing. Also omitted is a line

extending upwards from the 430° C. point on the start-of-transformation curve, indicating the formation of pro-eutectoid ferrite. At temperatures

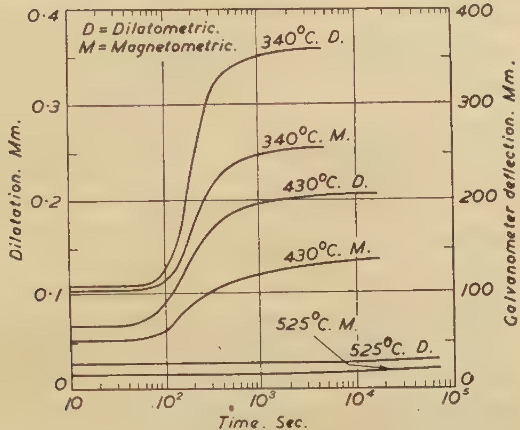

FIG. 4.—Austenite Decomposition Curves at Various Temperatures.

up to 525° C. the amount of this constituent formed was hardly appreciable. Only over a limited range of temperature (425–310° C.) does the end-of-transformation curve indicate that all the austenite has transformed, and at other temperatures examined the transformation appears to cease before reaching completion. These portions of the curve have therefore been shown as full and broken lines, respectively.

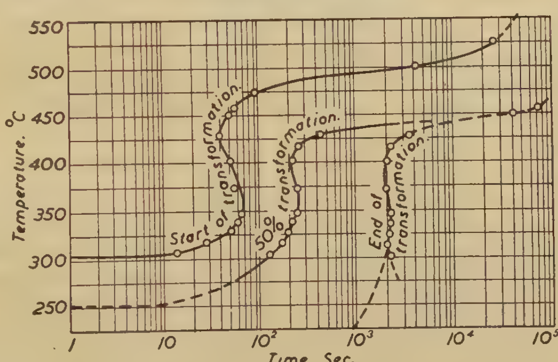

FIG. 5.—Isothermal Decomposition Curves.

The characteristics of the isothermal decomposition of austenite in steels of this type are now amply described in the literature of the subject,(8, 9, 10, 11) and render further discussion of Fig. 5 unnecessary.

Mechanical Properties Before and During the Austenite-Martensite Reaction.

(a) *Effect of Temperature.*

Fig. 6 records an examination of the mechanical properties in the range 525–150° C.; the results in these and following curves are averaged

from repeated tests. The temperature of the commencement of the martensite transformation (310–300° C.) is well shown in the elastic-limit

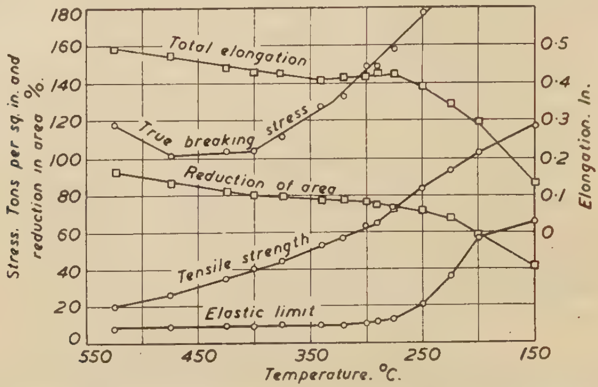

FIG. 6.—Effect of Testing Temperature on Mechanical Properties of Unstable Austenite in the Temperature Range 525–150° C.

curve, which begins to rise as soon as the harder phase appears. There is a corresponding although less definite change in the slope of each of the other curves (except the true breaking stress) at temperatures in this range. As the testing temperature is lowered progressively below 300° C. the elastic limit and tensile strength increase and the elongation is reduced, these changes being consistent with the fact that the martensite-austenite ratio, at the commencement of the test, is progressively increasing. At temperatures above that of the martensite transformation, the tensile strength is lowered and the elongation and reduction of area are increased with increasing elevation of the testing temperature. The low, almost constant value of the elastic limit (10 tons per sq. in.) and the large elongation indicate that the unstable austenite is capable of considerable plastic deformation under low stresses.

To observe the effect of straining on the structure, some specimens were quenched after testing and prepared for micro-examination. Figs. 12 to 14 are taken from a specimen tested at 425° C. and show bainite (dark) and martensite (unetched), the latter representing austenite which remained untransformed at the end of the test. Fig. 12 is from the unstrained head of the specimen and represents the normal transformation (approximately 45%) which occurs during the time (185 sec.) taken by the test. Fig. 13 is from the shank of the specimen, remote from the fracture, and Fig. 14 from the region near the fracture. It is plain that transformation has been promoted by strain, for there is considerably more bainite in the shank than in the head, and near the fracture, where deformation has been extensive, transformation appears to be complete. Fig. 15 (head) and Fig. 16 (shank) show similar effects in a test at 320° C.; the head of the specimen shows a partial breakdown to lower bainite (dark needles), while in the shank the change is almost complete.

Where decomposition of the austenite has been promoted by straining, the resultant structure is very fine. It is very probable that the increased transformation rate during the test is due to the production of strain nuclei within each austenite grain. Since these are defined as points of greatest strain, it is reasonable to suppose that breakdown of the austenite starts at these points and transformation nuclei are formed. The fineness of the

resultant structure may then be accounted for by the large number of nuclei produced.

While straining to fracture in the intermediate range (425–320° C.) results in a complete decomposition of the austenite, it is found that at higher temperatures, where the decomposition rate is very slow, only partial transformation is produced by straining. Fig. 17 shows the structure of the shank of a specimen tested at 525° C., and indicates that a slight breakdown only has been effected, the product consisting essentially of pro-eutectoid ferrite. Examination of the unstrained head of the specimen showed that, in the time interval of the test, no transformation occurred except for an occasional nodule of ferrite. It is clear then that only limited breakdown by straining occurs at those temperatures where the transformation is very sluggish. Even in the fractured region of specimens tested at 525° C., where deformation was very extensive (*e.g.*, 93% reduction of area), less than one-half of the austenite was decomposed. This may be seen in Fig. 18.

If, in this temperature range, austenite has a higher breaking strength than the ferritic product of transformation, then it is possible to account for the form of the curve of true breaking stress in Fig. 6; as the testing temperature is progressively raised above 400° C. more austenite is present in the region of the fracture, and at 525° C. the proportion of this phase becomes sufficient to give an increase in the value of the breaking stress.

A selection of typical stress-strain curves from the series is given in Fig. 7. The one obtained at 340° C. is characteristic of those associated

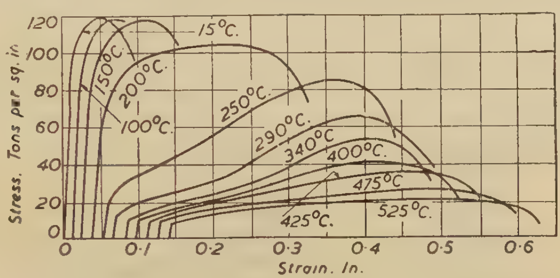

FIG. 7.—Stress-Strain Curves at Various Temperatures.

with a complete austenite decomposition during the test, the most interesting feature being the increase in the rate of strain-hardening which occurs after an initial period of low strain-hardening. This effect is consistent with the fact that a new structure of greater elastic strength and lower ductility is appearing during the test. The stress-strain curve is of this form only over a limited temperature range (250–425° C.); quite obviously, when the temperature is lowered appreciably below 300° C., sufficient of the harder constituent (martensite) is present before testing to nullify the effect. At higher temperatures, on the other hand, the austenite does not transform very extensively during the test, and curves obtained in this range do not show an increase in the strain-hardening rate.

(b) *Relation between Mechanical Properties and Transformation at* 340° *C.*

A series of tests was performed at 340° C., and, in each, a definite known amount of transformation to bainite was allowed to occur before commencing the test. The results are given in Fig. 8, which shows the dependence of the properties upon the percentage of austenite transformed at the start of the test; examples of the stress-strain curves are given in

Fig. 9. Simple relationships exist between the properties and the initial structural composition, the elastic limit, for example, being increased linearly with the amount of transformed phase present. This latter is to be expected if one considers the specimen to be an aggregate of two different materials (austenite and bainite), each exerting an influence proportional to its own elastic limit and to the amount present.

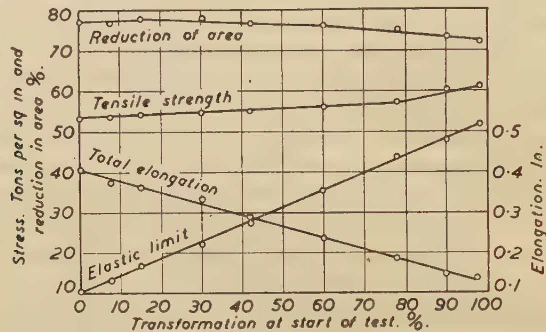

FIG. 8.—Properties during Isothermal Change at 340° C., showing variation with percentage of transformed austenite at commencement of test.

The tensile strength and reduction of area are almost independent of the amount of austenite originally present, and the slight dependence that there is may be accounted for by the fact that the total elongation varies markedly throughout the series. This causes differences in the cross-sectional area at the maximum load in different specimens; if the tensile strength and reduction of area are calculated from the area existing at the

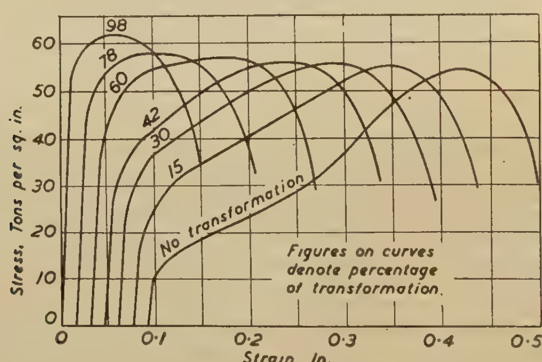

FIG. 9.—Stress-Strain Curves obtained during Isothermal Change at 340° C.

maximum load instead of the original area, then there is no significant variation in these properties through the series. These modified results, with the values of the true breaking stress, are given in Table I., and suggest that by the time that the maximum load is reached the structure is in all cases the same, *i.e.*, that transformation is essentially completed by the time when the period of general uniform elongation is ended.

TABLE I.—*Effect of Percentage Transformation on Mechanical Properties at 340° C.*

Transformation at Start of Test. % :	Nil.	7·5.	15.	30.	42.	60.	78.	90.	98.
Modified tensile strength. Tons per sq. in. . .	71·7	70·8	71·2	70·4	68·7	67·8	67·0	68·8	69·4
Modified reduction of area. %	69·2	69·8	70·8	72·1	69·3	71·6	71·0	70·3	69·4
True breaking stress. Tons per sq. in. . . .	128	133	130	122	127	124	130	123	118

Fig. 8 shows that the total elongation has a linear relation with the amount of austenite present; the elongation of any specimen is composed of (*a*) that which occurs in the untransformed austenite and causes it to decompose to the ferritic structure, and (*b*) that which occurs during the strain-hardening of the ferritic material. Making the reasonable assumption that the second component is constant through the series, it may be deduced that austenite can undergo only a fixed amount of strain during its decomposition under these test conditions. This suggests that transformation during the test is controlled by, and proportional to, the amount of strain applied to the specimen.

To confirm this idea, tensile tests were commenced (at 340° C.) on specimens that were initially wholly austenitic, and in each case a chosen amount of strain was applied, after which the specimen was immediately quenched. The amount of austenite transformed during each test was then determined by micro-examination of the shanks, and the results are given in Fig. 10, which relates the percentage change produced in the test

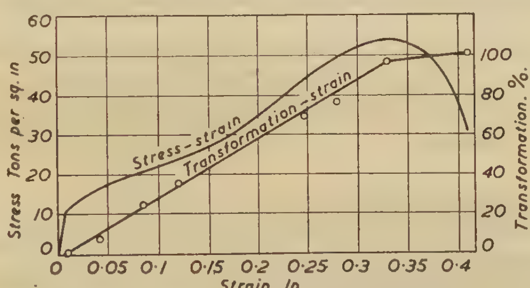

FIG. 10.—Progress of Transformation with Strain.

to the amount of applied strain. During the period in which the specimen elongates uniformly the transformation is directly proportional to the amount of plastic strain undergone. The breakdown appears to commence immediately the elastic limit is exceeded and is almost completed (96%) at the maximum load.

Mechanical Properties After the Austenite-Martensite Reaction.

Fig. 11 gives the results of tests performed at temperatures below 200° C. As the testing temperature is lowered to 135° C. marked property changes occur which are a continuation of those observed in Fig. 6 and which mark the final stages of the austenite-martensite reaction. It is concluded that the martensite reaction is not complete until the temperature of the specimen reaches 135° C. Below this temperature there is little change in

the mechanical properties except for a minimum ductility in the range 135–100° C. Some typical stress-strain curves from this series are given in Fig. 7.

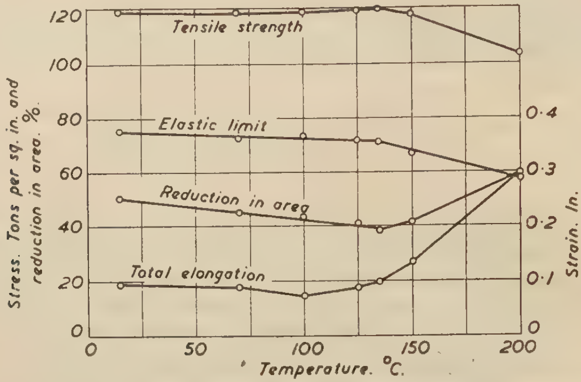

FIG. 11.—Variation of Mechanical Properties with Testing Temperature (below 200° C.).

In order to determine whether the slight ductility increase on cooling from 100° to 15° C. is a temperature effect or a tempering one, some specimens were given various treatments at or below 100° C. after air-cooling from 850° to 100° C. The results are summarised in Table II., an examination of which makes it plain that the ductility increase on cooling to room temperature is mainly a temperature effect.

TABLE II.—*Effects of Tempering and Quenching Treatments in the Range* 100–15° *C.*

Treatment.	Tensile Strength. Tons per sq. in.	Elastic Limit. Tons per sq. in.	Reduction of Area. %.	Total Elongation. In.
Water-quenched from 100° C. to room temperature; tested at room temperature	121·6	77·8	51·3	0·098
Air-cooled from 100° C. to room temperature; tested at room temperature .	119·0	75·1	50·2	0·099
Tested immediately specimen reached 100° C.	118·6	72·9	43·6	0·074
Held for 30 min. at 100° C. and then tested	119·2	69·5	45·4	0·085

Summary and Discussion.

(*a*) Tensile tests on unstable austenite have shown that:

(1) The austenite has a low elastic limit and good ductility; plastic strain causes breakdown to a ferritic structure and the product of transformation has a higher elastic limit and lower ductility.

(2) At temperatures where the austenite transforms readily (as indicated by the S-curve), complete transformation may be induced by straining, but at temperatures where the reaction is sluggish, partial transformation only is obtained.

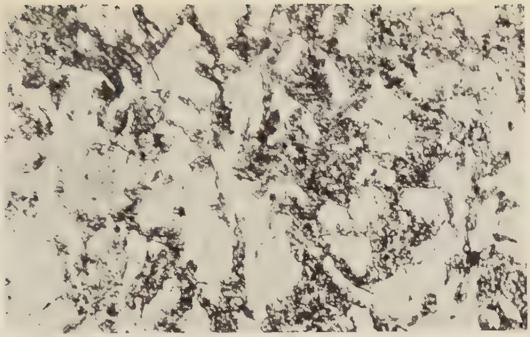

FIG. 12.—Unstrained Head of Specimen. Untransformed austenite is converted to martensite (unetched) by quenching.

FIG. 13.—Structure in Shank. More austenite decomposed during test.

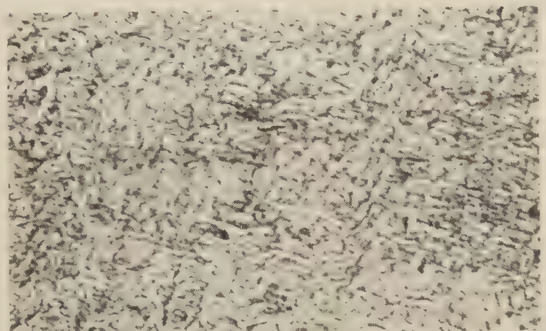

FIG. 14.—Structure near Fracture. Complete transformation during test.

FIGS. 12 to 14.—Structure of Specimen Tested at 425° C., showing extent of transformation to bainite during test. Specimen etched in 2% nitric acid in ethyl alcohol. × 500.

[*Cottrell.*

[*To face p.* 102 P.

PLATE VI.

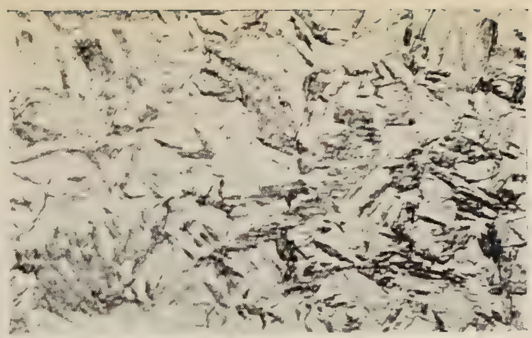

FIG. 15.—Unstrained Head of Specimen. Lightly etched regions represent austenite present at end of test.

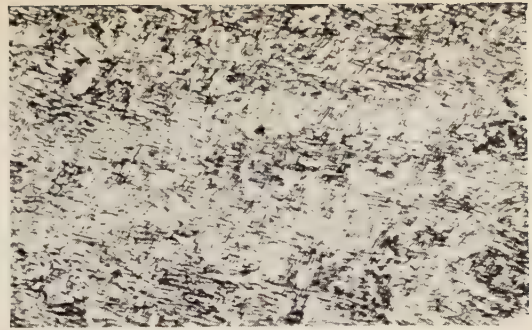

FIG. 16.—Structure in Shank. Decomposition virtually complete during test.

FIGS. 15 and 16.—Specimen Tested at 320° C., showing extent of transformation to lower bainite during test. × 400.

FIG. 17.—Structure in Shank. Slight transformation to ferrite during test. Matrix represents austenite present at end of test.

FIG. 18.—Structure near Fracture. Partial transformation to bainite during test.

FIGS. 17 and 18.—Specimen tested at 525° C. × 500.
Specimens etched in 2% nitric acid in ethyl alcohol.

[*Cottrell*
[*To face p.* 103 P.

(3) At 340° C. transformation by strain commences with the commencement of plastic yielding, the generation of transformed material is linearly proportional to the plastic strain (measured as elongation) and the change is substantially completed before localised elongation commences. Also, simple relations exist between the mechanical properties and the structural composition of the specimen.

Most of the behaviour outlined above is explainable on the basis of the instability of the austenitic structure. The high energy possessed by the atoms in the unstable lattice and the fact that a more stable (body-centred cubic) arrangement is possible by a suitable shearing of the lattice planes render the structure less resistant than is usual to the shearing movement involved in slipping, and a low elastic limit ensues. Also, when plastic strain occurs between planes, the component atoms have the opportunity of coming to rest in alternative positions corresponding either to the face-centred or the body-centred cubic lattice, and, since the latter represents the structure of lower free energy at these temperatures, the atoms will preferentially arrange themselves in that way. Thus, transformation will occur where plastic strain starts and will extend through each crystal as the plastic strain spreads through the crystal. Transformation due to strain should therefore commence at the elastic limit and progress in close relation to the amount of plastic strain, both of which are in accordance with the experimental conclusions.

(*b*) Tensile tests on martensite prepared by air-cooling from 850° C. show that :

(1) The fine-grained martensite obtained by this treatment possesses a high tensile strength and appreciable ductility.

(2) The mechanical properties are hardly affected by the testing temperature, although the ductility is less in the 100° C. range than at room temperature. This ductility change has been shown to be mainly an effect of temperature.

(*c*) In considering a cracking problem such as is encountered in the welding of this type of steel, three factors are important :

(1) The extent of shrinkage strains produced in the hardened zone by the effects of thermal contraction, external restraint and inhomogeneity of the weld region.

(2) The capacity for plastic deformation.

(3) The breaking strength.

Of the properties required to avoid cracking by the action of the effects enumerated in (1), the ability to deform plastically must be considered to be far more important than the breaking strength, owing to the fact that the rate of increase of stress with strain is very much less in the plastic range than in the elastic range, which leads to a correspondingly smaller stress increase when a given shrinkage strain occurs plastically instead of elastically.

It is therefore evident that cracking failure is exceedingly unlikely while the transformed zone is wholly or partially austenitic, since this structure possesses ideal properties for strain accommodation without producing high stresses. The unstable austenite will maintain only a small shrinkage tensional stress, and this will be largely relieved during the transformation to martensite by the volume increase accompanying the change.

Stress rise will be much greater in the martensite range, however, since the material then has a high elastic limit and a much lower ductility. In the present work, the properties of the martensite are rather too good

to account for cracking failure in the hardened zone, and it must be considered that the martensite encountered in such zones is much inferior mechanically to that produced by the treatment described above. The overheating in the region of the weld deposit may very well be the factor causing this deterioration in properties, for it is known [12] that martensite prepared from overheated austenite is weak and brittle.

Acknowledgments.

The author wishes to express his gratitude to Professor D. Hanson for his helpful suggestions and encouragement.

Thanks are also due to Mr. H. H. Burton and Dr. T. Swinden for their help in supplying and rolling the steel plate, and to Dr. N. P. Allen for supplying materials used in the furnace construction.

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A FURTHER NOTE ON THE MICROSTRUCTURE OF HIGH-SILICON ACID-RESISTING IRON.*

BY J. E. HURST, D.MET., AND R. V. RILEY, PH.D. (MESSRS. BRADLEY AND FOSTER, LTD., DARLASTON).

(Figs. 1 to 6 = Plates VII. and VIII.)

SYNOPSIS.

Further observations on the nature of the "barley-shell" structure (*see Journal of The Iron and Steel Institute*, 1944, No. I., pp. 221 P–224 P) are recorded. The well-defined barley-shell structure observed by W. Wrażej (*Journal of The Iron and Steel Institute*, 1944, No. I., pp. 227 P–236 P) on pure (electrolytic) iron and steel, was found to be produced only when the etchants containing hydrofluoric acid were contaminated with silica or silicon-bearing compounds. The dissolved silica in the etchant could be derived from a number of sources, as, for example, the glass vessel in which the reagent had been stored or from a specimen of high-silicon iron alloy previously etched in the same etchant. The barley-shell structure was invariably produced on high-silicon iron alloys even when the etchant was free from silica.

The fact that silicon or its compounds, either in the etching reagent or in the specimen, are essential to the formation of the barley-shell structure will have to be taken into consideration in any explanation of its exact nature.

THE experimental observations recorded by the authors in a previous note † led them to certain conclusions regarding the so-called "barley-shell" and "cracked film" structures obtained under prescribed conditions of etching iron-silicon alloys of high silicon content. Whilst the exact nature of these structures was not determined experimentally, the following were the conclusions drawn :

(*a*) The structures were not true structures of the alloys.
(*b*) The structures were of a film-like character.
(*c*) Their formation under the observed conditions as recorded was a characteristic of high-silicon iron alloys.

Although the barley-shell structure was produced readily on a silicon-iron alloy, many experiments with the etching reagents described failed to produce this structure on a variety of other ferrous materials, including a specimen of pure (electrolytic) iron supplied by Dr. W. Wrażej. In a recent paper ‡ by the latter the appearance of barley-shell markings is recorded on pure (electrolytic) iron, plain carbon steel and cast iron in addition to iron-silicon alloys. In view of this the authors have repeated their previous experiments and made a number of further investigations using specimens of electrolytic and Armco iron. These experiments, as described below, have resulted in the discovery of the conditions under which

* Received September 4, 1944.
† "A Note on the Microstructure of High-Silicon Acid-Resisting Iron," *Journal of The Iron and Steel Institute*, 1944, No. I., pp. 221 P–224 P.
‡ "The Apparent Microstructure Produced by Hydrofluoric Acid Etching Reagents on Pure Iron and Iron-Silicon Alloys," *Journal of The Iron and Steel Institute*, 1944, No. I., pp. 227 P–236 P.

the barley-shell structure occurs on materials other than silicon iron when using these etching reagents.

When using Corson's or the picric acid reagent * under the conditions specified by Wrażej, provided that the reagents were freshly prepared, not a single instance of the barley-shell structure on electrolytic or Armco iron was produced after a 5–20-sec. etch in a glass vessel. Both these reagents freshly prepared and used under the same conditions on specimens of acid-resisting silicon iron containing 14–16% of silicon invariably produced the barley-shell structure.

In the course of the experiments the observation was made that if these reagents were allowed to stand in a glass vessel several hours prior to use, then when used for the etching of electrolytic or Armco iron they produced a barley-shell structure having the same appearance as that found on silicon iron when etched with the freshly prepared solutions. The observation that the etching effect of these reagents when applied to electrolytic or Armco iron was influenced by their contact with glass vessels suggested a number of further experiments, the results of which are recorded in Table I.

TABLE I.—*Results of Etching Experiments on Electrolytic Iron, Armco Iron and* 15% *Silicon Iron, Using Corson's or the Picric Acid Reagent for* 5–20 *sec. at Room Temperature.*

Experiment No.	Material of Vessel.	Etching Solution.	Structure.	
			Armco and Electrolytic Iron.	15% Silicon-Iron. *
1	Glass.	Fresh solution.	Normal.	Barley-shell.
2	Glass.	Used after standing 3 hr. in vessel.	Barley-shell.	Barley-shell.
3	Lead.	Fresh solution.	Normal.	Barley-shell.
4	Lead.	Used after standing 15 hr. in vessel.	Normal.	No etching.
5	Lead.	As in experiment 4 plus 50% fresh solution.	Normal.	Barley-shell.
6	Platinum.	Fresh solution.	Normal.	Barley-shell.
7	Platinum.	Used after standing 15 hr. in vessel.	Normal.	Barley-shell.
8	Platinum.	Used after standing 15 hr. with powdered silica in bottom of vessel.	Barley-shell.	Barley-shell.
9	Platinum.	Used after standing 15 hr. with powdered glass in bottom of vessel.	Barley-shell.	Barley-shell.
10	Platinum.	Silicon iron dissolved in solution, and 50% fresh solution added.	Barley-shell.	...

* A "normal" structure for high-silicon iron is one which shows a silico-ferrite matrix and graphite only, with no trace of the barley-shell structure.

It was found that on using the reagents in dishes of platinum or lead instead of glass the length of time that the reagent stood in contact with the vessel was without influence, no barley-shell structure being produced on samples of electrolytic and Armco iron, whereas this characteristic pseudo-

* The etching solutions were made up as follows, using A.R. chemicals throughout:

Corson's Reagent : 2 parts hydrofluoric acid, 1 part nitric acid and 2 parts glycerol.

Picric Acid Reagent : 1 part hydrofluoric acid and 1 part saturated solution of picric acid in alcohol.

FIG. 1.—Electrolytic Iron, etched 15 sec. in Corson's reagent stored in platinum dish for 4 hr. × 120.

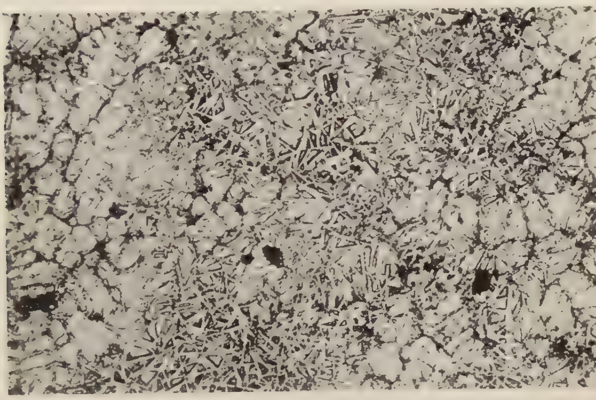


FIG. 2.—× 120.

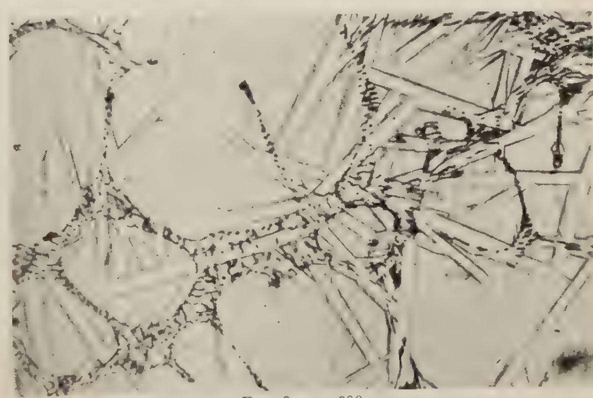


FIG. 3.—× 600.

FIGS. 2 and 3.—Silicon Iron, etched 15 sec. in Corson's reagent stored in platinum dish for 4 hr.

(Micrographs reduced to four-fifths linear in reproduction.)

[*Hurst and Riley.*

[*To face p.* 106 P.

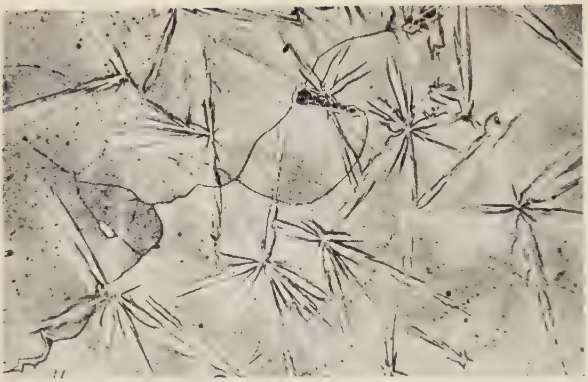


FIG. 4.—Electrolytic Iron, etched 15 sec. in Corson's reagent stored in glass dish for 4 hr. × 120.

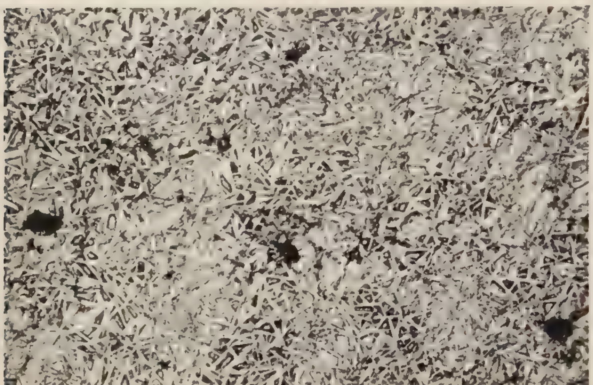


FIG. 5.—× 120.

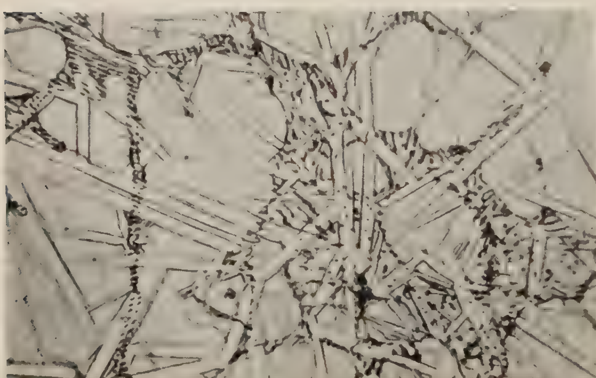


FIG. 6.—× 600.

FIGS. 5 and 6.—Silicon Iron, etched 15 sec. in Corson's reagent stored in glass dish for 4 hr (Micrographs reduced to four-fifths linear in reproduction.)

[*Hurst and Riley.*

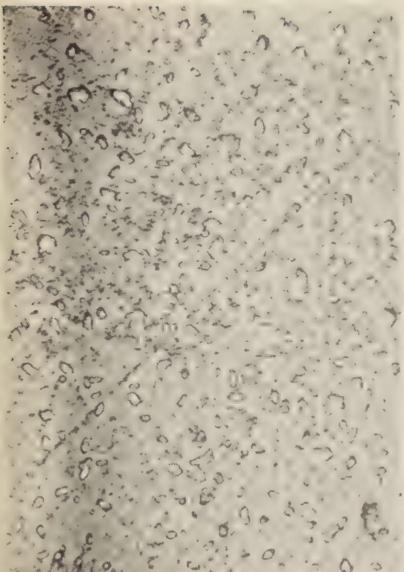


FIG. A.—Electrolytic Iron.

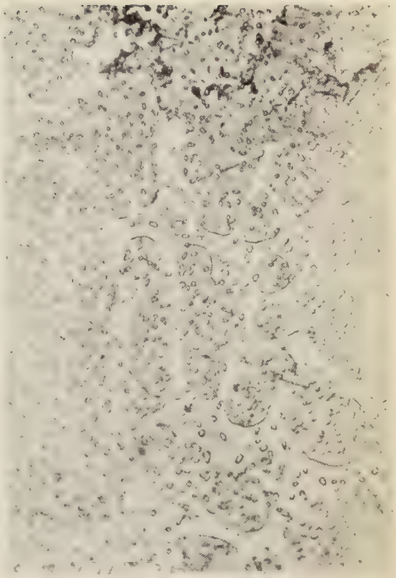


FIG. B.—Iron-Silicon Alloy (silicon 15%).

FIGS. A and B.—Surface of Specimens Etched for 30 sec. with One Drop of Hydrofluoric Acid and Treated with Alcohol. × 400.

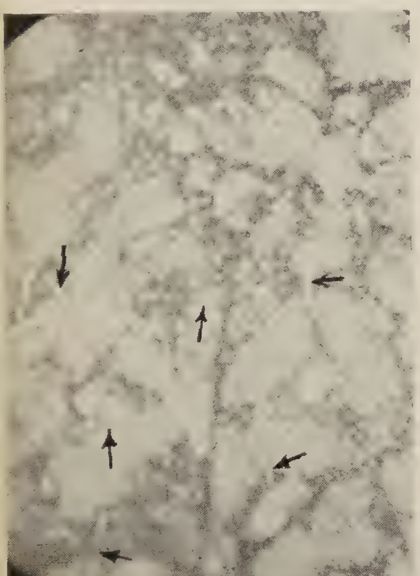


FIG. C.—Iron-Silicon Alloy Etched with HF + Alcohol (1 : 20). Single crystals of iron fluoride shown by arrows. × 300.

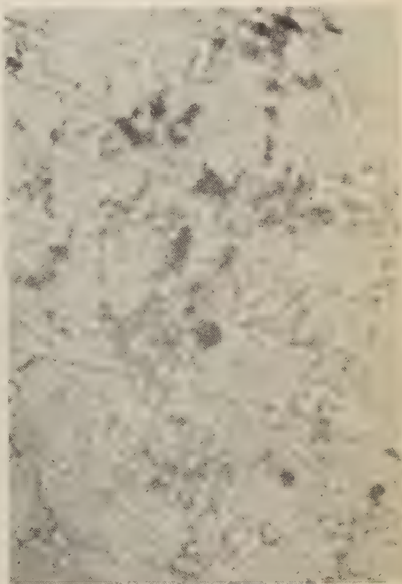


FIG. D.—Same Specimen as in Fig. C after rinsing in water. Etch marks show sites of crystals. × 400.

(*See* Wrażej's contribution.)

structure was formed invariably when etching silicon irons (experiments 2 to 9 of Table I.).

The experiments recorded clearly demonstrate that in order to produce the barley-shell structure with these reagents on electrolytic or Armco iron the reagent requires to be in contact with glass (*vide* experiments 2 and 9) or powdered fused silica (*vide* experiment 8). Experiment 10 shows also that the reagent after having been in contact with 15% silicon iron behaves exactly as though it had been in contact with glass or silica. Whether or not the reagents have been in contact with glass or silica, they invariably produce the barley-shell structure when used on high-silicon acid-resisting iron.

Figs. 1 to 6 are photomicrographs illustrating the main points brought out in Table I. When Corson's reagent was stored for 4 hr. in a platinum dish a normal etch, as shown in Fig. 1, was recorded with electrolytic iron, but the barley-shell structure resulted with silicon iron (Figs. 2 and 3). When, however, Corson's reagent was kept for 4 hr. in a glass vessel a well-defined barley-shell structure was obtained on etching both electrolytic iron and silicon iron, as shown in Figs. 4, 5 and 6.

It is therefore concluded that with etching reagents of the type described the formation of the barley-shell structure is a characteristic of the high-silicon iron alloys, the formation of this structure on pure iron (electrolytic) or Armco iron being due to the contamination of the etching reagent with silica or silicon-containing materials.

It is therefore quite clear that the presence of silicon or its compounds either in the etching reagent or in the specimen is essential for the formation of the barley-shell structure.

CORRESPONDENCE.

(Figs. A to D = Plate VIIIA.)

Dr. Ing. W. J. Wrażej (Imperial College of Science and Technology, Royal School of Mines, London) wrote : I have read with much interest the authors' statement that the hydrofluoric acid etching reagent must be contaminated with silica- or silicon-containing materials in order to produce the (characteristic) pseudo-structure ("barley-shell" markings) on a sample of electrolytic iron. I would congratulate the authors on their excellent micrograph of the barley-shell markings on electrolytic iron (Fig. 4), as, although they are shown at a magnification of only 120, they resemble Corson's micrographs more closely than the markings on the iron-silicon alloy (Figs. 3 and 6) at a magnification of 600.

In my own investigation,* the etching reagent was kept in a paraffin-wax container or in a plastic container coated with paraffin wax, so that it was not in contact with glass. I cannot now say, because I did not record it, which of the samples was etched in the freshly prepared reagent.

My investigation indicated that during attack with an etching reagent containing hydrofluoric acid and alcohol or glycerol, a deposit of iron fluoride was produced, and that after rinsing the sample in water the pseudo-structure (markings) remained on the etched surface. The presence of iron fluoride after the sample had been rinsed in alcohol was verified by X-ray and chemical analysis.

The authors' contention that only contaminated reagent causes marks on a sample of etched pure iron appeared to me to contradict the statement given in many text-books of chemistry (*e.g.*, "Gmelins Handbuch") that iron fluoride crystallises in the presence of alcohol.

* *Journal of The Iron and Steel Institute*, 1944, No. I., pp. 227P–236P.

In my recent paper (*loc. cit.*, Table I.) it was shown from the spacings obtained from X-ray films that the silicon present in a sample of iron-silicon alloy does not alter the crystalline structure of the salt. The specific gravity of the salt obtained from electrolytic iron, however, differed somewhat from that of the salt prepared from the iron-silicon alloy. The values were 1·96 and 1·97, respectively. The salt crystals from both samples were similar in shape and size and in the form of needles.

In the simple experiment described below I attempted to avoid any contamination of the hydrofluoric acid with silica in order to find out whether, under such conditions, the iron fluoride salt could be formed in the presence of alcohol.

Single drops of hydrofluoric acid were allowed to fall from a wax bottle on to the surfaces of samples of electrolytic iron and iron-silicon alloy (silicon 15%); after 30 sec. a few single drops of alcohol were carefully placed on these drops of hydrofluoric acid. Figs. A and B show that crystals of iron fluoride were formed by the action of alcohol on both samples, *i.e.*, hydrofluoric acid had dissolved iron (*see* the revealed structure) and the iron fluoride was crystallized by the addition of alcohol. When the drops of hydrofluoric acid were allowed to remain for 60 sec., the crystals on the iron-silicon alloy were either denser or longer, but on the electrolytic iron they were more numerous.

When a drop of an etching reagent made up of one drop of hydrofluoric acid and twenty drops of alcohol was placed on the surface of an iron-silicon alloy and allowed to remain for 30 sec. before rinsing off with alcohol, single needles of iron fluoride were observed, most of them being oblique to the surface. They were transparent and, owing to their position, difficult to focus. Some of them are shown in Fig. C, marked by arrows. After rinsing the sample in water, the scars (etch markings) show the original sites of the crystals, Fig. D. In this case an etching reagent consisting of pure hydrofluoric acid and alcohol behaved exactly like other reagents, *i.e.*, it produced the apparent microstructure of barley-shell etching scars.

The results both of the authors and of my own investigation lead to the conclusion that the use of reagents containing hydrofluoric acid in alcohol or glycerol solution should be avoided when the true structure of the specimen is to be revealed.

AUTHORS' REPLY.

The Authors wrote in reply: We wish to make the comment that Dr. Wrażej was aware of the fact that we were unable to duplicate his etching results on his own sample of electrolytic iron which he sent to us for test in our laboratory. For this reason, on the appearance of his paper, we repeated our previous experiments and made the further investigations reported by us. As a result, we have discovered that when using Corson's or the picric acid reagent silicon or its compounds must be present either in the etching reagent or in the specimen if a barley-shell structure is to be produced. In the case of the Wrażej sample of electrolytic iron the barley-shell structure could only have been produced by the contamination of the Corson's or picric acid reagents used with silicon or its compounds.

What the simple experiment described by him has to do with the barley-shell structure we fail to see. The somewhat indistinct markings shown out of focus in Figs. C and D appear to bear little resemblance to the pockmarked micrographs Figs. A and B, and still less do they resemble the genuine barley-shell structure shown in Figs. 2 to 6 in our paper. Of the many simple experiments that we could recommend perhaps the simplest would be to repeat the experiments described in our paper.

EMULSIONS OF OIL IN WATER AS CORROSION INHIBITORS.*

By P. HAMER, L. POWELL and E. W. COLBECK (Imperial Chemical Industries, Ltd., Research Department, Alkali Division, Northwich, Cheshire).

(Figs. 8 to 27 = Plates IX. to XII.)

Synopsis.

This paper describes an investigation which was undertaken with the object of preventing corrosion in recirculating cooling-water systems, and which was begun as a result of plant failures. The origin and object of the work are discussed, together with the factors influencing the choice of experimental method. The oil emulsions referred to are those produced by adding "soluble oils" to water. These resemble the cutting oils employed on machine tools.

Attention has been chiefly directed to the prevention of attack on mild steel. Three types of water have been used under both static and flow conditions at room temperature, and at 60° and 90° C.

The investigation was limited to an attempt to answer certain questions which appeared to be of practical value. These questions were:

(a) What is the lowest concentration of emulsified oil required to prevent the corrosion of mild steel?

(b) Does pre-treatment with oil emulsion give protection in the absence of oil, and/or can a heavy initial dose be followed by a light dose?

(c) Can oil emulsion stop pitting or general corrosion which has already commenced?

(d) Do weak oil emulsions cause pitting or intensified local attack?

(e) Do oil emulsions prevent corrosion of non-ferrous metals alone or in contact with mild steel?

The work described in the paper has provided the following very general answers to these questions:

(a) Oil emulsions containing 0·5% or more of emulsifying oil reduce to negligible proportions the corrosion of mild steel under all conditions tested.

(b) The protection afforded by oil emulsion persists for some time after the specimen has been transferred to untreated water.

(c) There is evidence that oil emulsions soften preformed rust and tend to reduce the rate of attack.

(d) Localised attack can occur at parts of the metal which are inaccessible to the emulsion.

(e) Oil emulsions are effective inhibitors of the corrosion of the non-ferrous metals studied, but when in contact with mild steel some attack occurred at elevated temperatures in certain cases.

A suggested mechanism of the protective effects of oil emulsions is that they form a thin adsorbed film of oil over the whole metal surface, which may break down at specially susceptible spots. The film is then thickened due to coagulation of the emulsion by the ferrous salts formed at these anodic points, thus stifling the corrosion.

The effect of the oil film in reducing heat transfer through clean metal surfaces was found to be negligible, but heat transfer was distinctly impaired when the metal surface was already covered with corrosion product.

In hard waters the oil emulsions tend to produce scums, which may be due to the formation of insoluble soaps or to inversion of the emulsion to water-in-oil. The importance of these scums in practice cannot yet be assessed, but it may be necessary to adopt special measures to prevent the fouling of heating surfaces when the water is recirculated.

* Received April 14, 1944.

Introduction.

As a result of the rapid failure of high-pressure mild-steel coolers in recirculating cooling-water systems, the work here described was undertaken in a search for inhibitors which might be suitable for large systems of this kind. The failed tubes had a wall thickness of 0·1 in. and a bore of 0·3 in.; the cooling water passed over the outside of the tube, and failure occurred in a few months as a result of severe pitting of the type shown in Fig. 9.

At the suggestion of Dr. U. R. Evans, who kindly consented to act as consultant for this work, the authors undertook a study of the behaviour and efficiency of oil emulsions made by the addition of "soluble oils" to water. The fact that these emulsions can act as corrosion inhibitors, particularly with respect to corrosion fatigue, has been recorded by a number of workers,[(1)] but the authors were unable to find in the literature that any study of their behaviour as general inhibitors had been made.

The corrosion problem in a recirculating cooling system, and consequently the choice of a suitable inhibitor, is complicated by changes which occur in the water as a result of the operating conditions. In order to give some idea of the conditions under which a corrosion inhibitor will be expected to be effective, a list is given below of the most important factors which must be taken into account:

(1) The volumes of water in circulation may be as high as 2000 cu. m. per hr. with a make-up of about 10% of this volume. The cost of the inhibitor is therefore important.

(2) Concentration of salts in the circulating water occurs as a result of loss by evaporation. There are also losses of water by spray, which tend to limit the concentration of salts, but which will remove inhibitor at the same time. Concentration is also sometimes limited by a small purge.

(3) Temperatures may be as high as 70° or 80° C.

(4) As a result of the concentration of salts and the rise in temperature, sludges and scales of calcium carbonate may be found on the metal surface and in other parts of the plant. These are derived from the temporary hardness of the water, even when this hardness is low in the make-up water, as is the case with most moorland surface supplies.

(5) The temperature and other conditions favour the growth of bacterial and vegetable slimes on the metal, and these, in conjunction with the other deposits, prevent free access of inhibitor to the metal surface, and may thus encourage pitting in place of general attack or complete inhibition.

(6) The pH value of the circulating water usually rises, owing to the loss of CO_2, but in certain circumstances it may fall, owing to absorption of acid gases from the atmosphere.

(7) Non-ferrous metals are usually present in some part of the plant, *e.g.*, circulating pumps, or may have to be used because of the nature of the substances to be cooled.

The problem can be simplified to some extent by treating the make-up water to remove the temporary hardness and by intermittent chlorination to prevent slime growth. In the authors' view, such steps are eminently justifiable on the grounds of the economies to be secured through improved heat transfer and reduction in cleaning costs; also because they simplify the task of the corrosion inhibitor. The best time at which to take these measures is obviously when the plant is first put into commission.

With these considerations in mind the authors tried to devise their experiments to cover the conditions of temperature, water flow and water composition in recirculating cooling-water systems. It speedily became

evident, however, that an enormous number of experiments would be required to make a complete study of the variables involved, and, in practice, the authors have fallen far short of their programme. The results and conclusions given in this paper must therefore be regarded as a general survey of the field. The conclusions are based on some six hundred experiments, of which only a few typical cases can be recorded here. Nevertheless, it is hoped that the evidence presented will prove sufficiently convincing to attract the attention which, in the authors' opinion, this type of inhibitor deserves.

Experimental Method.

The main idea throughout the investigation has been to find the amounts of inhibitor which will stop corrosion. No attempt was made, therefore, at any detailed study of comparative rates of corrosion in the presence of different amounts of inhibitor.

It was considered that a state of complete inhibition existed when the specimens, after treatment, showed no general attack or pitting judged by inspection and/or negligible weight loss.

The difficulty of reproducing corrosion effects at will (*e.g.*, pitting), even under the most carefully controlled experimental conditions, is a matter of common experience,[2] and this consideration must be taken into account when devising methods of investigation. A preferred method is to carry out simple types of experiment, in triplicate or quadruplicate, under a variety of conditions, so that the general trend of events can be determined with reasonable confidence.

This method was followed as far as possible, taking into account the large number of experiments envisaged, and all important observations were made in duplicate, *i.e.*, two separate experiments, sometimes with two or three specimens in each experiment. Where two or three similar specimens were submitted to the same environment at the same time, the total loss of weight of all specimens weighed together was divided by the total area. The usual time of exposure was one week, but longer periods were allowed in some cases, as will be seen from the Tables.

In the flow experiments compressed air was used to circulate the water (*see* Figs. 3, 4 and 5). It was found that a specimen suspended in the aerated column suffered greater loss in weight than the specimen in the liquid column. These results are not included in the figures given because such conditions may be regarded as abnormal; nevertheless, the point is of interest.

Variables Studied.

The following variables were taken into consideration :

(1) *Condition of Liquid Medium.*—Static and flow.

(2) *Temperature.*—Room temperature, 60° and 90° C., alternate heating to 60° C. and cooling to room temperature.

(3) *Condition of Metal Surface.*—No. 1 emery cloth and/or smooth filed finish; some experiments with mill-scaled and pre-rusted specimens.

(4) *Metals Employed.*—Mild steel, aluminium, copper, brass (60/40 and 70/30), cast iron. Bi-metal specimens: Aluminium/steel, copper/steel, brass/steel, aluminium-bronze/steel, manganese-bronze/steel, cupro-nickel/steel and copper/cast-iron.

(5) *Character of Waters.*—Three types of water were used in the first experiments, but when it had been established that the behaviour of the inhibitor in all three was much the same (*see* Figs. 6 and 7) only the most accessible source of supply was used. For brevity, the results obtained with modified river water only are tabulated.

The waters were typical samples of their class, and were modified by adding suitable salts to give the kind of composition which might be

expected from the concentration of waters of these types in a recirculating cooling system. Only the modified waters were used.

Typical analyses of the original and modified waters, the latter indicated by C.H., C.F., C.C., are given in Table I.

TABLE I.—*Analyses of Waters Used.*

	Expressed as—	Parts per 100,000.					
		Town Water.	C.H.	Well Water.	C.F.	River Water.	C.C.
Total alkalinity to methyl orange	$CaCO_3$	1·3	23·5	34·75	59·5	13·00	26·25
Alkalinity due to $NaHCO_3$	$CaCO_3$	Nil.	Nil.	16·10	39·0	Nil.	Nil.
Free CO_2 *	$CaCO_3$	Nil.	6·5	1·25	Nil.	Nil.	1·00
Sulphuric anhydride	SO_3	2·85	13·40	1·30	1·85	6·6	20·65
Combined chlorine	Cl_2	1·05	25·40	10·30	16·2	4·3	11·70
Silica	SiO_2	0·60	Nil.	1·65	Nil.	...	Nil.
Temporary hardness:							
Due to lime	$CaCO_3$	1·3	23·5	11·05	12·50	13·00	26·25
Due to magnesia	$CaCO_3$	Nil.	Nil.	7·60	8·0	Nil.	Nil.
Permanent hardness:							
Due to lime	$CaCO_3$	1·4	19·0	Nil.	Nil.	2·70	5·75
Due to magnesia	$CaCO_3$	1·35	5·0	Nil.	Nil.	7·50	22·50
Total hardness		4·05	47·5	18·65	20·5	23·2	54·5

* The free CO_2 figure represents the freshly made-up water, and, of course, decreases on storage. No attempt was made to maintain it.

Apparatus.

The various types of apparatus employed are shown in Figs. 1 to 5. The designs are self-explanatory and simple. Apparatus of such types can easily be multiplied.

Fig. 1 illustrates the apparatus used for partial- and total-immersion tests at room temperature.

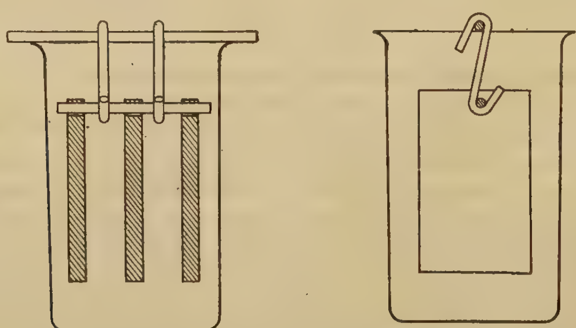


FIG. 1.—Types of Apparatus used for Immersion Tests.

Fig. 2 shows the general arrangement of apparatus used for heating and cooling tests. Instead of using a Bunsen burner, eight or nine of these pieces can be grouped around a gas ring.

Figs. 3 and 4 illustrate two types of cold-circulation apparatus, that shown in Fig. 3 being a copy of the apparatus used by Bannister.[3] The

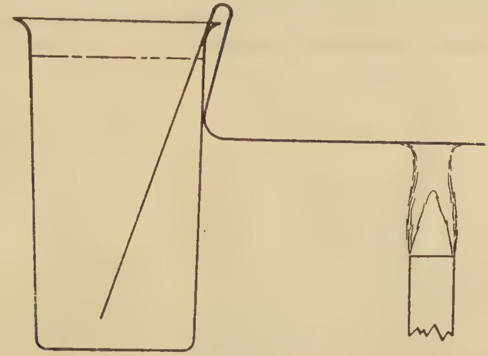


FIG. 2.—Apparatus used for Heating and Cooling Tests.

specimens were either hung in the limbs, as shown, or a specimen steel tube replaced part of the glass limb. The apparatus shown in Fig. 4, a modification of that illustrated in Fig. 3, is easier to make and clean.

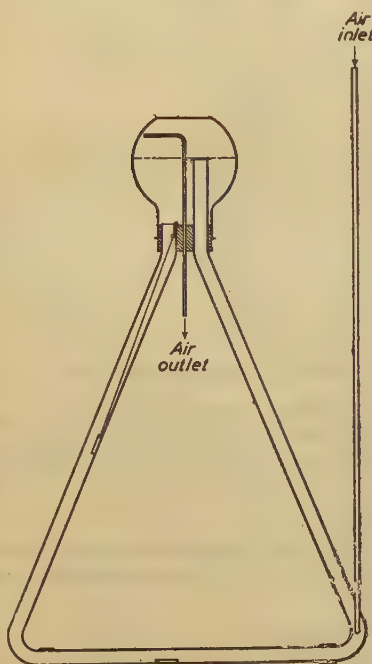


FIG. 3.—Cold-Circulation Apparatus as used by Bannister.

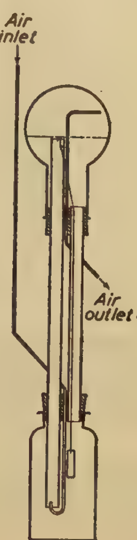


FIG. 4.—A Modified Type of Cold-Circulation Apparatus.

Fig. 5 shows three further modifications of the same basic apparatus. That shown in Fig. 5(*A*) incorporates an immersion heater, and was employed for the tests at 60° C. The units shown in Figs. 5(*B*) and 5(*C*) were both arranged so that they could be immersed in a thermostat and the circulating liquor was kept out of contact with rubber bungs. In Fig. 5(*C*) the specimen itself, a steel tube, is used as the air-lift tube.

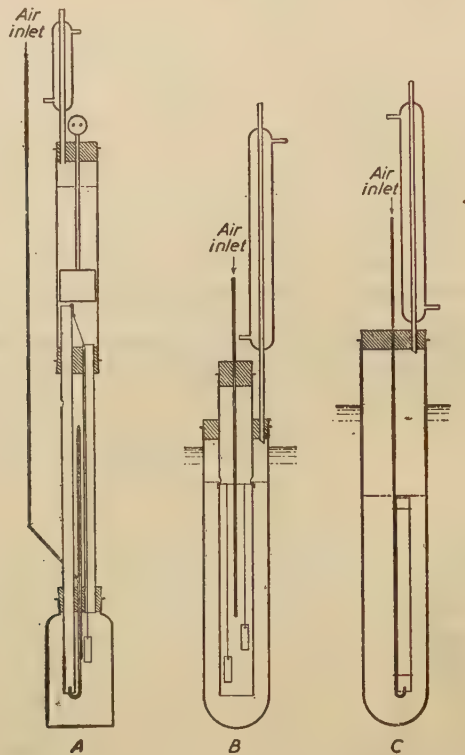


FIG. 5.—Types of Circulation Apparatus used at Elevated Temperatures.

In the apparatus illustrated in Figs. 3 to 5 the circulation was produced by an air lift; the water in contact with the specimens was, therefore, always be fully aerated, *i.e.*, saturated with oxygen at the temperature of the experiment.

Specimens.

(*a*) All static experiments were carried out with flat mild-steel specimens $1 \cdot 1 \times 2 \cdot 0 \times \frac{1}{8}$ in. in size, cut from the same piece of plate, the composition of which was:

Carbon	0·16%	Phosphorus	0·033%
Silicon	0·10%	Copper	0·075%
Manganese	0·56%	Nickel	0·06%
Sulphur	0·034%	Chromium	0·07%
Nitrogen	0·0055%		

The steel was of fully-killed acid open-hearth quality in the hot-rolled condition with normal microstructure.

The specimens were smooth-filed and finished with No. 1 emery cloth. They were stored in a desiccator and de-greased by several washes in alcohol and ether and weighed immediately before immersion. They were inspected immediately after withdrawal from the medium and then washed in warm water, clinging corrosion product being removed with a soft-bristle brush; they were examined again before drying with a cloth or filter paper and weighing.

(*b*) For experiments in flow 1-in. lengths of $\frac{3}{8}$-in. dia. mild-steel rod were used, the composition of which was :

Carbon	0·23%	Phosphorus	0·054%
Silicon	0·06%	Copper	0·07%
Manganese	0·60%	Nickel	0·07%
Sulphur	0·056%	Chromium	0·13%
Nitrogen	0·008%		

This steel was a fully killed basic open-hearth steel in the hot-rolled condition, and the microstructure was normal. The specimens were smooth-filed and received no emery finish.

(*c*) In experiments at 90° C. lengths of mild-steel tubing were used, the outside being finished bright with a smooth file. These were not weighed, results being judged by inspection.

(*d*) For heating and cooling experiments strips of 26 standard wire gauge drum sheet, 8 × 0·7 in., were used, bent as shown in Fig. 2, and finished bright with No. 1 emery cloth. The composition of this material was :

Carbon	0·11%	Phosphorus	0·056%
Silicon	Trace	Copper	0·18%
Manganese	0·34%	Nickel	0·09%
Sulphur	0·055%	Chromium	0·07%
Nitrogen	0·005%		

This was a basic open-hearth steel of rimming quality in the cold-rolled and close-annealed condition. The microstructure was normal.

The object of these experiments was to observe the effect of heat transfer along the length of the metal to the medium under quite severe conditions such as might occur in practice (*see* Fig. 2). The temperature of the medium was raised to about 60° C., and then the heat was cut off and the system allowed to cool. This cycle was repeated 26 times in each case, and during the night or week-end the system was left to stand. Losses by evaporation were made up with emulsion in the earlier experiments and with distilled water later. No difference was observed in duplicate experiments with both methods of make-up. The specimens could not be weighed and the results given are by inspection only.

(*e*) For experiments in flow with bi-metal specimens, these latter were made up as shown in Fig. 8, the two halves being screwed together. The suspension hole is in the steel. For static experiments with the single metals, plates were used of the same size as the mild-steel plates (as in (*a*)). All specimens were finished bright with No. 1 emery cloth. The compositions of the metals tested were as follows :

Copper.—High-conductivity 99·95% copper in the form of $\frac{1}{2}$-in. dia. rolled bar.

Brass.—60% copper, 40% zinc, nominal, in the form of 1-in. dia. extruded bar.

Aluminium.—0·14% silicon, 0·02% iron, other impurities low. In the form of $\frac{1}{2}$-in. dia. hot-worked and annealed bar.

Manganese bronze.—Cu, 56·87%; Mn, 0·83%; Pb, 1·50%; Fe, 1·04%; Sn, 0·75%; Zn (by difference), 39·11%. 1-in. dia. rolled bar.

Cupro-nickel.—Cu, 79·71%; Mn, 0·11%; Fe, trace; Si, 0·03%; Ni (by difference), 20·15%. 1-in. dia. rolled bar.

Aluminium bronze.—Cu, 91·10%; Al, 7·3%; Fe, 0·05%; P, 0·018%; Mn, 0·44%; Zn, 1·02%.

Cast iron.—Carbon, 3·34%; Si, 1·43%; Mn, 0·48%; S, 0·09%; P, 1·42%.

Specimens were machined from the flange of a pipe.

In each of the Tables of results reference will be found to the type of specimen used and this will be indicated by the same letters as in this section, *e.g.*, (*a*), (*b*), (*c*), &c.

TABLE II.—*Effect of Oil Emulsion Strength. Static Tests at Room Temperature.*

Experiment. No.	Time. Days.	Oil-Emulsion Strength. %.	Loss in Weight of Area Immersed. Mg. per sq. cm.	Remarks.*
		C.C. Natural. Partial Immersion. Specimen (a).		
55	7	Nil	0·47 } Average, 0·41	Distinct water-line attack and general corrosion.
68	7	Nil	0·38	
68	7	Nil	0·38	
59	7	0·1	0·29	Slight water-line attack and general corrosion.
61	7	0·25	0·13	No water-line attack. Some general corrosion.
63	7	0·5	0·10	} No water-line attack. Specimens bright.
65	7	0·75	Slight gain.	
109	4	0·75	0·07	
111	4	1·0	0·05	
		C.C. Alkaline.† Partial Immersion. Specimen (a).		
58	7	Nil	0·59	Water-line attack and general corrosion.
60	7	0·1	0·22	Slight water-line attack and general corrosion.
62	7	0·25	0·20	No water-line attack. Some general corrosion.
64	7	0·5	0·03	} No water-line attack. Specimens bright.
190	7	0·5	0·01	
66	7	0·75	0·09	
110	4	0·75	Slight gain.	
191	7	0·75	Nil	
192	7	0·75	0·02	
112	4	1·0	Very slight gain.	
193	7	1·0	0·013	
		C.C. Natural. Total Immersion. Specimen (a).		
46	5	Nil	0·21	} Water-line and general attack.
219	7	Nil	0·39	
220	7	0·1	0·25	Light attack on small isolated areas. Remainder of specimen bright. A little $CaCO_3$ scale formation.
221	7	0·25	Nil	Slight dulling at edges, remainder bright.
222	7	0·5	Nil	Specimen slightly tarnished.
223	7	0·75	Nil	Specimens bright.
		C.C. Alkaline.† Total Immersion. Specimen (a).		
224	7	Nil	0·51	Light general attack over whole of specimen.
225	7	0·1	0·23	A few areas of slight localised attack. Slight $CaCO_3$ scale formation.
226	7	0·25	0·17	Some areas of slight local attack.
227	7	0·5	0·05	One or two tarnished areas, remainder bright.
228	7	0·75	Nil	Specimens completely bright.

* *See* also Figs. 10 to 14.

† Alkalinity of medium was equivalent to 0·5 ml. of normal solution per litre. The solution was pink to phenolphthalein throughout.

Oil Emulsions.

The oil used was a proprietary product sold as an anti-corrosion emulsifying or soluble oil.* It is clear reddish-brown in colour and consists essenti-

* For the supply of this oil the authors are indebted to Messrs. Shell Mex B.P.

ally of a mixture of mineral oil, the sodium salt of a sulphonated mineral oil and a small amount of water. It forms a stable milky emulsion quite readily when added to water with stirring, particularly when the water is warm.

Discussion of Results.

Determination of the Minimum Concentration of Emulsified Oil Required to Prevent Corrosion of Mild Steel.

Experiments were carried out with the three types of water under flow and static conditions, hot and cold. Typical results are shown in

TABLE III.—*Effect of Oil Emulsion Strength. Flow Experiments at Different Temperatures in C.C. Natural Water.*

Experiment No.	Time. Days.	Oil-Emulsion Strength. %.	Loss in Weight. Mg. per sq. cm.	Remarks.
*Specimen (b) at Room Temperature.**				
31	10	0·005	2·73	Some local attack and general staining. Emulsion broke down.
32	10	0·1	1·37	
20	7	0·25	0·15	Specimens quite bright.
25	7	0·5	0·12	Specimen mainly bright but some slight staining.
39	7	0·5	Nil	Specimen bright.
43	7	1·0	Slight gain.	Very slight staining.
Specimen (b) at 60° C.				
77	6	Nil	55·56	Heavy general attack and some pitting.
35	2	0·005	3·66	Emulsion broke down; specimen pitted and generally rusted.
37	5	0·01	21·6	As for 77, above.
33	7	0·25	0·46	Emulsion thinned; specimens bright except for slight rust at edges.
34	7	0·5	0·03	Emulsion thinned, but no rust at edges.
117	3	0·5	Nil	One small patch of rust, remainder of specimen bright.
44	7	1·0	Nil	Specimen bright but some creaming of emulsion.
Tubular Specimens at 90° C.				
138	7	Nil	Visual examination only.	Specimen severely rusted; fairly extensive shallow pitting; perforation at water-line (*see* Fig. 25).
65	5	Nil		Numerous shallow pits.
155	2	0·05		Emulsion renewed twice. General attack and some pitting.
152	2	0·1		Emulsion broke down; tube fairly heavily attacked, particularly at points of contact with glass and string.
154	4	0·1		
153	4	0·25		Emulsion broke down. Some general attack on bottom half of tube.

* *See also* Fig. 15.

Tables II. and III., and Figs. 6 and 7. The following general conclusions are drawn :

(1) Under all conditions the oil emulsion reduces the rate of attack in the three types of water used and, with one or two exceptions, to be anticipated on statistical grounds, the attack becomes negligible at emulsion strengths of 0·5% or above (*see* Figs. 10 to 14 and Fig. 15).

(2) Water-line attack in partially submerged specimens decreases as the concentration of inhibitor is raised, and disappears at or before 0·5%. The water-line attack may even be absent when local attack is seen on the immersed parts of the metal surface.* Corrosion above the water line is also reduced, which may be due to the upward creep of an oil film.

* Dr. R. S. Thornhill suggests that the elimination of water-line attack may be due to slight creaming of the emulsion, leading to the formation of a very high oil concentration in the surface layer. (Private communication.)

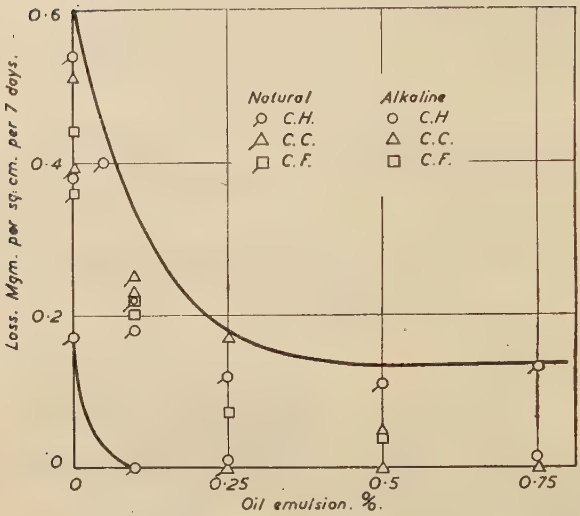


FIG. 6.—Total-Immersion Tests under Static Conditions at Room Temperature.

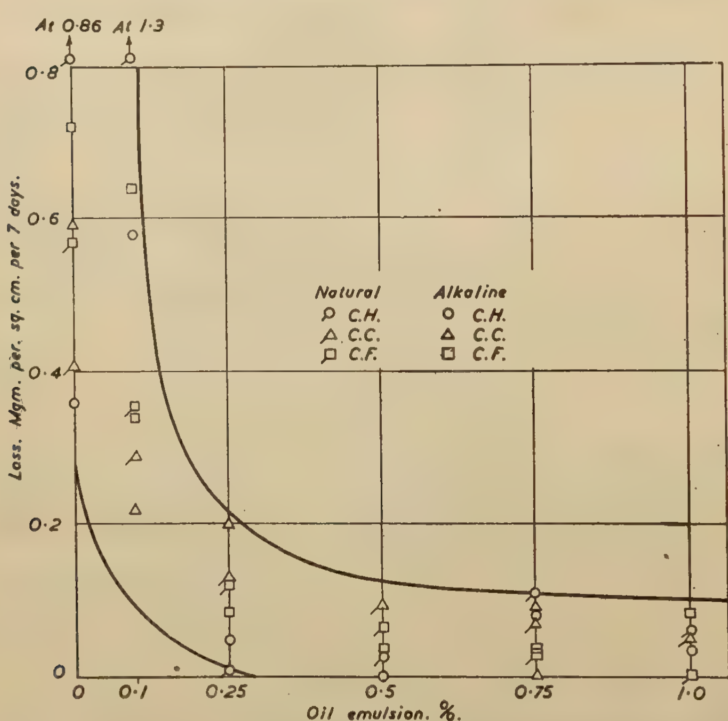


FIG. 7.—Partial-Immersion Tests under Static Conditions at Room Temperature.

(3) In emulsions weaker than 0·5%, attack is sometimes localised, and, in a few cases, definite pits were observed (*see* Figs. 11 and 17). This type of attack is, however, associated with emulsion thinning or breakdown, and this point is discussed more fully later.

TABLE IV.—*Effect of Replacing Oil Emulsion by Water or More Dilute Emulsion. Flow Tests at Different Temperatures with C.C. Natural Water.*

Experiment No.	Time. Days.	Oil-Emulsion Strength. %.	Loss in Weight. Mg. per sq. cm.	Remarks.
		Specimen (b) Totally Immersed at Room Temperature.		
49	4	0·5	0·01	Slight brown stain on specimen.
	4	Nil		
50	4	0·75	Nil	
	4	Nil		
53	4	1·0	Nil	Specimen completely bright.
	4	Nil		
110	8	0·5	Nil	Specimen bright until emulsion strength of 0·06% was reached, when an area of local attack appeared on each specimen.
	3	0·25		
	2	0·12		
	2	0·06		
		Specimen (b) Totally Immersed at 60° C.		
57	4	1·0	1·55	Some general rusting.
	4	Nil		
117	3	0·5	0·08	Small rust spot on specimen but no sign of attack underneath.
	5	Nil		
		Tubular Specimens Partially Immersed at 90° C.		
123	3	0·75	Visual examination only.	Heavy water-line attack. Immersed portion in good condition (*see* Figs. 21 and 22).
	7	Nil		
128	3	0·75		
	7	Nil		
74	3	0·5		
	4	Nil		
65	3	0·5		Emulsion broke down on dilution but specimen remained in good condition.
	3	0·25		
120	4	0·75		Emulsion broke down at 0·25% and heavy water-line attack developed. Some severe pitting at points where rust has streamed down from waterline.
	2	0·5		
	2	0·25		
	3	0·13		
140	2	0·75		Emulsion broke down at 0·25%. Heavy water-line attack. Immersed portion stained but not pitted. *See* Fig. 24.
	2	0·5		
	4	0·25		
	8	Nil		
141	2	0·75		As for No. 140 above. *See* Fig. 23.
	2	0·5		
	8	0·25		
142	2	0·75		Specimen in good condition; no rusting at water-line but some slight general staining.
	2	0·25		
	2	0·25		
	2	0·25		
	2	0·25		
	4	0·25		
		Tubular Specimens Totally Immersed at 90° C.		
156, 157	4	0·5	Visual examination only.	Specimen in good condition except for well marked groove under the string and small amount of pitting at top where specimen became exposed owing to evaporation (*see* Figs. 19 and 20).
	6	0·05		
	14	0·03		
158	4	0·5		Specimen in good condition apart from groove under supporting string and light staining at line of contact with glass rod.
	16	0·1		
159	4	0·5		
	8	0·1		
	14	0·05		
170	28	Nil		Water thick with rust. Heavy deposit of magnetic oxide inside tube. Tube surface showed patchy general corrosion, no acute pits. Figs. 26 and 27, two views of same specimen.

Experiments on Pretreatment with Oil Emulsion Followed by Natural Water and Heavy Initial Doses of Oil Followed by a Lighter Dose.

These were carried out with the three types of water under flow and static conditions, hot and cold. Typical results are shown in Table IV. Conclusions are as follows:

(1) In the cold, under static or flow conditions, a 0·5% oil emulsion has a protective effect which persists for at least four days when raw water is substituted for the oil emulsion. Twelve days in raw water is too long, as will be seen from Fig. 18. No definite time limits can be given for preliminary emulsion treatment required to produce this effect, but a period of not less than two days is suggested.

(2) The experiments to determine the protection afforded by an oil emulsion at 60° C. under flow conditions were not very conclusive, but at 90° C: the protection persisted for periods of up to seven days.

(3) The evidence of the dilution experiments is that good protection can be secured at temperatures up to 90° C. by a preliminary dose of 0·5% emulsion (or stronger) for a period of two days or more, followed by a continuous dose of a dilute emulsion (0·1–0·03%), *see* Figs. 19 and 20. Diluting the original emulsion with water is unsatisfactory because it leads to emulsion breakdown.

The implication of these experiments is that, in the cold, intermittent treatment with 0·5% emulsion appears practicable. The oil emulsion would not, of course, disappear as soon as the addition of fresh oil ceases, but would be diluted by any fresh water entering the system. The experiments suggest that the dilution should not be allowed to fall below 0·05% before fresh oil is added. The intervals at which fresh oil is required will naturally depend on the rate of dilution, but the absence of oil emulsion does not appear to be fatal, provided that this condition does not persist for more than a day or two.

At elevated temperatures, however, oil emulsion should always be present as, in its absence severe local attack is likely to occur. Continuous dosing is therefore recommended, but the 0·5% emulsion can be followed by a more dilute emulsion.

Experiments to Determine whether Oil Emulsions can stop Pitting or General Corrosion which has Once Started.

The object of these experiments was to find out what would be the effects of using oil emulsions in a cooling system in which corrosion has started and the metal surfaces are pitted or coated with rust and possibly scale.

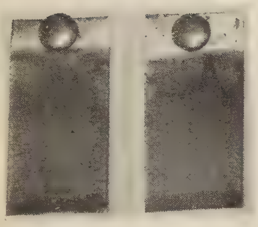
The study of this point in a quantitative manner is hampered by the difficulty of producing standard pre-rusted specimens and also by the variations in the texture of the rust. As a result of a number of experiments at room temperature the following qualitative observations may be made:

(1) Using specimens freshly rusted in water and specimens rusted for some time in air, it was found that 0·5% oil emulsions tended to remove the rust, and that in a period of 7 days the attack on the underlying metal, if any, was small compared with that of water alone.

(2) Using specimens covered with more or less unbroken mill scale at room temperature, 0·5% oil emulsion produced no apparent change in 7 days, but in weaker emulsions, or water alone, some attack was observed, presumably at cracks in the scale. At temperatures of about 60° C., in an experimental recirculating cooling plant, tubes covered with mill scale were badly pitted, but it is thought that this attack occurred when the oil emulsion strength fell below 0·5%.

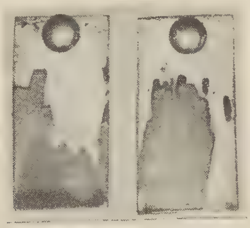
FIG. 8.—Bi-Metal Specimens for use in Flow Experiments.

FIG. 9.—Tube (0·3-in. bore) from Recirculation Cooling-Water System. Failure due to severe pitting.



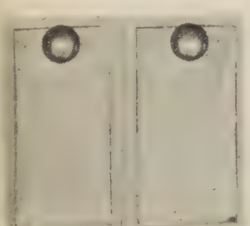
Natural. Alkaline.

FIG. 10.—Raw Water.



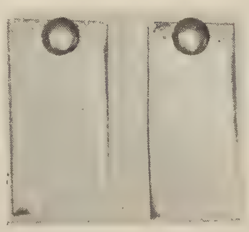
Natural. Alkaline.

FIG. 11.—0·1% Emulsion.



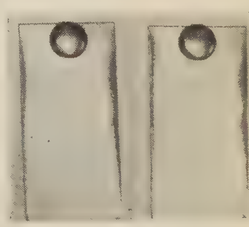
Natural. Alkaline.

FIG. 12.—0·25% Emulsion.



Natural. Alkaline.

FIG. 13.—0·5% Emulsion.



Natural. Alkaline.

FIG. 14.—0·75% Emulsion.

FIGS. 10 to 14.—Results of Partial-Immersion Tests.

Hamer, Powell & Colbeck.

[*To face p.* 120 P.

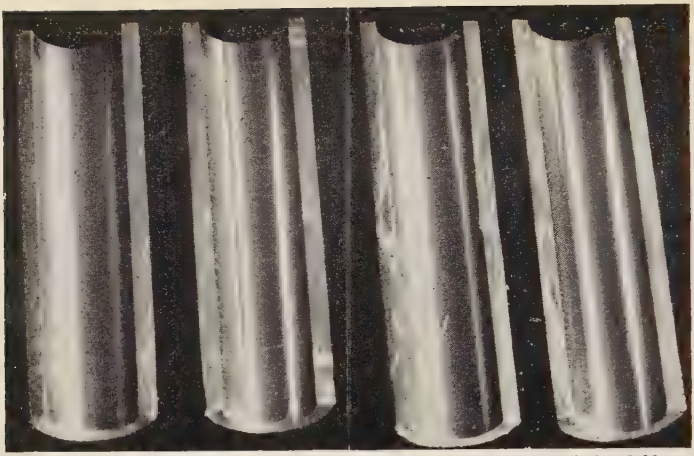


FIG. 15.—Mild-Steel Tubes after 160 Days with 0·75% Emulsion Circulating Cold.

Natural. Alkaline.
FIG. 16.—Raw Water.

0·1% Emulsion. 0·25% Emulsion. 0·5% Emulsion. 0·75% Emulsion.
FIG. 17.—Natural Water.

FIGS. 16 and 17.—Results of Heating and Cooling Tests.

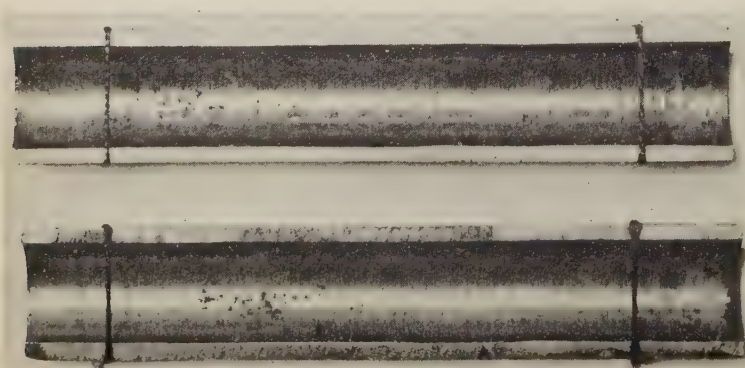


FIG. 18.—Mild-Steel Tubes; 26 days in 0·5% emulsion, then 12 days in untreated water at room temperature.

[*Hamer, Powell & Colbeck.*

DILUTION EXPERIMENTS.

FIG. 19.—Expt. 156, Table IV.

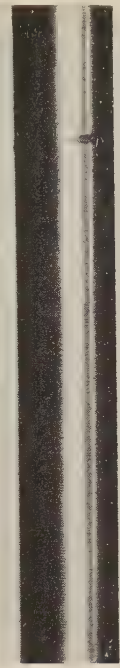


FIG. 20.—Expt. 157, Table IV.

FIGS. 19 and 20.—Corrosion under String at 90° C.; 0·5% followed by 0·05% and later 0·03% emulsion.

FIG. 21.

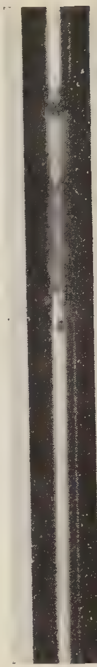


FIG. 22.

FIGS. 21 and 22.—Expts. 123 and 128, 0·75% emulsion followed by raw water.

FIG. 23.—Expt. 141, Table IV.

FIG. 24.—Expt. 140, Table IV.

FIGS. 23 and 24.—0·75% emulsion followed by weaker emulsions, which finally broke down, leaving raw water.

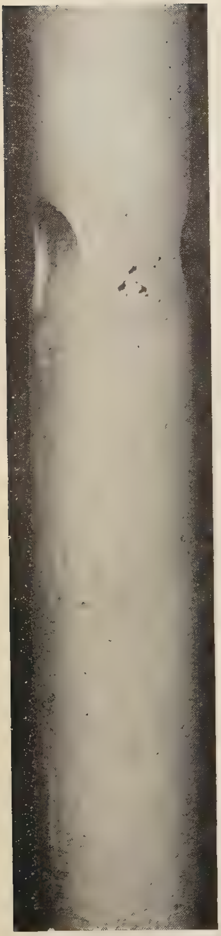


FIG. 25.—Expt. 138, Table III. Seven days. Note the three perforations.

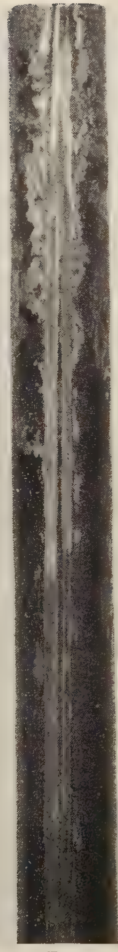


FIG. 26.

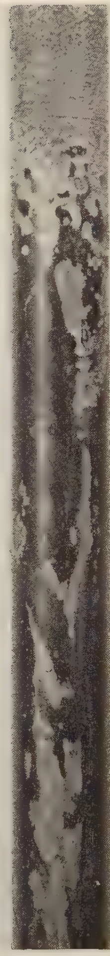


FIG. 27.

FIGS. 26 and 27.—Expt. 170, Table IV. Twenty-eight days.

FIGS. 25 to 27.—Effect of Raw Water Circulating at 90° C.

(3) Freshly formed rust appears to absorb oil and may therefore de-emulsify a weak emulsion. The slush produced can markedly reduce heat transfer rates (*see* later).

The laboratory evidence is incomplete and it is unlikely that reliable data will be secured without trials in full-scale cooling plant. It would appear advisable to maintain the emulsion strength at not less than 0·5%.

Effect of Weak Oil Emulsions in Promoting Pitting.

Many attempts were made to find out whether oil emulsions weaker than 0·1% were likely to cause pitting or intensified local attack, but in most cases the emulsion broke down, and any pitting which occurred was due to the water alone. The higher weight losses shown in Tables II. and III. in experiments with weak emulsions are, it is suspected, in part due to corrosion by water alone. It has been noted, however, that on the few occasions when the dilute emulsions remained stable and corrosion occurred, the attack was local.

Intense attack has been observed at points of contact with foreign bodies, particularly at elevated temperatures, *e.g.*, under the string which attaches the tubular specimens to their support (*see* Figs. 5(*C*), 19 and 20), in 0·5% oil emulsions, as well as in experiments in which 0·5% emulsion is followed by dilute emulsion (Table IV.). The attack is heavier than in water alone, as might be expected, since the attack in water (Figs. 26 and 27) is spread over most of the surface (anodic and cathodic areas large), whereas in oil the only attack is under the string (large cathode, small anode).

The authors' general conclusion is that there are insufficient data to make any definite statement on the occurrence of pitting in dilute emulsions, and no means of preserving a dilute emulsion were found except by the addition of other substances, thus confusing the issue.

The Effect of Oil Emulsions on the Corrosion of Non-Ferrous Metals Alone and in Contact with Mild Steel.

Since non-ferrous metals are frequently included in cooling plant, *e.g.*, bronze impellers in pumps, and it is known that severe corrosion can occur in a system in which dissimilar metals are in contact, it was felt that the

TABLE V.—*Effect of Oil Emulsion on Non-Ferrous Metals Alone and in Contact with Steel. Static Tests at Room Temperature in C.C. Natural Water.*

Experiment No.	Type of Specimen.	Time. Days.	Oil-Emulsion Strength. %.	Loss in Weight. Mg. per sq. cm.	Remarks.
5	Three aluminium plates.	4½	Nil	0·07	White flocculent powder found around points of attachment (probably $Al(OH)_3$). Otherwise, specimens quite unchanged.
9	Three aluminium plates.	4	0·5	0·01	No visible attack.
6	Three copper plates.	4½	Nil	0·13	More or less general staining of specimens. No localised attack.
10	Three copper plates.	4	0·5	0·03	No visible attack.
7	Three brass plates.	4½	Nil	0·13	Specimens fairly generally stained.
11	Three brass plates.	4	0·5	0·02	No visible attack.
43	Copper/steel.	8	0·5	Nil	Evidence of very slight pitting on copper and steel.
44	Aluminium/steel.	8	0·5	0·34	Some general attack but no sign of pitting on aluminium or steel.
45	Brass/steel.	8	0·5	Slight gain.	

investigation into the inhibiting properties of oil emulsions should include experiments with some of the commoner non-ferrous metals and alloys.

The type of specimen used in these experiments is seen in Fig. 8, the two halves being tightly screwed together to ensure good electrical contact. The specimens were suspended from holes in the steel or cast iron. The composition of the various alloys used has already been given.

TABLE VI.—*Effect of Oil Emulsion on Non-Ferrous Metals in Contact With Steel. Flow Tests at Room Temperature in C.C. Natural Water.*

Emulsion strength 0·5% in each case.

Experiment No.	Type of Specimen.	Time. Days.	Loss in Weight. Mg. per sq. cm.	Remarks.
63a *	Copper/steel.	7	Gain.	Slight tarnishing of copper parts of specimens.
63b *	Brass/steel.	7	Slight gain.	Very slight tarnishing of brass parts of specimens.
63c *	Aluminium/steel.	7	0·017	Some emulsion breakdown and tarnishing of aluminium.
112	Aluminium-bronze/steel.	7	Slight gain.	Both parts of specimens bright. No signs of staining.
119	Cupro-nickel/steel.	7	Gain.	Specimens in excellent condition. No signs of attack or staining.
121	Aluminium/steel.	7	Slight gain.	Emulsion began to break down at beginning. Dull film on aluminium. No signs of pits on either part of specimens.
122	Copper/steel.	7	Slight gain.	Emulsion began to break down at once. Copper dulled with fairly heavy oxide film. Steel bright.
132	Copper/cast-iron.		Gain.	Emulsion stable. Copper slightly dulled with thin oxide film. Cast iron unchanged.
103	Manganese-bronze/steel.	7	Gain.	Emulsion broke down. Manganese-bronze slightly dulled. Steel bright.
136	Aluminium-bronze/steel.	16	Slight gain.	Both aluminium-bronze and steel portions of specimens perfectly bright.

* C.F. natural water used in these tests.

The results of experiments under static and flow conditions at temperatures up to 90° C. are given in Tables V., VI. and VII. These experiments cannot be regarded as conclusive because of lack of duplication, but the following comments give some indications of what was found:

(1) Under static conditions at room temperature, 0·5% oil emulsion prevents corrosion of aluminium and steel in contact, but copper, brass and steel in contact are all slightly pitted.

(2) Under flow conditions at room temperature in 0·5% emulsion neither steel nor cast iron suffered attack when in contact with the other metals. Copper, brass and aluminium were slightly tarnished; aluminium bronze, manganese bronze, and cupro-nickel were unattacked.

(3) At 60° and 90° C. in 0·5% emulsion under flow conditions, steel in contact with brass and cupro nickel, and cast iron in contact with copper, suffered slight local attack at the points of suspension. Brass, copper, aluminium-bronze, and manganese-bronze were all tarnished, but cupro-nickel remained bright.

(4) At 90° C. under flow conditions without oil emulsion, steel in contact with aluminium-bronze, manganese-bronze, copper and brass was quite severely attacked, both the bronzes and brass were stained, but copper was unattacked. In the copper/cast-iron couple, the latter was heavily attacked, the copper remaining untouched.

Mechanism of Protection.

U. R. Evans [4] states that it has been shown by Thornhill at Cambridge that small traces of iron salts precipitate the oils from their otherwise stable

TABLE VII.—*Effect of Oil Emulsion on Non-Ferrous Metals in Contact With Steel. Flow Tests at 60° and 90° C.*

Experiment No.	Type of Specimen.	Time. Days.	Temperature. ° C.	Loss in Weight.* Mg. per sq. cm.	Remarks.
			Untreated C.C. Natural Water.		
183*a*	Aluminium-bronze/steel.	7	90	4·45 7·10	Bronze portion slightly tarnished; steel part severely corroded, mainly in areas of local attack.
183*b*	Manganese-bronze/steel.	7	90	8·22 4·66	Bronze portion slightly stained; steel part heavily corroded, mainly in large areas of local attack.
184*a*	Copper/steel.	7	90	8·5 5·4	Copper portions unattacked; steel parts heavily corroded, the attack being entirely general in nature.
184*b*	Copper/cast-iron.	7	90	4·0 5·9	Copper portions unattacked; cast-iron parts heavily corroded, the attack being mainly general in character.
184*c*	Brass/steel.	7	90	...	Brass portions slightly stained; steel parts heavily corroded, mainly in areas of local attack.
			0·5% Oil Emulsion in C.C. Natural Water.		
66*a*	Copper/steel.	8	90	Nil	Copper portions slightly tarnished; steel parts slightly dulled.
66*b*	Aluminium/steel.	8	90	Slight gain.	Thin white film on aluminium portions; steel quite bright.
66*c*	Brass/steel.	8	90	Gain.	A few light-brown spots on brass portions; steel parts bright.
130	Brass/steel.	7	60	Slight gain. 0·33	Slight local attack at points of suspension on steel portions; slight staining of brass portions.
167	Cupro-nickel/steel.	7	60	0·89 0·27	Both portions of specimens quite bright.
173	Copper/cast-iron.	7	60	Slight gain. 0·70	Copper portions heavily tarnished; cast-iron parts unchanged.
168	Manganese-bronze/steel.	7	90	0·51 0·52	Steel parts quite bright, but some local attack under string; bronze portions dulled.
169	Aluminium/bronze steel.	7	90	Slight gain.	

* Total loss in weight. Duplicate results bracketed.

emulsions, and it is therefore possible that the oils are thus precipitated at susceptible points (anodes) by the iron salts there formed. This preferential deposition would be a great thermal advantage since it should not interfere with heat transfer. Buchholtz, and Krekeler [5] and also Hauttmann [6] state that the use of emulsifying oils does not adversely affect heat transfer, but they picture the formation of a continuous oil film.

The authors' own observations have confirmed Thornhill's work in showing that small amounts of iron salts coagulate the oil emulsions. It has also been shown that, when submitted to a potential difference, the oil separates at the positive pole so that the emulsified particles are negatively charged. Furthermore, when a drop of oil emulsion is placed on a clean iron surface it develops a white yolk in the course of a few hours and the yolk forms in the centre of the drop, *i.e.*, at the anode on the metal surface. These observations lend strong support to the theory put forward by Evans and by Thornhill.

The experiments on the effect of small amounts (2% and 0·5%) of oil in water emulsions on heat-transfer (*see* later) show that when the metal surface is clean the heat-transfer coefficient is unaffected, which is in agreement with the statements of Buchholtz, Krekeler and Hauttman. These results do not, in the authors' opinion, rule out the possibility of film formation, since the film might be too thin to affect the determinations.

To investigate the matter further, flat-plate specimens were examined under a mercury lamp after they had been immersed in 0·5% oil emulsions containing oil-soluble fluorescent dyes. These observations were, however, inconclusive, because little or no fluorescence could be seen. On the other hand, the nature of the emulsions and the experiments in oil emulsion followed by water, which have shown that passivity persists for an appreciable time when oil emulsion is changed for raw water alone, seem to support the film theory.

The analysis of the soluble oil used for our experiments shows that it contains an emulsifying agent believed to be a sodium salt of a sulphonated mineral oil. Some such highly polar body is likely to be used in all " soluble oils " and in the emulsion it will be adsorbed into the oil-water interface. It is difficult to believe that under these circumstances adsorption on metallic or other surfaces will not occur with the formation of a film of oil over both anodic and cathodic surfaces.

Chyżewski and Evans,[7] in experiments on the classification of anodic and cathodic inhibitors, have shown that when emulsifying oils were added to 0·1 molar solutions of K_2SO_4, marked inhibition was obtained at the anode, whilst there was no influence or only weak inhibition at the cathode. In similar experiments with a divided cell, using a much more dilute salt solution (C.C. water containing about 700 mg. of total dissolved solids per litre), the strong inhibition at the anode was confirmed, but, in addition, a well marked but weaker effect at the cathode was observed.

To sum up, it appears reasonable to believe that protection is secured by the formation of a thin film of oil adsorbed on both anode and cathode, but that, if this film is too thin to prevent the formation of anodes at specially susceptible spots, the oil film is thickened at these places by coagulation of emulsion to such a degree that the corrosion is stifled.

Properties of the Emulsions.

No careful study of the properties of the emulsions was made, but some observations obtained during the experimental work are recorded here.

When the soluble oil is added to water it forms a surface layer of oil, but this is rapidly emulsified on stirring by hand. Some soluble oils emulsify much more readily than others. No data regarding particle size are available, but it is probable that emulsions made up in this way at different times differ to some extent in stability and particle size, so that for any future series of experiments it may be advisable to adopt some form of standard stirring or to employ a colloid mill in preparing the emulsions. When made up by hand the emulsions often tend to separate a layer of oil on standing for periods of one week or more, but this does not happen if they are kept stirred. They can be evaported down to half bulk without separation.

A more satisfactory method of making up emulsions is to pipette the required quantity of oil into a small portion of the water to be used, warmed to a temperature of 50–60° C., shake vigorously to emulsify and add to the remainder of the water.

With hard waters a scum forms slowly, and this is no doubt due to the formation of calcium and magnesium soaps from the hardness salts and the emulsifying agent. With soft water, as might be expected, there is considerable foaming, but no scum. Both of these phenomena might be objectionable in full-scale plant. There is evidence that the addition of substances such as " Calgon " or tannins, which are known to form soluble complexes with calcium and magnesium ions, can reduce the scum, but further experiments are necessary.

It was found on several occasions that the emulsions, particularly

those of low concentration, broke down during corrosion experiments; this breakdown is attributable to three factors:

(1) The removal of emulsifying agent as scum, followed by adsorption of oil by the scum, or conversion of the emulsion into a water-in-oil type.
(2) Adsorption of oil and emulsifier by surfaces of glass or metal, or by insoluble corrosion products.
(3) Coagulation of emulsion by small amounts of metallic ions.

Further work is needed to establish the conditions under which dilute emulsions can be kept stable, since these are the least costly to use. The risk of breakdown of dilute emulsions in contact with metals appears to be less if there has been a preliminary treatment with a more concentrated emulsion.

Experiments With Emulsifying Agents Alone.

In order to gain some idea of the contribution of the emulsifying agent to the inhibiting properties of a soluble oil some static immersion tests were carried out, using a known emulsifying agent made into a soluble oil by mixing with a good grade lubricating oil, and the same emulsifying agent alone, in the same quantity as in the soluble oil. The weight losses were much greater with the emulsifying agent alone than with the soluble oil.

Effect of Oil-Emulsion Treatment on Heat Transfer.

Since the protective action of the oil emulsions may involve the formation of a film on the surface of the metal, the effect of oil emulsions on heat transfer in the case of clean tubes and dirty rusty tubes was determined. These experiments led to the following conclusions:

(1) With an initially clean tube the addition of 0·5% and 2% oil emulsion did not affect the heat-transfer coefficient. (In the absence of inhibitor a considerable loss in heat-transfer rate would have occurred.)
(2) When oil emulsion was introduced after water had been in circulation with rusty tubes, the heat-transfer coefficient showed a slight improvement at first, but after a comparatively short time there was a steady decrease. Over a period of three weeks the coefficient fell to less than 60% of its original value.

Conclusions.

(1) Oil emulsions can be effective inhibitors of the corrosion of mild steel in natural waters of widely differing types, at temperatures up to 90° C., under both static and flow conditions.
(2) Metal surfaces which have been exposed to oil emulsions retain their passivity in untreated water for a limited time. Satisfactory protection can be obtained by pretreatment with a heavy dose followed by subsequent continuous application of a much lighter dose.
(3) Non-ferrous metals alone or in contact with mild steel are protected by oil emulsions both under static and flow conditions at room temperature. At 60° and 90° C., however, local attack takes place on both steel and cast iron in contact with non-ferrous metals.
(4) With mild steel or non-ferrous metals it is of primary importance to maintain a stable emulsion, since breakdown may be accompanied by pitting. Attention must also be drawn to the intense corrosion which occurs at 90° C. at points which may be relatively inaccessible to the oil, *e.g.*, the groove under the supporting string seen in Figs. 19 and 20. It would appear from comparison with a similar groove in Fig. 27 where no

oil emulsion was used that the loss of thickness at these points was greater in the presence of oil emulsion. A similar groove where the specimen touched a glass rod is shown in Fig. 26. This is another view of the specimen in Fig. 27. There is, therefore, a potential danger in actual plant if debris sticks to the metal surface.

(5) The progress of corrosion is definitely retarded and possibly stopped when oil emulsion is in contact with metal surfaces on which the corrosion products are still present. The most efficient protection is obtained, however, when the metal surfaces are clean, and under these conditions less oil is likely to be needed to maintain a stable emulsion.

(6) With clean metal surfaces exposed to oil emulsion the heat transfer is only slightly affected, but it should be borne in mind that, if untreated water is used, a much greater fall in heat-transfer rate will rapidly occur as a result of the formation of rust. The use of oil emulsion in previously corroded tubes improves the heat-transfer coefficient for a short time compared with that obtained with natural water alone. There is, however, a subsequent decrease in the efficiency of heat transfer.

(7) Evidence is presented which appears to show that the inhibiting properties of the oil emulsion may be due to the formation of a thin film of oil on the metal surface, with thickening at specially susceptible anodic points.

(8) Any treatment of cooling water for the prevention of corrosion must take into account the formation of scales and slimes. The full effect of these on the prevention of corrosion by oil emulsions or other inhibitors has yet to be investigated. It appears advisable to remove scale- and slime-forming constituents from the water before use in cooling systems.

(9) The results given in this paper must be regarded as a preliminary survey of the subject, but, in the authors' opinion, sufficient evidence has been accumulated to show that the soluble mineral oil used is worthy of consideration for trials in full-scale plant.

(10) While the effects of a light-grade mineral-oil emulsion have been considered herewith, it might be useful to extend the study to vegetable and animal oils. It might also be an advantage to have a more viscous film for high-temperature work, but the authors were not successful in forming stable emulsions with heavy lubricating oils, and time did not permit a further study of these aspects.

Acknowledgments.

The authors wish to acknowledge their indebtedness to Dr. U. R. Evans for his help and encouragement, and to L. H. Tomlins and G. Drinkwater of this laboratory for suggestions and much painstaking experimental work.

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CORRESPONDENCE.

Dr. T. P. HOAR (Corrosion Research Section, University Chemical Laboratory, Cambridge) wrote: I have for some time been interested in a corrosion system rather similar to that studied by Messrs. Hamer, Powell, and Colbeck, and it is satisfactory to find that our results appear to be in very substantial agreement. Our experiments at Cambridge have been conducted mainly with a high-carbon, low-chromium steel, exposed to a variety of soluble-oil emulsions, made with distilled water, at 25° C.; the general result, for both "milky" and "clear" emulsions, made from either commercial or laboratory-prepared soluble-oil concentrates, is that 0·5% of oil in the emulsion prevents attack to a very great extent, although 2·0% is usually necessary to prevent all visible traces of pitting.

We have found that proper emulsification of the oil concentrate is all-important, and that various oils differ markedly in their ease of emulsification. Mr. Hamer very kindly made available a sample of his oil for our work, and our first results with it were highly disappointing—very considerable attack occurred, and 2·0% emulsions were worse than 0·5%. This discrepancy was traced to our method of emulsification—hand-shaking with cold distilled water. This is satisfactory for most of the oils that we have studied, but gives very unstable emulsions, having the poor corrosion properties just referred to, with Mr. Hamer's oil. When we altered our technique by emulsifying in *warm* water, we found the oil to give emulsions at least as good as any others that we have tested. In this connection, I note that the oil concentrate is stated to contain mineral oil and petroleum sulphonates only; many commercial oils contain at least a proportion of an oleate (or similar) soap, and there seems to be good evidence that such emulsifiers assist in easy emulsification in the cold. It is, however, fairly clear that very good, perhaps optimum, results from the corrosion point of view are obtained from emulsions containing petroleum sulphonates as an ingredient, provided that proper emulsification is secured by one means or another.

The authors' suggested mechanism for the corrosion inhibition—marked anodic inhibition due to precipitation of the emulsion at anodic points, and lesser cathodic inhibition due to oil deposition all over the surface—appears very probable. They state that in a divided cell of the type used by Chyżewski and Evans * they found a marked influence of the soluble oil at the anode and a smaller but still important influence at the cathode. They do not, however, state the current densities prevailing, and it is always possible in experiments of this type that current densities far removed from those of "natural" corrosion occur, so that there is a slight doubt as to whether the effect observed would also be present in "natural" corrosion. However, corroborative evidence can be obtained by the measurement of the single-electrode potential of a corroding specimen; if the addition of an inhibitor causes this to become more noble, anodic inhibition is predominating over cathodic, as I have pointed out elsewhere.† In many hundreds of measurements on various steel/soluble-oil-emulsion systems, I have always found that the potential becomes more noble (*a*) on the addition of oil, (*b*) in the presence of higher concentrations of oil, and (*c*) with the passage of time—so long as no corrosion is visible. Thus, there is no doubt that the inhibition is more anodic than cathodic. The fact that the area of attack is diminished by an increase of soluble-oil concentration (in the dilute emulsions that are not entirely

* Chyżewski and Evans, *Transactions of the Electrochemical Society*, 1939, vol. 76, p. 215.

† Hoar, *Chimie et Industrie*, 1939, vol. 41, p. 118 c.

protective) is further evidence of the restriction of the anodic reaction; and it is right to emphasize that, in consequence of this restriction, soluble-oil emulsions too dilute to give complete protection may lead to an increase in the rate of corrosion penetration in the well-known way.

It seems very likely that the foundation of the thin film of oil deposited on the oxide-film *cathode* surfaces, giving the moderate cathodic inhibition, is an oriented monolayer of the petroleum sulphonate, soap, or other emulsifier, with the polar group attached to the oxide-film surface. One may imagine that a spherical oil particle, encased with the amphipathic emulsifier, (*a*), Fig. A, hits the oxide-film surface and spreads out (*b*), owing to the displacement of water molecules from the surface by the sulphonate or carboxylate groups. Eventually, it is likely that a nearly continuous layer is formed (*c*), offering some resistance to the passage of oxygen molecules, and considerable resistance to the passage of water molecules, to the cathodic surface; the oxide-film surface has been rendered hydrophobic by a process akin to that operative in mineral flotation.

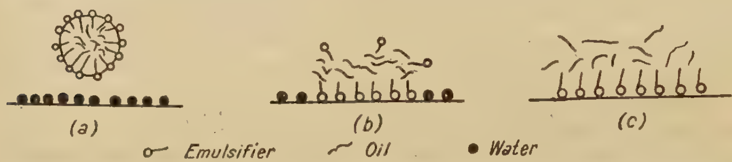

Fig. A.—Formation of a Thin Film of Oil on the Oxide-Film Cathode Surfaces.

We may note also that a similar oriented adsorption of oil-soluble petroleum sulphonates from oily media on to oxide-film surfaces is probably responsible for the improved corrosion-preventing properties of slushing oils and other temporary protectives containing sulphonates.

Mr. G. McNeill (Messrs. Colvilles, Ltd., Motherwell) wrote: I also have found, in common with other investigators, the difficulty of reproducing the pitted type of corrosion of the severity usually found in industrial practice, but it can be demonstrated in a less severe state. The paper indicates that the type of water has not a great bearing on whether the system will, or will not, be subject to corrosion attack such as pitting.

I am very much struck by the paragraph on the "Mechanism of Protection." It mentions that "small traces of iron salts precipitate the oils from their otherwise stable emulsions, and it is, therefore, possible that the oils are precipitated at susceptible points (anodes) by the iron salts there formed. This preferential deposition would be a great thermal advantage, since it should not interfere with heat transfer." Does this not indicate that oil is the main factor in giving ideal conditions for the development of pitting?

The authors' conclusion, drawn from the work done, indicates that the water is the cause of the pitted type of corrosion. My interpretation of the paper is that the presence of the oil is the condition necessary for producing pitting. My experience, over many years, has been mainly concerned with failures due to pitted corrosion. I have always found that oil has been present in every case where pitting has occurred; it has always been confined to a particular piece of plant, and the other plants using the same water have been free from the trouble.

I have just received a length of locomotive boiler tube which is a mass of pitting; it has been in use only one week. Do the authors consider that water alone could produce such selective action on a piece of steel

in such a short time? Must there not be some other influence to cause such selective and intensive action? Furthermore, I am at present interested in intensive pitting in a hydraulic water system. It is usual to add soft soap to the water for lubricating and this has been used for years on some plants without any sign of abnormal corrosion. In this particular case there was difficulty in procuring the soft soap and a soluble oil was used. The result has been very serious trouble due to pitting.

The use of oil for keeping polished steelwork from rusting is so impressed on our minds that it is difficult to realize that when a water-oil phase is produced the action is reversed.

In view of my comments I would be very reluctant to use a corrosion inhibitor which contains any free oil and would be pleased to have the authors' views on my remarks.

Dr. J. E. O. MAYNE (Corrosion Research Section, University Chemical Laboratory, Cambridge) wrote: The authors have shown in a very clear and convincing manner that emulsions of oil in water can act as corrosion inhibitors. The way in which these emulsions inhibit, however, is by no means clear. It is unfortunate that details of the composition of the emulsifying agents used have not been published, since it is always difficult, and sometimes dangerous, to interpret results obtained with unknown commercial products. Two facts, however, are known: the emulsions had a negative charge, and the calcium salts, and therefore, presumably the iron salts, of the emulsifying agents were insoluble in water.

Stability of the emulsion is due to the negative charge on the particles, and this may be reduced or destroyed in at least two ways:

(1) The charge may be reduced in the bulk of the emulsion by interaction of either:

Ferrous or ferric ions with the emulsifying agent, with the formation of insoluble products; or

Positively charged ions, or ferrous or ferric hydroxide particles, with the negatively charged oil particles.

Either of these reactions would lead to unstable oil particles, which would coalesce and deposit on the nearest surface. They would, however, adhere preferentially to the steel, owing to the polar nature of the remaining emulsifying agent. Protection may then be due to either:

(*a*) The formation of films of insoluble iron salts; or

(*b*) The formation of an oil film, as suggested by the authors.

It would, therefore, be interesting to know to what extent deposits of oil were formed on the glass walls of the apparatus used.

(2) The destruction of the charge may not take place in the bulk of the solution but at the anode where the ferrous salts are formed; this reaction would be assisted by cataphoresis, and protection may be due to either (1(*a*)) or (1(*b*)).

Modern emulsifying agents are available, which, it has been claimed, are unaffected by considerable concentrations of electrolytes, since they form soluble complexes with divalent and trivalent ions. These claims have been confirmed in a number of cases by King and Wrzeszinski.* It would, therefore, be very interesting to know if emulsions prepared with similar emulsifying agents have been tested, since it would show whether protection was due to (*a*) or (*b*) above.

* *Transactions of the Faraday Society*, 1939, vol. 35, p. 741.

The authors have shown that the emulsifying agent dissolved in water is not as effective as the same quantity in an oil emulsion. This does not rule out the possibility of the protection being due to the emulsifying agent, since it is strongly adsorbed on the surface of the oil particles, and may be delivered at the anode in a greater concentration than that obtained with an equivalent solution. The oil may, therefore, be only a carrier.

With respect to the deposition of the emulsion at the cathode, this may take place in accordance with suggestion (1) above. In the divided cell, however, it would be interesting to know what precautions were taken to remove all anodic centres on the cathode, since only a small number of local anodes would account for the observation of the authors.

The part played by cataphoresis could be investigated by the use of positively charged oil emulsions, and I should like to know if the authors have done any work along these lines.

AUTHORS' REPLY.

The AUTHORS wrote in reply that they were very interested in the comments of both Dr. Hoar and Dr. Mayne on the possible mechanism of protection by oil emulsions. Unfortunately, at the time when the work was done, they were unable to give this matter the attention which it deserved.

In reply to Dr. Hoar, no attempt was made to estimate the current densities prevailing in the divided-cell experiments. They agreed that the current densities in this type of cell might be far removed from those in "natural" corrosion, because of the use of copper cathodes and zinc anodes. In these experiments they were following the lines adopted by Chyżewski and Evans, who also found some inhibiting effect at the cathode when using soluble oil. They agreed, however, that the use of the single-electrode potential measurement as described by Dr. Hoar and applied to corroding specimens would have been a better method of following the course of the inhibition.

In reply to Dr. Mayne, although no attempt was made to measure the amount of the oil deposits on the walls of the apparatus used, there seemed to be no doubt that such deposits were present, as judged by the greasy feel and the fact that glassware required cleaning with solvents before it became readily wettable by fresh water. The authors had not carried out any experiments with the emulsifying agents described by Dr. Mayne, but it was possible that the simple addition of calgon would have a similar effect. Neither were any experiments carried out with positively charged emulsions. They pointed out that in the divided-cell experiments copper plates coupled with steel were used to study the effects on the steel as anode and zinc coupled with steel to study the effect on steel as cathode. In the former case, was it to be expected that anodic areas would develop on the copper? In the latter case it was presumably conceivable that anodes might develop on the steel cathode. In neither case were any precautions taken to prevent the formation of anodic areas on cathodes, but individual experiments were all of short duration, lasting only 15 or 20 min.

Whilst the authors agreed with Mr. McNeill that the presence of oil could under certain conditions give rise to pitting, it was also well known that pitting could occur in the absence of oil.

The authors felt unable to comment fully on the statement referring to the pitting which occurred on a locomotive boiler tube during the course of one week. Mr. McNeill gave no information regarding the depth or distri-

bution of the pits, whether this was the only tube in the boiler which suffered attack, from what part of the boiler the tube came, or, finally, whether the boiler was fitted with a copper fire-box. Broadly speaking, the authors thought that all that could be said was that if the pits were not deep they would not be surprised that the water was the sole cause, but if they were deep they would suspect some additional influence. In describing the intensive pitting found in a hydraulic system, Mr. McNeill did not state what strength of oil emulsion was used or whether any steps were taken to prevent dilution of the emulsion during service. Pitting could occur in the presence of dilute emulsions.

SOME OBSERVATIONS ON THE AUSTEMPERING AND ISOTHERMAL TRANSFORMATION OF STEELS, WITH SPECIAL REFERENCE TO THE PRODUCTION OF MARTENSITE.*

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(Figs. 27 to 45 = Plates XIII. to XVI.)

Paper No. 16/1944 *of the Alloy Steels Research Committee.*

SYNOPSIS.

Employing the sensitive recording dilatometer already described (Stanton, *Journal of The Iron and Steel Institute*, 1943, No. I., p. 95 P) in conjunction with hardness measurements and microscopic examination, the transformation of austenite at temperatures in the martensitic zone was investigated.

A theoretical explanation based on the rate of recrystallisation is offered to account for the shape of the typical S-curve representing the isothermal transformation of austenite; from the results of experimental work it is suggested that the retention of austenite and the subsequent "period of induction" depend on the relaxation of stresses set up in the material at the time of quenching.

The experimental work covered the effect of the mass of the specimen, the nature of the steel, its pretreatment and the effect of initial quenching temperature and of variation in that temperature, together with the implications arising from the latter two points with relation to grain size.

The variation in behaviour on isothermal transformation of apparently identical specimens was considered, and it is suggested that this may be due to some variation in the material not taken into account in ordinary chemical analysis.

Various possibilities accountable for the shape of the austenite transformation curve are reviewed, and, whilst the nucleation theory is not ruled out, it does not seem possible that the transformation proceeds according to any one specific physico-chemical reaction.

PART I.—THEORY OF ISOTHERMAL TRANSFORMATION.

Introduction.

THE earliest work on the quenching of steel in metal baths appears to have been carried out in connection with the lead-patenting ("sorbitising") of wire rod, the beneficial effects of which have long been appreciated. That a similar process can, in certain circumstances, be usefully applied to ordinary heat-treatment operations is clearly suggested by Brearley [1] in the following words: ". . . it should be possible to arrange the quenching operation in such a manner that hardening and tempering could be accomplished in one and the same liquid. If, for example, it were desired to temper a hardened object back to 350° C., it might be thought that direct quenching in molten lead at that temperature would be equally as effective as first hardening in water and then tempering. And so it would if the rate of cooling in the first instance from the quenching tem-

* Received April 13, 1944. This paper is published by authority of the Alloy Steels Research Committee. The views expressed in it are the authors', and are not necessarily endorsed by the Committee as a body.

perature down to 350° C. were as rapid in lead as in water. As a matter of fact it is very much slower, and the use of lead or any similar metal or alloy, or other heated substance is practicable only in comparatively few cases where great toughness and a very moderate degree of hardness is sufficient; in such cases, however, it is both simple and safe." For the time at which this was written, not later than 1916, it represents a very far-sighted view.

The foundations of the metallographic study of the process were laid by Dejean [2] and Chevenard,[3] the latter of whom introduced the nomenclature which is generally adopted today. Portevin and Garvin,[4] in a research of the greatest importance, in addition to quenching their specimens down to room temperatures, also carried out experiments on "interrupted" quenching, in which the steel was removed from the quenching medium before the changes were complete, the forerunner of the process of quenching in a bath maintained at some intermediate temperature. In their conclusions these workers state that "the formation of troostite corresponds with a rapid transformation at a high temperature . . . and that of martensite with a transformation which is relatively slow as compared with the former and takes place at a low temperature. . . ."

Modern "austempering" treatments may be considered to have germinated, however, from the time when Dartry Lewis [5] showed that the transformation of austenite to martensite could be delayed for several minutes by carrying out the quenching operation in a salt bath at 232° C., and suggested that practical use might be made of this fact. It was the work of Davenport and Bain,[6] however, which brought the possibilities of the process to general notice, and their determination of the shape of the so-called S-curve has proved of immense value, not only with respect to austempering itself, but also in connection with heat-treatment processes in general. Many other workers have contributed in varying degrees to our knowledge in this field, the literature of which is now considerable.

The present researches, which have extended over nearly three years, are still very far from complete, but in view of the importance of the subject at the present time, it may be worth while to place on record the results which have so far been obtained, even if some of them require more examination, and the ideas to which we are tending, even if some may later prove to be unsound. It has become very clear that, simple as the technical process may appear at first sight, it is, in fact, subject to certain disturbing factors; the difficulties of reproducing even one's own results, and still more those of other investigators, are often formidable.

Experimental Technique.

The Testing Apparatus.

In earlier work microscopic examinations in conjunction with measurements of hardness, and magnetic, dilatometric and electrical resistivity determinations were employed. After very careful consideration it was decided that for the present purpose the progress of the decomposition of the austenite could be most readily followed by dilatometric measurements. For this a very sensitive instrument was designed and constructed by one of the authors (L. R. S.) (*see* Fig. 1), a full description of which has already been published.[7] Changes of length of the order of 0·001 mm. may readily be measured and one-tenth of this determined by estimation. The length of the sample is normally recorded autographically every 2 sec. In order to obtain sufficient stability to justify measurements of this order, the silica tube, on the sealed end of which the specimen to be treated rests, must be quite rigid; the walls are about 0·1 in. thick and the external diameter is $\frac{3}{4}$ in. A series of $\frac{1}{4}$-in. holes was made in

that portion of the tube over which the specimen extends in order to permit ready access of the quenching medium.

Alongside these dilatometric measurements, the microstructures and Vickers pyramid hardness numbers were recorded. The authors are not unaware of the practical importance of some indication of the toughness of the treated samples, but for the immediate object of this research it did not appear that it would add anything to the information for which they were seeking. Further, the short stubby specimens which they have mainly employed are not very suitable for such determinations; in their view, it would have been most undesirable to make modifications—by increasing the length or by the provision of a notch, for instance—which would have rendered the specimen much less satisfactory for the work on the factors involved in the process itself.

To obtain a surface for the microscopic and hardness measurements a small flat was prepared along the length of the sample, while, in addition, many specimens were later sectioned to enable the change of structure from surface to centre to be followed.

The dilatometer generally employed was designed for work over comparatively long periods, from several minutes to a day or more. To supplement these measurements, particularly in connection with the changes which go on in the earliest stages of the rapid, troostitic transformations, the push-rod of the dilatometer was arranged so as to operate an optical lever, the movements of the spot of light from which were recorded photographically on a rotating drum. This, where necessary, provided a record of the changes occurring in the specimen in the first minute or two, after which the normal dilatometric recordings were sufficient.

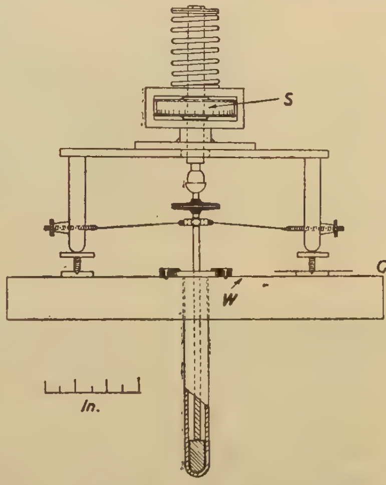

FIG. 1.—The Dilatometer and Water-Cooler.

The Quenching Apparatus.

In the design of the quenching apparatus two main factors were regarded as being of pre-eminent importance, namely, great rigidity and the reduction of the personal element to a minimum.

The silica tube containing the specimen (*see* Fig. 1) is held by its flange against the upper surface of the water-cooler W, which, in turn, stands on the table A (*see* Fig. 2), through a hole in which the tube projects. The table A is supported on three massive legs, which are screwed to the base-plate of the whole apparatus.

The two furnaces F_1 and F_2 (*see* Fig. 27), in which the initial heating and the subsequent quenching respectively are carried out, are situated on the semi-circular table B, Fig. 2, which can be rotated in a horizontal plane so that either F_1 or F_2 is centrally disposed with respect to the specimen tube. Further, B can be raised or lowered on the pillar E by means of the lever L. All these movements, horizontal or vertical, are controlled by stops which render them both rapid and almost automatic.

In order to reduce to a minimum any vibration transmitted to the dilatometer during the transference of the sample from the heating to

the quenching furnace, the movements were made as smooth as possible. The table B rotates on three small rollers D about the point C, and the lever L, which raises or lowers it, is exactly counterbalanced. The specimen and the dilatometer remain stationary throughout, it being the furnaces which are interchanged; the advantages of this have also been appreciated independently by Flinn, Cook and Fellows.[8]

Furnace F_1 is wound to give a maximum temperature of about 1000° C., and during the experiments is constantly traversed by a stream of pure dry nitrogen to minimise oxidation. The temperature is measured by means of a platinum/platinum–10%-rhodium thermocouple, the hot junction being on the same level as the centre of the specimen.

The quenching bath F_2 is a fireclay pot, 7 in. long and 4 in. in dia., wound with Nichrome wire and with an auxiliary heating element under

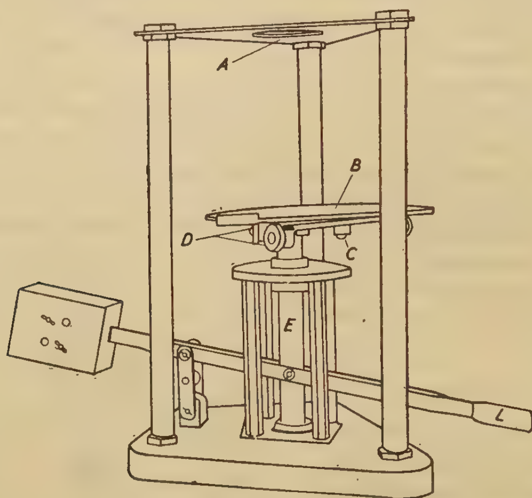

FIG. 2.—The Quenching Apparatus.

the base; it is well lagged with kieselguhr. The heating coils are so arranged that they can be connected either in series or in parallel, the temperature being measured by a copper-constantan thermocouple protected by a fireclay sheath. The thermal losses from the surface are minimised by a well-fitting asbestos lid, split into two halves, with the holes necessary to take the specimen tube and thermocouple sheath. In the earlier experiments the bath temperature was controlled by means of a mercury-in-steel thermometer operating a relay, which short-circuited a resistance connected in series with the furnace windings, the temperature variation being of the order of $\pm 2\frac{1}{2}$° C. In the later work, in order to reduce this fluctuation still further, personal control of the external resistances was substituted, the temperature changes being cut down to $\pm \frac{1}{2}$° C.

In some preliminary tests a second thermocouple was inserted into a small hole drilled in the centre of the specimen; after the first 10 min. of heating its reading varied by less than 5° C. from that of the furnace couple, whilst after 10 min. at temperature no recordable difference was to be observed.

For the quenching bath the quaternary lead-tin-bismuth-cadmium

eutectic and the lead-tin eutectic were used. The former was tried in anticipation of quenchings at or below 100° C., but has been little used. It suffers from the two defects of rapid oxidation, with the production of a scum which may retard the quenching operation, and, more troublesome still, a marked expansion on solidification, which may cause fracture of the container. Comparable experiments showed no appreciable difference in the quenching efficiency of these two media.

The introduction of the heated silica tube and metal specimen, which latter normally weighed about 7 g., into the quenching bath, weighing about 5000 g., caused some increase in the temperature of the latter. In the case of quenchings from 800° C. in a bath at 200° C. this temperature rise amounted to about 12° C., which fell to 3° C. in 15 min. With a bath temperature of 300° C. the rise was around 6° C., which became steady after another 10 min. To improve this state of affairs, pieces of the cold alloy or steel were introduced into the quenching bath immediately before the quenching operation; as a result the temperature remained essentially constant around 300° C., and rose only 5° C. at 200° C. In the latter case the temperature returned to its original value in 5–10 min.

These facts must be borne in mind in connection with the initial portion of the dilatometric curves, where the rate of expansion will at first be somewhat retarded. In the case of a treatment in which the steel remains in the fully austenitic condition in the bath, the same effect will result in an apparent slow contraction for the first ten minutes or so after quenching.

The Dimensions of the Specimens.

After a number of preliminary experiments, specimens 0·3 in. in dia. and 0·875 in. long were standardised for the majority of the tests. One end is rounded to sit firmly on the bottom of the silica tube; into the other end a hole $\frac{1}{4}$ in. deep and $\frac{1}{16}$ in. in dia. is drilled, into which is inserted a steel wire which passes up the dilatometer push-rod.

The early use of specimens 2 in. in length was soon discontinued, owing to the impossibility of quenching them uniformly in this apparatus. The question of the dimensions of the sample will be reconsidered at length in a later section.

The Testing Procedure.

In raising the specimens to the quenching temperature, which took 40–50 min., the dilatometer was in continuous operation, and any abnormality was at once revealed.

The cycle of operations after the specimen had been set up in the silica tube and the dilatometer recorder put into operation was as follows:

(1) The recorder chart is calibrated in steps of 0·01 mm. by means of the screw S on the dilatometer, Fig. 1.

(2) The table B is placed with the heating furnace F_1 directly under the specimen tube, raised into position round it and the top closed by two half-plugs. Dry nitrogen is passed through to sweep out the air and the heating current switched on. This heating can be slow or rapid as desired, a matter of importance in the tempering tests to be described later.

(3) When the desired temperature is approximately attained the resistances are controlled to maintain it within a degree or so for the required soaking period.

(4) When the moment for quenching has arrived, the table is lowered steadily against a stop, swung over against a second stop,

which brings the quenching furnace directly under the specimen, and then quickly raised again. Photographic tests showed that the complete operation was effected in about a couple of seconds.

After more than a year's use of this apparatus the authors are completely satisfied with its performance.

The Steels used.

The analyses of the steels used in this investigation are given in Table I.

TABLE I.—*Analyses of Steels used.*

Steel.	C. %.	Si. %.	Mn. %.	S. %.	P. %.	Ni. %.	Cr. %.
A . . .	0·82	0·23	0·32	0·027	0·011	...	...
D . . .	0·44	0·21	0·66	0·020	0·016	3·09	1·30
L . . .	1·19	0·22	0·47	0·010	0·018	...	...
K . . .	0·32	0·20	0·21	...	...	0·14	12·5

Reproducibility of Results.

Before proceeding to discuss the variations in the behaviour of different materials after varying treatments it is obviously imperative to demonstrate that identical dilatometric curves are obtained when the same material is given the same treatment.

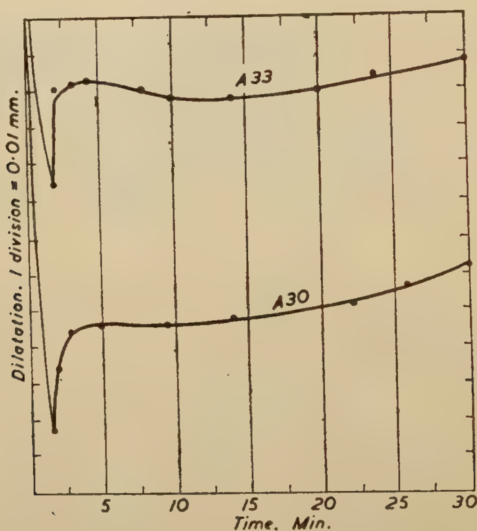

FIG. 3.—Dilatation/Time Curves for 0·3-in. dia. "Identical" Specimens of Steel *A*, quenched from 845° C. in a bath at 200° C.

In Fig. 3 are recorded the curves for duplicate quenchings made on untreated samples machined side-by-side from the eutectoid plain carbon steel *A*. These were successively heated during 50 min. to a temperature of 845° C., kept there for 10 min. and then quenched in the lead-tin eutectic

at 200° C. The curves are essentially identical. At the moment of quenching, both samples underwent a marked contraction to a length less than that at which they started, showing that they were still, in part at any rate, in the austenitic condition. This was followed by an immediate rapid expansion which came to an end in about 2 min., and then by a period of about 10 min. during which there was very little change. This was succeeded in turn—12 min. after the quenching—by the commencement of a slow decomposition of the residual austenite, with its concomitant expansion.

It is, then, quite clear that the apparatus is capable of giving curves of a high degree of reproducibility, and, therefore, that observed variations must be due either to differences in the material or in its treatment.

These results agree very satisfactorily with those of Davenport and Bain [6] on a material of similar composition, as is also the case for hyper-eutectoid steel *L*.

At the same time, there are many instances where the results for materials of closely similar ordinary analysis, even when tested on the authors' apparatus, were markedly dissimilar. A good example of this is shown in Fig. 4, which refers to tests on samples of steel of analysis

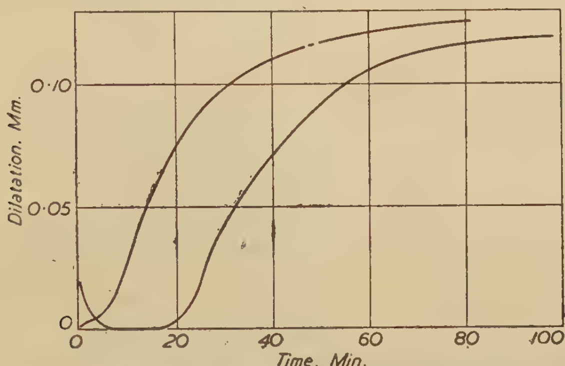

FIG. 4.—Dilatation/Time Curves for 0·3-in. dia. Specimens of Steel corresponding to *D*, quenched from 810° C. in a bath at 280° C.

corresponding to *D* when quenched from 810° C. in a bath at 280° C.; the rates of transformation are very clearly different.

Other workers have, the authors believe, experienced a similar variation in the isothermal transformation of steels of identical ordinary analysis but from different casts.

Effect of " Oxygen " and Abnormality.

An explanation—at present lacking sufficient experimental support—of this failure to reproduce results obtained on material of similar composition but at different times, and also a possible explanation of " abnormality," may be put forward if it be assumed that some constituent of the material, not normally estimated, has the effect of moving the S-curve bodily to the left. (In certain temperature ranges a comparatively small shift may modify the results very considerably indeed.) The authors throw out the suggestion, for what it is worth, that the element involved, directly or indirectly (say, through its reduction of the " effective " manganese or silicon content of the steel or of the formation of non-

metallic inclusions which may act as nuclei), is oxygen. If it be thought improbable that the addition of a new element should move the curve to the left, *i.e.*, increase the reactivity of the austenite, when most additions shift it in exactly the reverse direction, attention may be drawn to the influence of cobalt, which is agreed[9] to be of this character.

In Fig. 5 this hypothesis is illustrated diagrammatically. If curve II. be the "inherent" characteristic of a steel of given ordinary analysis, which, owing to the presence of "oxygen," is transposed to the left, curve I., and if III. be the cooling curve associated with a given treatment then this latter will intersect the first curve, II., at the point A and yield a martensitic structure, whilst in the second case intersection takes place at B, at which temperature the material will transform to a pearlitic constituent. The authors are aware that the time element in the case of curves I. and II. is not identical with that for curve III., but the nature of the argument is not essentially modified thereby.

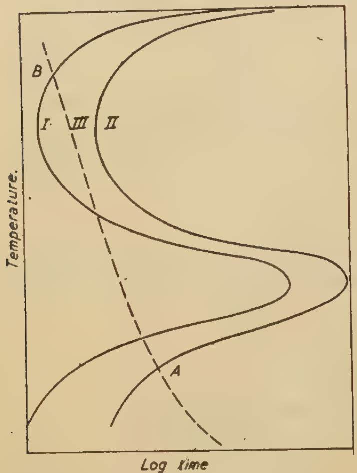

FIG. 5.—Effect of "Abnormality" on the S-Curve.

In this connection, it is relevant to refer to results obtained by Roberts and Mehl[10] on the rate of formation of austenite from pearlite, the inverse of the transformation which we are considering. These workers have shown that the velocity of change is markedly different in two steels, one of small, the other of large grain size, but of almost identical "ordinary" analysis. The rate of growth of the austenite (from pearlite) is "appreciably higher for the fine-grained steel," where, as a result of the addition of aluminium, the degree of deoxidation should be the greater. This is quite consistent with the suggestion just made that "oxygen" in a steel shifts the S-curve, *i.e.*, modifies the stability of austenite at temperatures below A′, and, therefore, the velocity with which the transformations take place.

As an example of a vertical movement of the curve in steels of very similar composition may be cited results obtained by Davenport and Bain[6] and the present authors, respectively, on a stainless steel (Fig. 6). Working with a steel of which the composition was merely recorded as 0·35–0·4% of carbon and 12–14% of chromium, the American workers recorded a maximum stability of the austenite just above 200° C. In the authors' experiments the material (steel K) reacted at 206–210° C. even more quickly than did a plain carbon steel at a similar bath temperature, and yielded a Vickers pyramid hardness number (V.P.N.) of only 450. (It may be remarked in passing that the authors have on more than one occasion obtained relatively low hardness figures—which could not readily be explained on the basis of retained austenite—during low-temperature tests. For these abnormal results no explanation can in all cases be offered.) At 225° C. the reaction velocity of steel K was still very high, but as the temperature of the quenching bath was increased it became progressively slower, until, quite suddenly at 360° C., the austenite became exceedingly stable and showed no measurable change in 24 hr. The quenching temperature employed in these tests was 915° C. It will be seen that there is a difference of some 150° C. between these results and those of

Davenport and Bain. Possible causes of this great difference may lie in some element of the composition, the larger size of the specimens with which the authors were concerned or differences in the temperature before quenching.

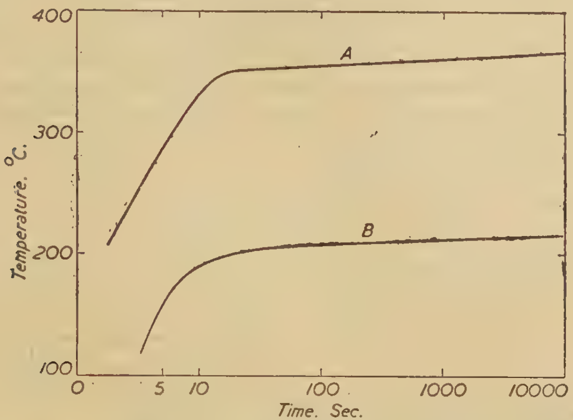

FIG. 6.—Start of Austenite Transformation for (*A*) steel *K*, (*B*) similar steel, according to Davenport and Bain.[6]

The Shape of the S-Curve.

The velocity of crystallisation—or of recrystallisation—is well known, mainly, perhaps, as a result of the work of Tammann,[11] to follow a curve similar to that in Fig. 7. At the transformation temperature crystallisa-

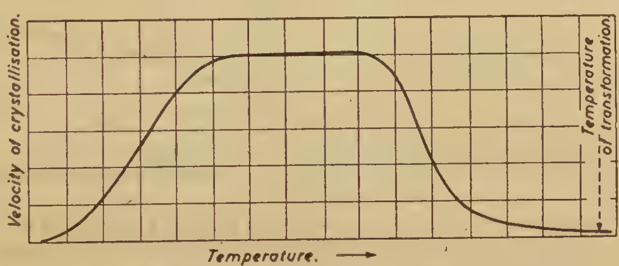

FIG. 7.—Velocity of Crystallisation as a Function of Temperature (Tammann[11]).

tion may be delayed indefinitely, but as the temperature falls the velocity, as a result of the under-cooling, increases more or less rapidly. At some still lower temperature it falls off, owing to the decreased mobility of the system, and at last again becomes negligibly slow. Carpenter, for instance, has shown the persistence of martensite in quenched steels many thousands of years old.

When austenite of eutectoid composition decomposes, it does so in two quite distinct stages: The first of these results in the precipitation of carbide together with the change from γ- to α-iron, yielding a conglomerate pearlite, sorbite or troostite; the other, lower-temperature, transformation to martensite is, in the authors' opinion, concerned with the

allotropic change alone. They are emphatically of the belief that no eutectoid plain carbon steel ever shows more than the two points Ar′ and Ar″. The third transformation, observed, for instance, by Allen, Pfeil and Griffiths,(12) is, in their view, the result of a change associated with the precipitation of some other constituent, probably a carbide. An explanation of the bainite structure, as will be shown later, can be found on the lines of the undercooling theory propounded by Hallimond,(13) the ferrite being precipitated at the boundary of the labile state.

Attention will therefore be confined to the two points Ar′ and Ar″ alone. In Fig. 8 is shown the inverse of the rates of transformation at

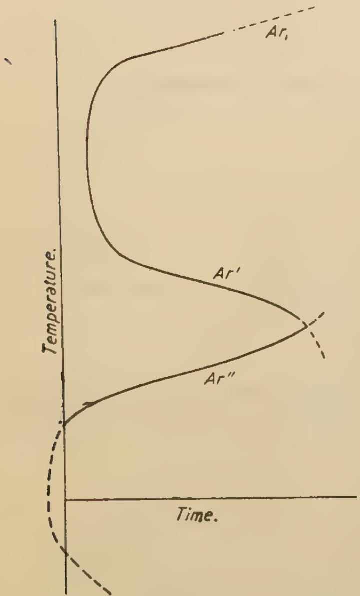

Fig. 8.—Origin of S-Curve for Eutectoid Steel.

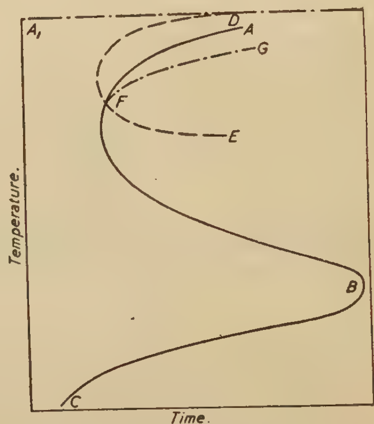

Fig. 9.—Origin of S-Curve in Hypo- and Hyper-eutectoid Steel.

different temperatures, namely, the times required for these to attain a given magnitude. The curves for the Ar_1, Ar′ and Ar″ changes, respectively, when superposed, yield a graph which is exactly of the shape of the normal S-curve.

In the case of steels other than eutectoid a further complication is introduced by the precipitation of a pro-eutectoid phase. From the evidence available it is clear that this is, mainly at any rate, thrown out of solution before the austenite begins to break down, though in some cases it would be desirable to have clearer evidence that the precipitation *does* occur at the bath temperature and not previously during quenching.

If, say, ferrite first comes out of solution, a third curve *EFD* must be superposed on the two, *AFB* and *BC*, shown in Fig. 9. This formation of ferrite will in its turn increase the carbon concentration of the residual austenite and require a shift of the curve *AF* to *FG*.

The effect of ferrite on the structure does not, however, end here. The possibility of a precipitation of ferrite on some sort of supersolubility

line has already been mentioned, and there is abundant evidence of this, especially in the case of some of the S-curves for the more sluggish alloy steels. The effect is most clearly shown in Fig. 34 of the paper by Davenport,[9] in describing which the following statement is made : "At about 1000° F. (540° C.) the rapid-etching product is quite acicular and, *strangely enough, there is an unexpected increase in the volume of the pro-eutectoid ferrite* (*if such it be*), *the distribution of which is distinctly acicular* . . ." These remarks are in complete accord with the idea of a labile shower (of ferrite), the crystallographic characteristic of which normally is long markedly angular crystallites.

That the material which comes out of solution is ferrite is evidenced by its etching properties, by X-ray examination, and by the fact that its appearance is often marked by a change in the hardness/treatment-temperature curve. In some instances the hardness actually falls in this range as the bath temperature is reduced.

It does, however, differ from normal ferrite in the increased amount of carbon retained in solid solution. This is not surprising in view of the diminished rate of diffusion at these relatively low temperatures, and is in accord with Hanemann's observations on the carbon distribution in acicular sorbite. On more prolonged soaking in the bath some of this carbon is rejected.

The "*X*" constituent of Davenport, then, reduces to this labile shower of acicular crystallites of ferrite of higher than normal carbon content.

In the lower range of Ar′ temperatures the austenite itself breaks down to an acicular form of troostite, often termed "bainite." The point of metallographic interest here is the obvious trend towards a martensitic type of structure, even at temperatures well above those at which that constituent can be produced. There is, however, nothing very remarkable about this, the tendency of one solid phase to precipitate along the cleavage planes of a second being a well-known metallographic phenomenon, which tends to occur more and more readily as the temperature falls.

Although, then, the carbide is not (entirely) retained in solution in these higher-temperature constituents, as it is, the authors believe, in martensite itself, the cause, whatever it may be, of the acicular pattern of the martensite is beginning to make itself evident at temperatures well above the Ar″ point, increasing in intensity until martensite itself is retained.

With a progressively lower temperature and, therefore, a still lower rate of diffusion of carbon through the iron lattice, the tendency for carbon to be retained in solution in the ferrite needles increases. The etching properties are, therefore, modified, the needles becoming progressively darker, and the change from this acicular troostite, through "troosto-martensite," to martensite itself is progressive.

The Peak at about 500–600° *C.*

The only other feature of the S-curve with which we need concern ourselves is the extra peak, around 500–600° C., which is found in certain of the alloy steels.

In the case of plain carbon material, low-manganese (up to a little over 1%) or the nickel and cobalt steels, this peak has not been found, and it is of interest that in all these cases the carbide (in annealed material) has been shown by Arnold and Read[14-17] to be of the same general form, M_3C. Further, there are reasons—the gradual change in the etching properties, for instance—for the belief that the new carbide enters into solid solution with the cementite. We are dealing here, then, with steels in which one carbide only is present.

It has been shown equally clearly by the same workers[16, 18–20] that in the chromium, tungsten, vanadium and molybdenum steels a distinct second carbide is formed, and here a well-marked peak in the S-curve is found. There are good reasons, then, for the view (which other workers have expressed) that the peak and the presence of another carbide are interconnected.

In a paper of the type of the present one, speculation may be permitted, and, since more is known with some certainty about the constitution of the chromium steels than that of any others in this group, attention will be confined to these.

The work of Edwards and Norbury,[21] which is in accord with other results, shows that, at any rate in the annealed state, the chromium steels with a chromium-carbon ratio of less than 4·3 contain both cementite and the double carbide $Fe_3C.Cr_3C_2$. Both their tempering experiments and Monypenny's work[22] suggest most strongly that these carbides exist independently of each other at elevated as well as at room temperatures.

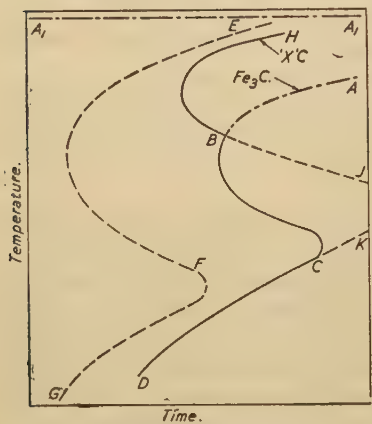

FIG. 10.—Origin of S-Curve where two Separate Carbides co-exist.

It is obviously possible, therefore, to superpose separate S-curves for two such carbides, and this has been done in Fig. 10. The iron-carbide curve, *ABCD*, would not be expected to coincide with that for a plain carbon steel, *EFG*, since there is now chromium present in solution in the austenite, rendering the latter much more stable. The curve will, therefore, be shifted to the right. The chromium-carbide curve is shown at *HBJKD*. The reason for suggesting this order lies in the fact, clearly established by Edwards and Norbury, that the chromium carbide is retained in solution at temperatures below about 500° C., even in the severely stressed martensite, whilst the iron carbide is precipitated at temperatures above about 200° C.

Since the speeds with which martensite is formed from austenite in the low-chromium and in plain carbon steels do not differ very markedly, the lower parts of the curves around Ar″ have not been differentiated.

On the assumption that the austenite starts to decompose when the solubility curve for the specific carbide is reached, the net result will be the curve *HBCD*, in which the peak is clearly shown.

A critical test of the validity of the argument just advanced would be the determination of the S-curve for the eutectoid steel with a chromium-carbon ratio of 4·3, say, carbon 0·6% and chromium 2·6%. If such material still showed the extra peak the latter obviously could hardly be the result of two separate carbides.

Depression of the Ar_1 Point by Pressure.

So far, it has been tacitly assumed that the austenite is in a metastable state and will transform as soon as conditions permit. It has already been shown, however, that there are grounds for the belief that, when quenched, it is in a condition of internal stress, a stress which, above the

Ar″ transformation temperature, may reach, but cannot well exceed, the elastic limit of the austenite at the prevailing temperature, *i.e.*, a few tons per square inch. Since the breakdown is accompanied by an expansion this stress will lower the temperature of the change, and there is, therefore, the *possibility* of this being depressed below that of the bath, when the austenite would be stable—so long as the stress persists.

This depression can be calculated from the well-known Clapeyron equation :

$$\frac{dp}{dT} = \frac{\lambda}{T(v_2 - v_1)},$$

where λ is the heat of reaction per gramme, v_1 and v_2 are the specific volumes before and after the reaction, and T is the absolute temperature. Although the data are not known with any high degree of accuracy, it is quite clear that the depression can be only of the order of, at the most, a very few degrees centigrade per ton of pressure, a figure consistent with Roberts-Austen's experimental results,[23] and may, therefore, be neglected.

Period of Induction.

With certain steels, and over certain ranges of temperature at any rate, it is well known that a period of time, which may be considerable, is required before any observable change in the austenite occurs. This "period of induction" is of both theoretical and industrial importance and justifies closer consideration than has yet been given to it.

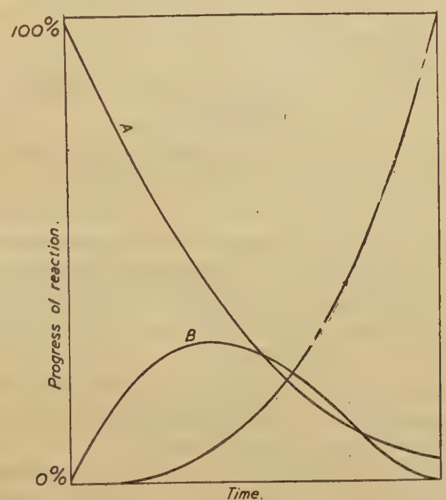

FIG. 11.—Variation of the Quantity of the Product in a Reaction $A \rightarrow B \rightarrow C$.

It may be well to start with a general account of some of the known causes for the existence of an incubation period prior to reaction.

In the first place, there is the case where the reaction is really a successive one, A forming C through an intermediate product B. In such circumstances the graph of the quantities of each of the constituents as a function of time will be of the type shown in Fig. 11. It is conceivable that one may be dealing with this sort of change in the breakdown of

austenite if the carbon is originally in elementary solution but must combine with iron to form the carbide before precipitation.

There is next the possibility of auto-catalysis, in which, for instance, $A \longrightarrow B + C$, where C is the catalyst. It is not difficult to interpret the results of Portevin and Garvin [4] on the acceleration, due to the existence of a second phase, of the formation of troostite from austenite in a somewhat similar manner.

Then comes the effect of an inhibitor, which must first be eliminated before the main reaction can proceed. In steels it is known that the Ar′ change is slowed down very materially by the presence of a special alloy, *e.g.*, chromium. If the carbon (in solution as such) first reacts with the chromium to form an insoluble carbide, the concentration of chromium still in solution in the austenite is diminished (0·1% of carbon should remove nearly 0·5% of chromium) and the speed of reaction increases.

In addition, cases are known in which an effect somewhat of the nature of friction has to be overcome. Just as in the case of friction itself, once movement has commenced the force rapidly decreases, and the reaction then proceeds at an enormously accelerated rate. Some such cases may be those in which a minimum "activation energy" must be induced.

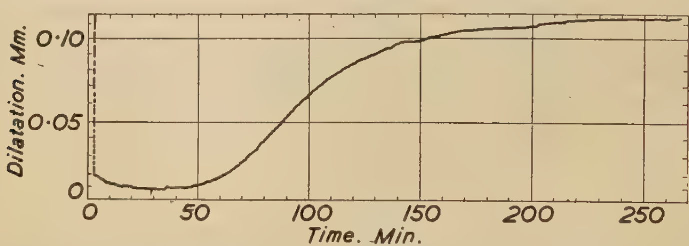

FIG. 12.—Dilatation/Time Curves for a 0·3-in. dia. Specimen of Steel L, quenched from 850° C. in a bath at 210° C. and completely transformed. Traced from an actual autographic dilatometric curve.

Such a state of affairs may be compared with the production of mechanical twins, where the atoms remain in their original position until a certain stress is reached, when they swing over into the twinned orientation almost instantaneously. Since there are certain analogies between twinning and the formation of martensite,[24, 25] we have here a case in which this "frictional" effect may be operative.

In connection with austempering and isothermal transformations, two matters in particular seem to demand special attention, namely, the reality of the supposed effect and the changes which must be presumed to be going on in the material during this time.

With regard to the former, the first point to be definitely established is whether the change is sudden once the conditions for its initiation are present, or whether it may be going on all the time, but initially at a rate so slow as to be unobservable. The high sensitivity of the apparatus with which the authors have worked seems to render it very suitable for deciding this issue. Steel L was chosen for these tests, since it had given a suitable induction period of about 1 hr. at a temperature of 200° C.

The curve in Fig. 12 refers to steel L quenched from 850° C. in a bath at 210° C. in which it was left for 270 min.; the change is at first exceedingly slow, increasing in velocity as time goes on. This and many other similar graphs show clearly that there is no sudden commencement of the change, and are consistent with the view that it may be progressing,

though at first at an almost negligible rate, during the whole time that the material is in the quenching medium.

The clue to the explanation for which we are seeking was obtained by quenching six other specimens of the same steel (L) under identical conditions, with the exception that they were withdrawn from the bath after periods of 5, 10, 20, 30, 40 and 50 min., respectively—that is, before any definite transformation was indicated by the dilatometer. They were then quenched in water to retain as far as possible that condition to which they had attained. In every case, just under the skin there was a layer of martensite needles (Fig. 28), which was completely different from the almost structureless interior. The "case" clearly represented the material which had transformed in the bath, the remainder that which went over to martensite during the subsequent water-quenching. Further, the thickness of the case was constant, showing at once that it had been formed in the earliest stages of the quenching and thereafter ceased to increase during the induction interval. After longer periods than were employed in this series of treatments, when the general transformation had commenced, the newly formed martensitic needles were more or less uniformly disposed over the whole cross-section.

We may here recall the classical work of Benedicks [26] on the conditions which permit of the retention of austenite in quenched plain carbon steels. These researches showed beyond doubt that for such retention the existence of compressive stresses is essential.

In the treatments just described, at the surface where normal compressional stresses were absent the austenite at once transforms. As a result expansion takes place and a compression is exerted on the interior. When this stress rises to a sufficiently high value any further change to martensite is inhibited, a metastable state of affairs being set up and the structure temporarily fixed. At all temperatures, however, relaxation of these stresses can take place, with the result that further breakdown of the austenite is, sooner or later, rendered possible. Although nucleation, an idea invoked by Bain [27] in explanation of the induction period, is not ruled out, and is in fact supported to some extent by other evidence, the authors are of the opinion that the imposition and subsequent relaxation of compressional stresses are the real factors which are responsible for the initial reluctance to transform and then the subsequent breakdown. Nucleation may well be secondary to this, but if such nucleation can occur at all it is, in the authors' view, the direct result of stress relaxation. From this argument it follows that in a given steel the induction period is determined by the net effect of the tendency to change at a given temperature and the rate at which, at the same temperature, stress relaxation can take place.

In this connection, it is an interesting and not unimportant fact that these curves for the commencement and completion of the transformations are in many cases remarkably similar to each other. Now, the former curve represents the duration of the incubation period, the latter the time necessary, once the induction stage is over, for the complete breakdown of the austenite, *i.e.*, the average rate of the reaction.

There is no obvious reason why these two quite dissimilar effects should show any parallelism, and one is led to speculate on the probable explanation being found in the fact that even the first curve, in reality, merely represents the time required for the change to become large enough to be measurable. If this be the case, where one curve is a measure of the time required for the attainment of a fairly definite minimum amount of change, the latter being that necessary for its (essential) completion, the velocity of the reaction is equally responsible for both curves and their similarity is more understandable. This presupposes that even during

the induction period the transformation is already going on, but at an immeasurably slow rate.

This conception of the importance of stress relaxation upon the rate of transformation of austenite to martensite is of interest in still another connection, namely, the shape of the curve expressing the completion of the change in the low-temperature range, regarding which there is still some uncertainty. In some diagrams this curve is shown still bending downwards to the left, whilst in others it moves still further towards the right.

Much, if the authors are correct, must clearly depend upon the shape and dimensions of the actual specimen used. The smaller this sample is, the more readily would one expect the stresses to be relieved. No curve can do more then than express the results for those particular experimental conditions adopted, though, in general, there will be a more and more marked trend towards the left, *i.e.*, to higher rates of reaction, as the surface/mass ratio becomes greater.

The General Shape of the Transformation Curve.

Several attempts have been made to obtain equations which will fit the experimental data obtained on the transformation of austenite to martensite, but without overmuch success. As an example may be cited that of Austin and Rickett,[28] who suggested:

$$\log \frac{P}{100 - P} = K \log t + \frac{a}{T} + b,$$

where P is the percentage of austenite changed in time t at an absolute temperature T, while K, a and b are constants. There does not appear to the authors to be any fundamental justification for this expression, which certainly does not coincide with all the known facts.

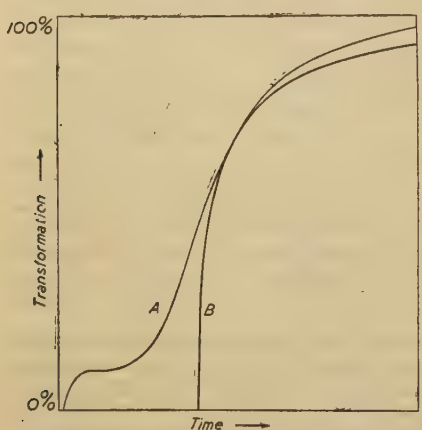

FIG. 13.—*Curve A.*—Dilatation/Time Curve for the Austenite Transformation of Steel *A*, quenched from 810° C. in a bath at 200° C. *Curve B.*—First-Order Reaction Curve: $kt = \log_e a/(a - x)$.

In Fig. 13, *A*, is shown, in its most general form, the dilatometric curve of the breakdown of austenite to martensite in the case of the eutectoid plain carbon steel, *A*, quenched from 810° C. in a bath at 200° C. Starting with a small, but initially rapid expansion which quickly slows down, a stage is reached in which further transformation temporarily ceases. This is in turn succeeded by a second period of expansion which becomes increasingly rapid, attains a maximum velocity, and then diminishes once more.

The primary expansion can be correlated with the formation of the martensitic "case," which the authors have invariably found on the surface of carbon steels, and which has already been discussed. The succeeding period of temporary stability is essentially identical with the

period of induction which we have just considered, and, as the stresses are relaxed, breakdown of the austenite once more commences, slowly at first, and then progressively faster and faster.

Theoretically, there are reasons for the belief that the formation of "nuclei" of the transformation product should lead to an equation of the general form :

$$\frac{dV}{dt} \propto V^{\frac{2}{3}},$$

where V is the fractional volume of the product. This relationship is plotted on the left in Fig. 14 and, as will be seen, can be fitted very satisfactorily on to the early part of the experimental curve. In passing, it may be remarked that such nuclei need not be, and probably are not, of any very definite nature; they merely represent regions in the austenite lattice where suitable atomic configurations have momentarily occurred. As will be seen from Fig. 14,* however, this type of internal change would

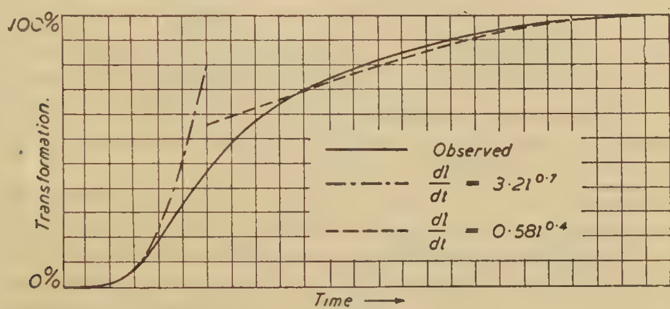

FIG. 14.—Equations representing the Earliest and Latest Portions of the Typical Dilatation Curve.

result in an extremely rapid acceleration of the rate of transformation which would very soon be more or less complete. Nucleation, whatever this may really mean in this connection, cannot, therefore, be more than one factor, and possibly a quite subordinate one, in determining the rate of change in the earliest stages. Further, other types of curve could be fitted almost equally well to the initial part of Fig. 14, and the authors are therefore disposed to agree with the view of Mehl,(29) that the formation of martensite is not one of nucleation and growth.

Passing on to consider the main portion of the curve, Bain (27) has already shown that the curve does, over some central portion of its course, agree well with that required for a first-order reaction. This is also shown in the authors' work, Fig, 13,*B*. In such a reaction, the velocity is proportional to the concentration of the reactant still unchanged :

$$kt = \log_e \frac{a}{a - x},$$

where a is the initial concentration and x the amount transformed in time t. Such a curve involves a maximum velocity at the moment at which it starts, and it is, perhaps, not without importance in this connection that the initial stage of breakdown, in which the first skin of martensite is produced before the range of metastability is reached, is of this nature. If the induction period is imagined to be eliminated, and the

* Fig. 13 with omission of first stage of decomposition at skin.

earliest portion of the curve fitted on to the main one, the similarity between a first-order curve and the one experimentally obtained becomes still more marked.

There are, however, several points to which attention must be directed. In the first place, it is only over a relatively small portion of the whole transformation that the experimental curve can be fitted to the first-order curve, and the question inevitably arises as to whether this is not merely fortuitous. In this connection, however, the remarks just made regarding the first part of the entire curve, if of any cogency, tend to increase the probability of the curve being of this physico-chemical nature. Secondly, the same theoretical curve is given by reactions of a higher order—the so-called pseudo-monomolecular reactions—where it is the concentration of one substance alone which suffers any appreciable change. If, for instance, it were the case that the carbon in austenite were in elementary solution, whilst in martensite it was present as Fe_3C, then, since the concentration of the iron is to all intents and purposes unchanged, the same type of curve would be obtained. Further, a reaction involving diffusion, a supposition which need hardly be defended in such a change as that which we are considering, would proceed at a speed which, in the main, is determined by the rate at which this diffusion takes place. It is then very far from certain, despite the excellence with which the experimental curve and that of a first-order reaction coincide over a limited range, that the change is one which the physical chemist could properly regard as being of the monomolecular type.

The final part of the curve which now demands attention, as both Bain's work and Fig. 13 show, indicates a speed of transformation which is appreciably greater than that which the first-order curve would require. The evidence available supports the view that this part of the curve does not become asymptotic to the ordinate representing the complete change, but cuts the latter after the expiration of a definite time. Explanations of this acceleration are possible from more than one point of view. There is, in the first place, the possibility of some auto-catalytic effect, just conceivably the result of actual deposition of carbide from the super-saturated solution. In this connection the extreme rapidity of the austenite-troostite change as compared with the austenite-martensite inversion may be called to mind. Further, there is the obvious possibility of " surface " effects exerting some influence. When the first nuclei are formed (if we assume that nucleation does occur) they will be in the nature of more or less rounded inclusions in the austenite lattice. The growth of these into the matrix in which they are embedded can take place only at the convex surfaces, and such a state of affairs tends, particularly in the earliest stages, to retard the change. At the end of the transformation, however, the conditions are reversed, the residual austenite being enclosed in the products of the transformation, and further disappearance of the austenite is accelerated.

The whole process is, however, subject to still another complicating factor, which has not yet, so far as the authors are aware, been taken into account by any earlier worker, namely, the possibility that tetragonal martensite is first formed with an expansion, a secondary contraction occurring when this passes over into the cubic form. Whether the tetragonal variety is invariably produced first and then changes to the cubic type is as yet by no means settled, and some observations on this point may not be without their interest.

Tempering Experiments.

It has been known for a long time that a water-quenched, medium- or high-carbon steel shows a quite marked abnormality when it is tempered at a temperature in the neighbourhood of 100° C., and α and β

forms of martensite have, therefore, been postulated. Excellent examples of this change are provided, for instance, by the work of Andrew, Rippon, Miller and Wragg,[30] Enlund,[31] Träger[32] and others. At such a low temperature only the martensite can be responsible for the effect, since it is clear from this and other work [31, 33] that austenite, if it be retained, remains relatively stable until a distinctly higher temperature is reached.

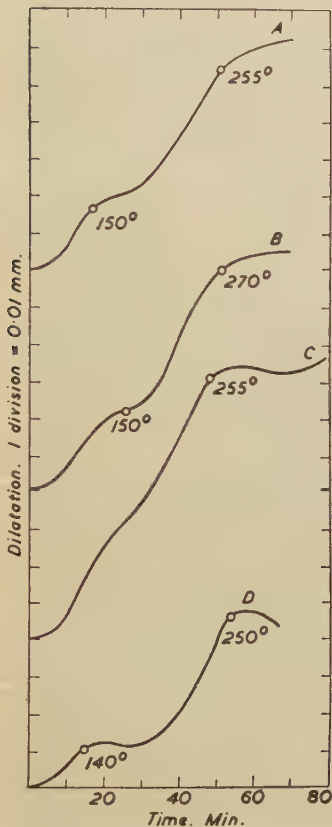

FIG. 15.—Dilatometric Tempering Curves for Steel D (curves A and B) and Steel A (curves C and D).

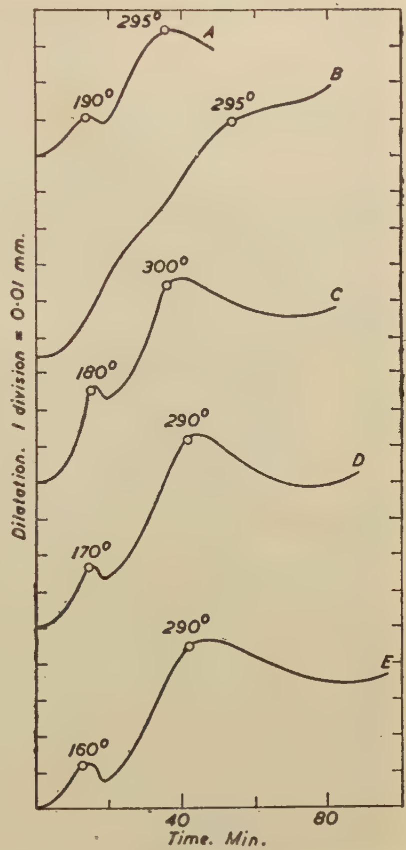

FIG. 16.—Dilatometric Tempering Curves for Steel L.

The real nature of this transformation was first elucidated by Honda and Nishiyama,[34] from whose work it is clear that martensite can exist in two crystallographic forms, (1) a tetragonal form, which on mild tempering passes over to (2) a cubic form. It is equally clear from the work already cited, as well as from the authors' results, that the tetragonal-cubic inversion is accompanied by a small, but quite definite contraction.

In addition to this change, however, two others may occur; first, the tempering of residual austenite, resulting in a marked expansion, and,

secondly, a contraction, resulting from the precipitation of carbide from solution. For the purpose of studying the effect of tempering, the sample was suitably quenched, placed in the dilatometer and slowly reheated at a rate of about 4° C. per min.

In Fig. 15, *D*, is shown the tempering curve for the eutectoid plain carbon steel *A*, which was water-quenched from 810° C., the initial hardness being 830 V.P.N. The curve shows two marked contractions, one around 140° C., the other commencing at about 250° C. It does not seem unreasonable to associate these points with the tetragonal-cubic inversion of the martensite and the tempering of the cubic martensite with the formation of a troostitic constituent, respectively. There is no sign of any sudden marked expansion, from which it may be concluded that no considerable amount of residual austenite is present. At the same time it must be remembered that this constituent may be present, but as a result of its small concentration or slow decomposition may cause merely an imperceptible change in the slope of the curve.

Fig. 15, *C*, shows for comparison the tempering curve of the same steel quenched from 900° C. in a bath at 200° C., in which it remained for 110 min. Two main points of difference at once present themselves. First, the definite contraction at 140° C. undergone by the water-quenched specimen has become a mere hump, suggesting most strongly that in the treated specimen the martensite is, for the most part, already in the cubic form; secondly, the total expansion is very much greater, a fact clearly consistent with the view that there is still a considerable amount of unchanged austenite; the expansion from the latter on transformation extends over a quite wide temperature range, even exerting its influence after the contraction due to actual precipitation of carbide has set in.

In Fig. 15, curves *A* and *B* show the results for the nickel-chromium steel *D* in the water- and oil-quenched states. The expansion is rather greater for the oil-quenched condition, a result consistent with the view, fairly generally held, that oil-quenching is more likely to lead to the retention of γ-iron than is a corresponding water-treatment.

In Fig. 16 curves *A* to *E* show the results of similar work carried out on the hyper-eutectoid steel *L*. So far as the water-quenched sample *A* is concerned the only noteworthy point is the relatively high temperature—190° C.—at which the inversion of the martensite occurs. The four curves *B* to *E* were obtained on samples which were quenched from 850° C. in a bath at 210° C. and allowed to remain there for varying periods of time before removal. From Fig. 12 it will be seen that noticeable transformation did not commence until about 50 min. after quenching. The samples to which curves *C*, *D* and *E* of Fig. 16 refer were all removed from the quenching bath before any measurable change took place, while in the case of curve *B* the specimen had remained in the bath for some 300 min., after which, as the dilatometric record indicates, the austenite had almost completely broken down.

Curves *C*, *D* and *E* are very similar to each other, the only noteworthy difference being the progressive change of temperature at which the tetragonal-cubic change of the martensite occurs. This is not inconsistent with the view that there is also a progressive change of internal stress. The fully transformed steel, on the other hand, shows practically no sign of the martensitic inversion, suggesting very strongly indeed either that the austenite transformed in the bath directly to cubic martensite or, alternatively, first to the tetragonal form and then almost instantaneously to the cubic. It is unfortunate that there is no very direct means of differentiating between these two possibilities. It is quite clear, however, that in the samples removed from the bath while still mainly austenitic, the martensite which formed on cooling is, largely at any rate, of the tetra-

PLATE XIII.

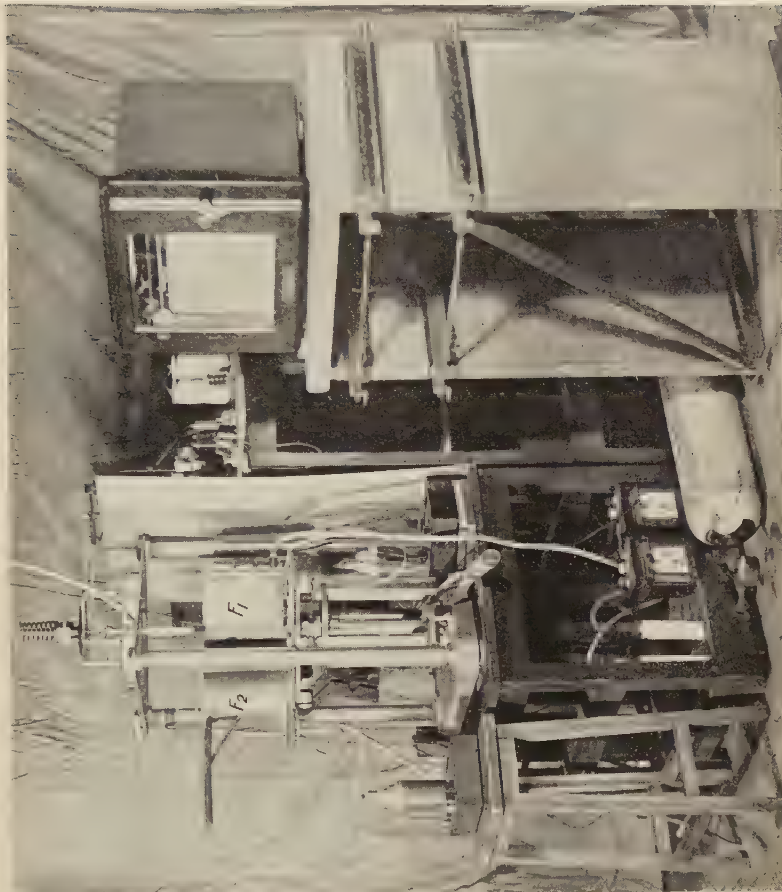

FIG. 27.—The Complete Apparatus.

[*Thompson and Stanton.*

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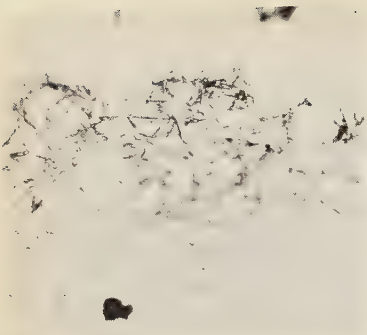

FIG. 28.—Steel L, quenched from 850° C. in an isothermal bath at 210° C., water-cooled after 20 min. (before start of austenite transformation). Nickel casing alongside martensite layer. Etched 10 sec. × 250.

FIG. 29.—Steel L, quenched from 810° C. in a bath at 210° C., water-cooled before transformation started. Nickel casing alongside martensite layer. Etched 5 sec. × 250.

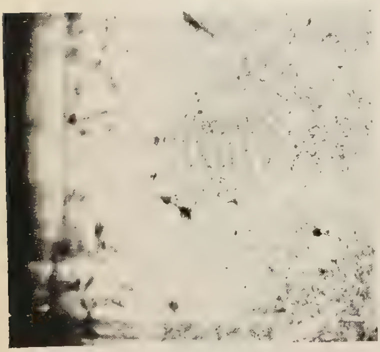

FIG. 30.—Same Specimen as in Fig. 29. Etched 10 sec. × 250.

FIG. 31.—Steel L, quenched from 810° C. in a bath at 210° C., allowed to transform partially before water-cooling. Etched 5 sec. × 250.

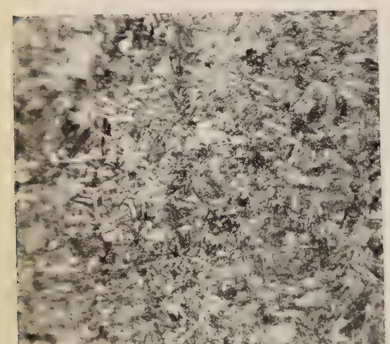

FIG. 32.—Completely transformed at 210° C.

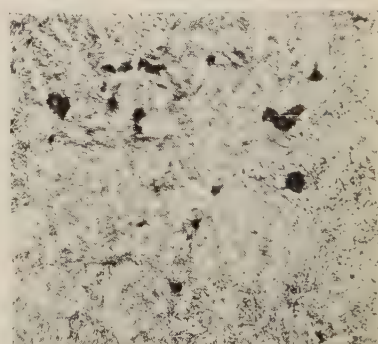

FIG. 33.—Water-cooled before transformation started, T. 150° C.

FIGS. 32 and 33.—Steel L, quenched from 850° C. in a bath at 210° C. Etched 5 sec. × 250.
(Micrographs reduced to four-fifths linear in reproduction.)

[*Thompson and Stanton.*

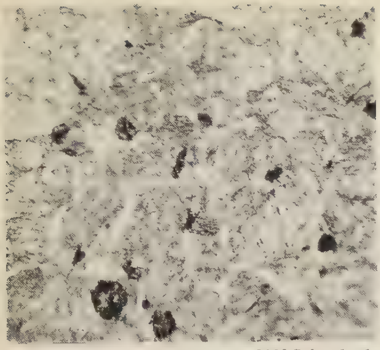

Fig. 34.—Steel L, quenched from 810° C. in a bath at 210° C., water-cooled before transformation started. T. 250° C. Etched 5 sec. × 250.

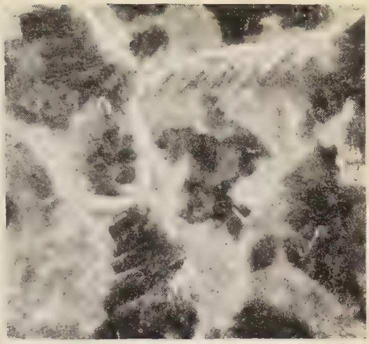

Fig. 35.—Large Specimen of Steel A, 2 in. × 0·45 in., quenched from 830° C. in a bath at 150° C. Etched 5 sec. × 350.

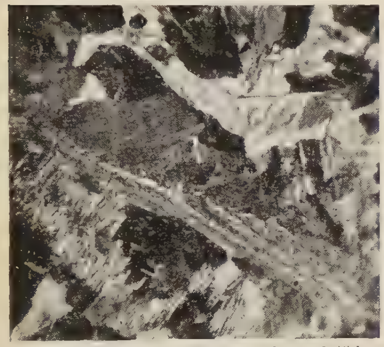

Fig. 36.—Steel A Specimen, 0·875 in. × 0·45 in., quenched from 830° C. in a bath at 150° C. Etched 5 sec. × 350.

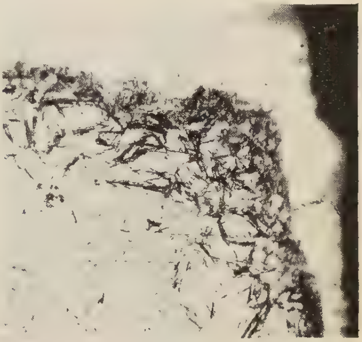

Fig. 37.—Steel L Specimen, 0·875 in. × 0·45 in., quenched from 860° C. in a bath at 200° C., removed from bath after 70 min. (before start of transformation, *see* Fig. 19), water-cooled. Nickel case beyond martensite layer. Etched 10 sec. × 250.

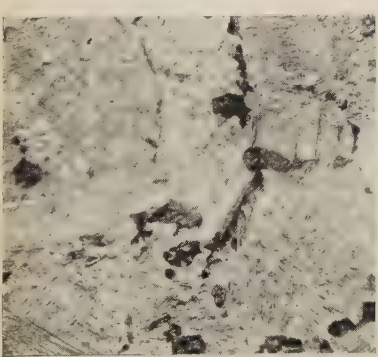

Fig. 38.—Specimen $A28/1$, showing grain-boundary troostite. Etched 5 sec. × 250.

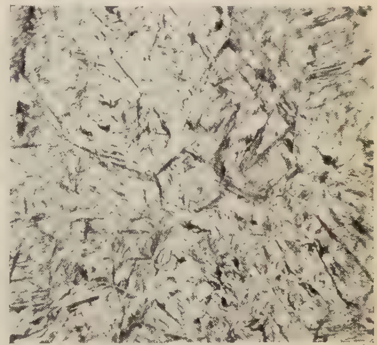

Fig. 39.—Specimen $A27/1$, entirely martensitic. Etched 5 sec. × 250.

(Micrographs reduced to four-fifths linear in reproduction.)

[*Thompson and Stanton.*

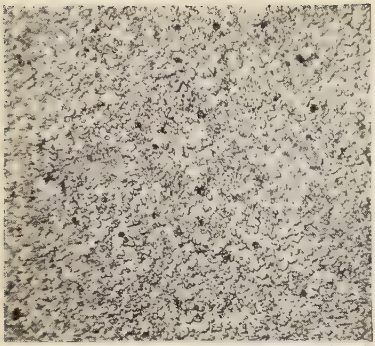

FIG. 40.—Steel A, pretreated to produce spheroidal pearlite. × 300.

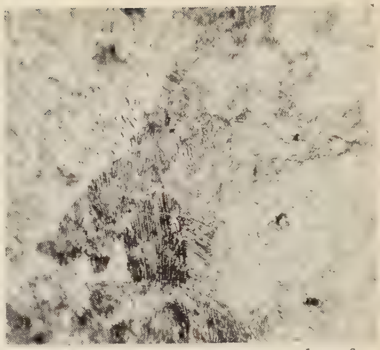

FIG. 41.—Steel A, pretreated to produce fine pearlite. × 900.

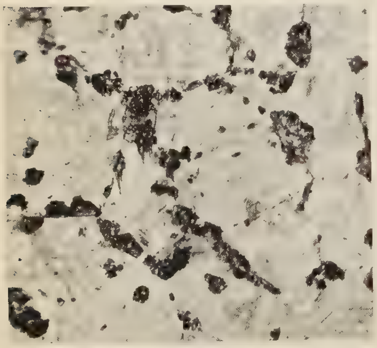

FIG. 42.—Pretreated to give spheroidal pearlite. × 300.

FIG. 43.—Pretreated to give fine pearlite. × 250.

FIGS. 42 and 43.—Steel A, pretreated, after quenching from 900° C. and transforming at 200° C. Etched 5 sec.

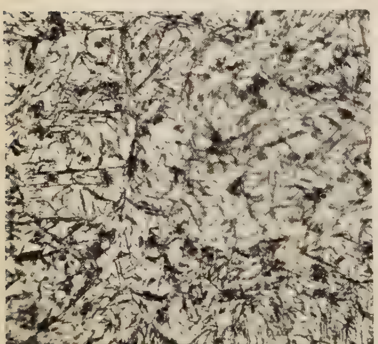

FIG. 44.—Pretreated to give coarser pearlite.

FIG. 45.—Pretreated to give sorbitic structure.

FIGS. 44 and 45.—Steel A, pretreated, after quenching from 900° C. and transforming at 200° C. Etched 5 sec. × 260.

(Micrographs reduced to four-fifths linear in reproduction.)

[*Thompson and Stanton.*

[*To face p.* 153 P.

gonal variety, whilst in the sample allowed to transform at the higher temperature of the bath, the product is essentially cubic. *May one not in this definite fact see a possible explanation of the importance of the process ?*

Certain of the autographic curves for specimens in which the transformation is proceeding slowly tend to progress in a series of steps (*see* Fig. 17). For some considerable time it was uncertain whether these steps were not due to small temperature fluctuations in the bath coinciding with the time intervals at which the thermostatic control came into operation. This explanation is inconsistent, however, with the clearly established fact that these periodic irregularities are not present in all the curves. An excellent example of this is shown in Fig. 12, where the length of a specimen, still austenitic, and, therefore, with a higher coefficient of expansion than normal, remained essentially unaltered for some three-

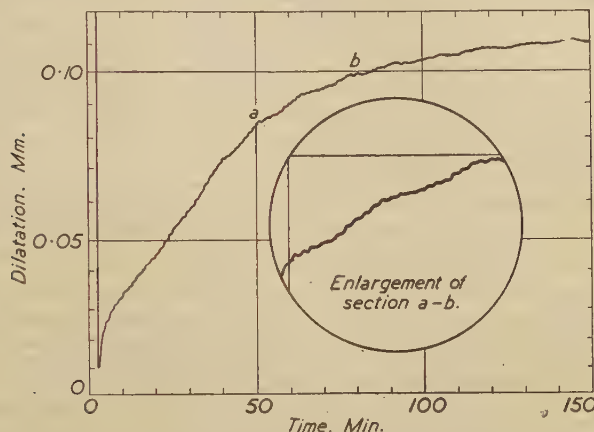

Fig. 17.—Dilatation/Time Curve for 0·3-in. dia. Specimen of Steel *D*, quenched from 830° C. in a bath at 240° C. Traced from an actual autographic dilatometer curve.

quarters of an hour. We may, then, regard this periodic effect, found in only some of the curves, as being an inherent part of the transformation process.

An adequate explanation is provided if one imagines small " parcels " of austenite * as transforming with considerable rapidity into tetragonal martensite with an easily measurable expansion. After a small interval of time, depending on composition and temperature, this tetragonal martensite is tempered to the cubic form, with consequent contraction. On the basis of this hypothesis it would be reasonable to believe that the tetragonal martensite is a frequently formed intermediate product, especially when the rate of change is slow, but, unless it tempers essentially instantaneously, is not formed when the transformation is rapid.

The structures found in these isothermally transformed and tempered materials can, perhaps, be best discussed in connection with the samples of steel *L* quenched from 810° C. which have been allowed to remain in the metal bath at 210° C. for varying periods of time. In all cases in which the sample was removed from the bath before any measurable expansion was noted in the dilatometer, the specimen (apart from the skin), after etching for 5 sec. in 4% nitric acid in alcohol, was essentially structureless (*see* Fig. 29), so much so that at first it was mistaken for

* *Cf.* Wiester's kinematograph film (The Iron and Steel Institute and The Institute of Metals possess a copy).

austenite. That it was martensitic, however, was shown (1) by a more prolonged etching, which developed a faint acicular pattern (Fig. 30), (2) by its hardness, about 700 V.P.N., and (3) by the shape of the dilatometric tempering curve, in which the tetragonal-cubic martensite inversion is clearly shown (Fig. 16, curves *C*, *D* and *E*). This material, then, in the opinion of the authors, is martensite in the tetragonal form.

The martensite in a drastically water-quenched sample treated from the same initial temperature was similar except for the fact that its hardness was appreciably higher—830 V.P.N.

In contrast to these results, the martensite which forms at the bath temperature, and not during the subsequent cooling, is characterised by rapidly etching needles (*see* Fig. 31), which soon become almost black. In proportion as the transformation proceeds at the higher temperature the magnitude of the tetragonal-cubic abnormality in the dilatometer tempering curve diminishes. This is, in all probability, the cubic martensite, which generally occurs in a somewhat coarse acicular form. One metallographic point of difference which must be noted is that, whereas in this latter form the needles are dark on a lighter background, in the tetragonal variety, when the structure has been developed, it is the needles which are the lighter. This (so far as the authors are aware) constant difference is shown in Figs. 30 and 31.

The question at once arises as to whether it is proper to regard these dark-etching needles as martensite at all, or whether they may not really be troostitic.* An answer can be provided by the results from a specimen which transformed isothermally under the same conditions as before, except that it was allowed to remain in the bath for 275 min., when, as the dilatometric curve (Fig. 12) shows, the change was essentially completed. The structure is shown in Fig. 32, the hardness still being 710 V.P.N. This hardness is clearly inconsistent with a structure which is troostitic, and is of the order to be expected from martensite, a fact which is also supported by the dilatometric curve, Fig. 16, *B*. It will be seen that the only strongly marked abnormality is that which starts just below 300° C., which can be associated only with carbide precipitation, *i.e.*, with the formation of troostite or, perhaps more exactly, a troosto-martensite complex, for the change is certainly progressive. The authors are satisfied, therefore, that the needles formed in the bath at 210° C. are cubic martensite.

When samples containing the tetragonal form are tempered there is little change of microstructure so long as the temperature of tempering is maintained below that of the tetragonal-cubic change point (*see* Fig. 33). At temperatures above this, however, there is a progressive darkening of the needles until a structure typical of that due to the higher temperature transformation is obtained (*see* Fig. 34). This change is not instantaneous, however, but, as both the dilatometric and the metallographic results show, extends over a definite range of temperature or time.

In connection with these tempering changes the variations in hardness are not without interest. Samples removed from the quenching bath before any change could be detected gave V.P.N. hardness readings in the neighbourhood of 700. They were then tempered for 10 min. at 190° C., *i.e.*, just above the martensite inversion point. The hardness rose, as a result, to an average of 750 and in one case to 810 V.P.N., the hardest material the authors have yet obtained as a result of such transformations.

On tempering at 345° C., *i.e.*, above the troostite formation temperature, the results fell into two quite sharply divided groups. When

* Throughout this paper the authors have retained the term troostite rather than adopt that of "fine pearlite" or "bainite," as they are not here concerned with the nature of this constituent.

the martensite had formed on air-cooling from the bath the hardness after tempering was constant at 467 $\pm$ 4 V.P.N., whilst the steel allowed to transform *in* the bath, when similarly tempered, had a hardness of 563 V.P.N.—almost a hundred points higher. *This increased resistance of the isothermally transformed material to tempering is just what would be expected from a material already relatively very free from internal strain.*

Part II.—Factors which Affect the Process.

The Size of the Specimen.

Most of the earlier work on austempering was carried out on wire or thin sheet. One of the requirements kept in mind in the design of the apparatus previously described, however, was that it should permit of working with samples of distinctly greater size.

Legge,[35] however, has given an account of some attempts made to determine the maximum sizes of various steels which can be treated to give a Rockwell hardness figure of 50–60 (500–700 V.P.N.). One of these, with 0·8–0·9% of carbon and 0·3–0·5% of manganese (very similar to the authors' steel *A*), gave a Rockwell figure of 55–58 (590–650 V.P.N.) for diameters up to 0·156 in. As will be seen later, the authors have obtained fully martensitic structures with a diameter twice as great. The corresponding diameters of different steels obtained in the American work were as follows :

Steel :	Cr 0·5%.	Mn 1·6–2·0%.	Cr 1%, Mo 0·2%.	Ni–Cr–Mo.
Diameter. In. . .	0·31	0·625	0·5	> 1·0

In the initial experiments the specimens of plain carbon steel employed had a length of 2 in. and a diameter of 0·45 in. The authors found it impossible, however, to quench these materials effectively or uniformly in their apparatus. Even with the nickel-chromium steel *D*, only isolated areas of martensite formed when such specimens were treated at 230° C., while the eutectoid plain carbon steel *A* was almost entirely pearlitic (*see* Fig. 35), even when quenched in a bath at as low a temperature as 150° C.*

The length of the specimen was, therefore, reduced to 0·875 in. for the remainder of the tests. With this length and a diameter of 0·45 in. the critical cooling-rate was more nearly approached, the structure of steel *A* consisting of troostite and martensite (*see* Fig. 36). By further reduction of the diameter to 0·3 in. fully martensitic structures were obtained (even in such a plain carbon specimen as steel *A*). The dilatometric curves for these two diameters are reproduced in Fig. 18, while the hardnesses are given below:

Length. In.	Dia. In.	Bath Temp. ° C.	Time in Bath. Min.	Mean V.P.N.
2·0	0·45	237	75	198
1·5	0·45	220	60	339
0·875	0·45	200	5	348
			30	491
			60	383
			100	367
0·875	0·3	200	120	702

* For this, two factors are responsible; in the first place there is the relatively large mass of hot silica in the immediate vicinity of the steel, while, secondly, the space between the steel and the silica tube was insufficient to permit of an adequately rapid flow of the quenching metal. It would appear, therefore, that where these effects do not operate, the treatment could be satisfactorily carried out on sections somewhat larger than those with which the authors have been successful; especially is this the case when a troostitic or similar structure only is required.

The figures indicate that the hardness of the short test-piece of 0·45-in. dia. (removed from the bath after 5 min.) was roughly 150 less than that of the sample maintained in the bath for half an hour. The authors

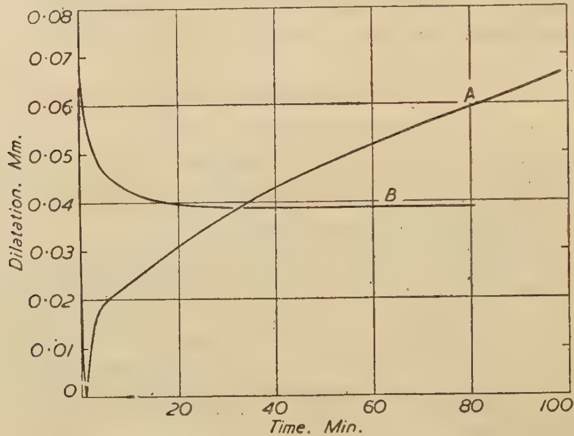

FIG. 18.—Dilatation/Time Curves for Steel *A*. *Curve A*.—0·875 × 0·45-in. specimen. *Curve B*.—0·875 × 0·3-in. specimen, quenched from 830° C. in a bath at 150° C.

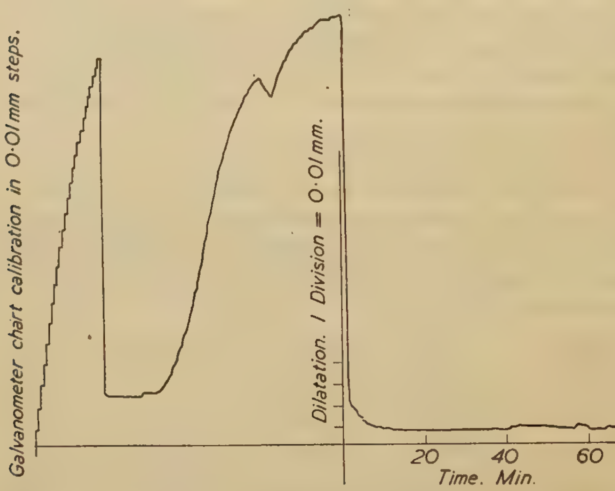

FIG. 19.—Dilatation/Time Curve for 0·875 × 0·45-in. Specimen of Steel *L*, quenched from 860° C. in a bath at 200° C.

have come across many similar anomalous results, the most probable explanation of which is that the rate of cooling is so slow in the interior that even after several minutes the temperature of the specimen is still appreciably higher than that of the metal in which it is immersed. This

is inconsistent with the statement of Greninger and Troiano[36] that ¼-in. cubes attained throughout the temperature of the lead bath in under 5 sec., a statement which the authors find somewhat difficult to accept. This question of the rate of cooling will be considered again later.

Passing on to consider the hyper-eutectoid carbon steel *L*, quenched from 860° C. in a bath at 200° C., a specimen of 0·45-in. dia. underwent no recordable transformation in the bath during 70 min., indicating that it was, even then, fully austenitic (Fig. 19). After removal from the bath the samples were water-quenched, examined microscopically and their hardness values measured. With the exception of the coarse martensitic case the structures were all typical of the martensite formed from austenite on cooling from the bath temperature, as shown in Fig. 37. The V.P.N. numbers for specimens 0·875 in. long but of varying diameters, together with the corresponding results for steel *A*, are given below :

Steel *L* :	Diameter. In. .	0·45	0·35	0·30	0·25
	V.P.N. . . .	687	723	...	717
Steel *A* :	Diameter. In. .	0·45	0·35	0·30	0·25
	V.P.N. . . .	367	...	702	712

In considering these results it should be pointed out that the manganese content of *L* was higher than that of *A*, and manganese is an element which has a marked influence in accentuating the effects of isothermal transformation, a result which is entirely consistent with its known influence on normal quenching operations.

A still more cogent reason for the difference, however, which is at first sight at variance with the conclusions of Portevin and Garvin (who found a minimum critical cooling velocity was required for the eutectoid steel), is found in the increased carbon concentration of the austenite. The quenching temperature of 860° C. was sufficiently high to allow of the carbide being dissolved, and all the evidence is in agreement in showing that, other things being the same, the stability of the austenite increases with the carbon concentration in solution. It is here that the conditions differ from those of Portevin and Garvin, who quenched their material from a temperature only slightly exceeding the Ar_1 point, *i.e.*, when undissolved cementite still remained to act as nuclei.

Specimens of 0·45-in. dia. of the nickel-chromium steel *D*, after quenching from 810° C. in a bath at 245° C., even after 90 min. consisted throughout of a fine martensite produced from austenite on the subsequent cooling. The hardness after this treatment was practically identical at 525, whether the diameter was 0·45 in. or 0·3 in., and it is clear that this material could be satisfactorily treated in sizes distinctly greater than those with which it has been possible to deal.

The Grain Size.

The grain size of a steel may be determined by two quite distinct causes. In the first place there is a grain refinement (or coarsening) due to changes of composition, and in this category must be placed steels, such as those of controlled grain size, which owe their fine grain to additions which may not be *directly* responsible. (Steels with traces of boron may also belong to this class.) Secondly there are the alterations of grain size due to heat treatment. It does not necessarily follow that these two different reasons for variations of grain size will have the same effects on the austempering properties of the material.

It is known that chemical changes of the steel modify, often appreciably, the behaviour of the austenite when it decomposes at temperatures below the carbon change-point. The evidence for the effect of grain size varia-

tions brought about by heat treatment is, however, by no means so clear. It must not be forgotten that the grain size of a metal is, so far as the authors are aware, invariably reduced so long as diffusion is taking place within it, *i.e.*, when quickly cooled from a temperature just below A_1 up to that at which solution of the ferrite or carbide is complete. Further, so long as this is taking place the crystalline structure is rendered less perfect, with the result that its breakdown is appreciably modified. This is excellently shown, for instance, in the cases of cemented bar or a black-heart malleable cast iron. In both cases, as a result of the long-continued heating at the high temperature, an exceedingly angular pattern would be expected. Actually, this occurs in neither, owing to the orientation of the crystals being interfered with by the diffusion which is going on, and the well-formed cleavage planes necessary for the angular precipitation of the pro-eutectoid constituent being unavailable. During the solution of ferrite or carbide on heating, this effect must be operative, with the result that the austenite is not crystallographically identical with that formed when equilibrium of composition obtains. It is not unreasonable to expect that when the steel is austempered the stability of the austenite will be affected by its structure, and effects put down to grain size variations may well be, in part at any rate, due to this crystallographic variation.

There is still another possible factor to be taken into consideration. We know practically nothing regarding the actual molecular weight of even iron carbide, much less of those carbides present in the alloy steels. Several workers have suggested that there may be association, the degree of which will presumably diminish as the temperature is raised. Further, it has been suggested (with some degree of probability) that double carbides of iron and an alloying element may dissociate at high temperatures. In either case the molecular concentration of the austenite would be increased and its stability thereby enhanced.

The fact that austenite breaks down in two quite distinct ways also suggests that the influence of grain size may be different when an aggregate is formed from that where the product is martensite. Considering the former, it is well known that this does tend to come out on the grain boundaries (*see* Fig. 38), and would, therefore, be expected to be accelerated by an increase in such surfaces. This is in accordance with the evidence at present available. The formation of martensite, on the other hand, is not associated with the inter-crystalline boundaries of the austenite (Fig. 38), and should not, therefore, be appreciably dependent on the size of the crystals from which it forms.

On the most general grounds, therefore, one would expect small grain size to facilitate the formation of a troostitic constituent, but to have little, if any, effect on the production of martensite.

The Time at the Quenching Temperature.

As the time of soaking at the quenching temperature is increased two effects may be expected, first, an increased uniformity of carbon content in solution, and secondly, an increased grain size.

Roberts and Mehl [10] have shown very clearly that, even after 10 min. at 800° C., the carbon in a eutectoid plain carbon steel is still in a state of distribution resembling that of the original pearlite. Since it appears that the stability of the austenite at lower temperatures increases as the carbon content is increased, the elimination of this " ghost " pattern on more prolonged soaking might be expected to affect the austempering characteristics of the steel, probably in the direction of a decrease in the velocity of the austenite breakdown.

The effect of the time of soaking at the quenching temperature on the

subsequent transformation was examined for steel *D*, the relative sluggishness of which it was expected would render it possible to obtain a clear picture of any changes which might be induced. The samples were heated to 810° C. for varying periods and then quenched in baths at 190° C. and 280° C., respectively. Typical dilatometric curves are shown in Fig. 20,

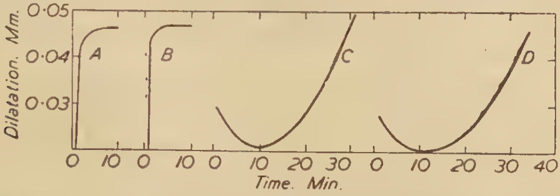

FIG. 20.—Dilatation/Time Curves for 0·3 × 0·875-in. Specimens of Steel *D*; curves *A* and *B* transformed at 190° C., *C* and *D* at 280° C. Soaking time at quenching temperature, 810° C., curves *A* and *C*, 10 min., *B* 90 min., *C* 30 min.

and the change of length during the first 10 min. in the bath are recorded in Table II., from which it will be seen that in neither case has any appreciable variation resulted from the increased time at the high temperature.

TABLE II.—*Effect of Time at Quenching Temperature on Steel* D.

	Bath Temp.				
	190° C.			280° C.	
Time at quenching temp. Min.	10	30	90	10	30
Vickers pyramid hardness number	620	625	580	697	618
δl in first 10 min. Mm. .	+0·053	+0·058	+0·054	−0·0162	−0·015

No change of microstructure could be detected within either of the series of quenchings.

The Quenching Temperature.

The effect of variations of quenching temperature on the isothermal transformation of the alloy steels is far too large a subject to be dealt with here; attention will be confined, therefore, to the eutectoid plain carbon steel *A*. In this steel grain-size effects alone may be expected to be of importance.

Esser and Majert [37] state that an increase in the initial quenching temperature tends to reduce the critical cooling velocity, and they ascribe this to the effect of the increased size of the austenite grains.

Two specimens of the steel were quenched from a temperature of 900° C. in a bath at 200° C. for comparison with the samples similarly treated from an initial temperature of 845° C. (already referred to). In all cases the time at the high temperature was 10 min. The curves are shown in Fig. 21.

The only variation in behaviour appears to be a slight increase in the rate at which the transformation proceeds in the case of the higher temperature quenchings. The V.P.N. of the samples treated at 900° C. gave an average of 707, *i.e.*, slightly lower than that for those quenched

from 845° C. (V.P.N. 721). It is doubtful, however, whether it is justifiable to believe that any real difference exists.

Neither the quenching temperature nor the time at this temperature, then, appears to have any effect on the Ar″ transformation, though both should result in a variation of grain size. This result is in accordance with that of Greninger and Troiano,[36] who say, "grain-size variations were found to have little, if any, effect on the amount of martensite present," as well as with the results on a chromium-molybdenum steel of Griffiths, Pfeil and Allen.[38]

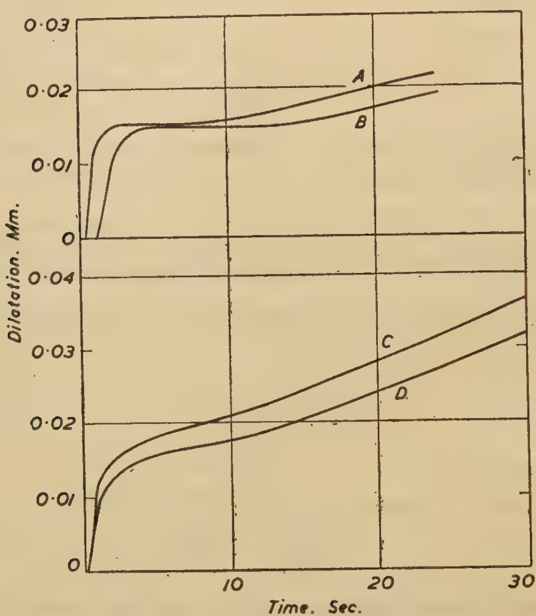

Fig. 21.—Steel *A* Specimens, 0·3 × 0·875 in., transformed at 200° C. Curves *A* and *B*, quenched from 845° C.; *C* and *D*, from 900° C.

The Cooling Rate from the Quenching Temperature.

It is well known [26, 39] that the cooling curve in a normal quenching medium consists of an initial portion during which the rate of temperature fall is low, and which has been ascribed to the blanketing of the specimen by gases given off by the liquid.

In quenching in a metal bath the extent to which vaporisation occurs is exceedingly small, and an immediate quenching effect may be expected as soon as the sample comes into contact with the quenching bath. This is shown to be the case in Fig. 22, which represents the photographic record of a steel quenched from 810° C. in a bath at 283° C. Some compensation for the relative slowness of the quenching operation in these hot baths is thus obtained.

Provided that the critical rate of cooling is exceeded and the sample enters the bath while still in the austenitic state, *i.e.*, that the Ar′ point has been suppressed, the rate of cooling would not be expected to exert any appreciable influence on the formation of martensite, nor have the authors been able to detect any.

When the quenching is rendered less drastic and troostite is formed during the cooling to the bath temperature, this constituent, by providing nuclei, may accelerate the velocity of transformation. The curves for samples 27/1 and 28/1 in Fig. 23 show, however, that a small amount of

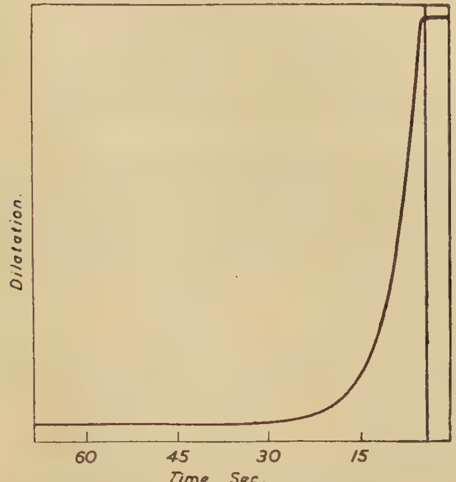

Fig. 22.—Photographic Dilatation/Time Curve for Steel D Specimen, quenched from 810° C. in a bath at 283° C.

troostite formed during the quenching has little, if any, further effect on the subsequent transformation. As Fig. 38 shows, a small amount of grain-boundary troostite was present in the case of sample 28/1, whilst sample 27/1 (Fig. 39) was entirely martensitic. Despite this difference

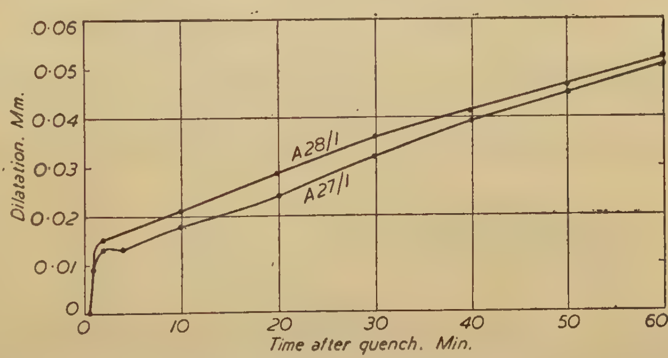

Fig. 23.—Dilatation/Time Curve for 0·3 × 0·875-in. Specimens of Steel A, quenched from 900° C. in a bath at 200° C. $A$28/1 produced some troostite. $A$27/1 was entirely martensitic (*see* Figs. 38 and 39).

of structure the rate of change indicated by the slopes of the curves in Fig. 23 was essentially identical.

The microscopic evidence is equally clear in demonstrating the absence of any influence of the troostite on the subsequent decomposition of the

austenite. As will be seen from Fig. 38, the dark martensite needles formed in the bath show no tendency to cluster around the areas of troostite, but are uniformly dispersed throughout the austenite. (The white areas in Fig. 39 are, at room temperature, martensite of the slow-etching variety, which the authors have attempted to identify with the tetragonal form.)

The Pre-existing Structure.

Several workers have suggested that the results of heat treatment are dependent on the previous structure of the steel, an idea which, if confirmed, the work of Roberts and Mehl,[10] for instance, goes some way towards explaining.

Specimens of steel *A* were, therefore, treated so as to give a range of structures varying from a coarse spheroidal pearlite, through one in which this was mixed with still laminated carbide, to fine pearlite and sorbite. The structures of the spheroidal and finely laminated pearlites are shown in Figs. 40 and 41. These were then heated to 900° C., kept there for 10 min., and then quenched in a bath at 200° C. The structures are shown in Figs. 42, 43, 44, and 45.

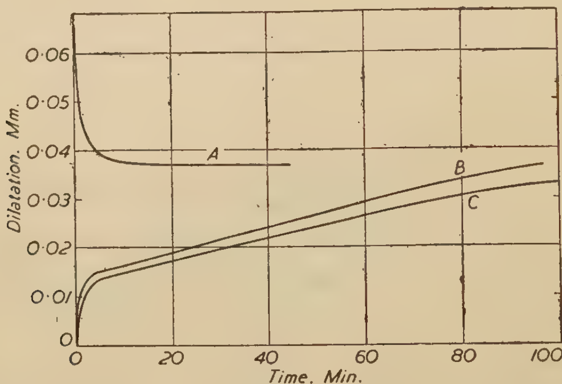

Fig. 24.—Dilatation/Time Curves for Specimens illustrated in Figs. 42 (curve *B*), 43 and 44 (curve *C*) and 45 (curve *A*), quenched from 900° C. and transformed at 200° C.

The corresponding dilatometer curves are shown in Fig. 24, while the average hardness values after the transformation are shown below:

Structure.	V.P.N.
Spheroidal pearlite	676
Spheroidal and laminated pearlite	697
Finely laminated pearlite	700
Sorbite	483
As-rolled	706

Apart from the spheroidal specimen, which is somewhat softer than the pearlitic ones, and the sorbitic sample, which is very distinctly softer, these figures are all of the same order, and indicate that, within fairly wide limits, the degree of lamination of the pearlite does not exert any marked influence.

The result for the sorbitic material was so surprising that the tests on this and on the spheroidal pearlite were repeated, and gave the same result. Why a steel initially sorbitic should react so much more markedly is inexplicable, and is an observation which clearly requires further substantiation.

That the pre-existing structure may in some circumstances affect the result can be demonstrated in still another manner. If a normal sample of steel is given a treatment which is then repeated the structure before the second quenching will consist of sorbite (formed from martensite during the heating up for the second quenching operation), whereas in the first instance it consisted of a laminated pearlite. Still further quenchings will be carried out on a sample always with a similar, sorbitic, structure. If, then, a given sample is repeatedly treated, the first curve may be expected to differ from the remainder, which should be similar to each other. This is the case in the curves reproduced in Fig. 25, which refer

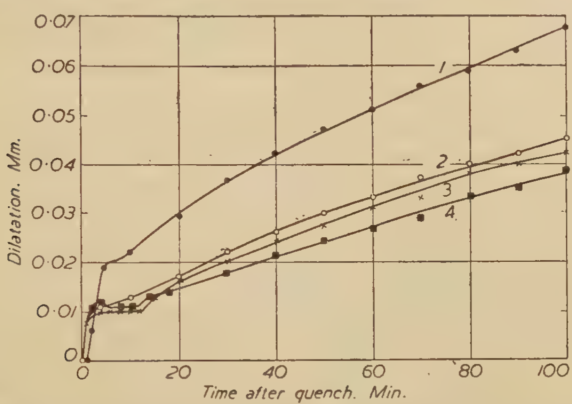

FIG. 25.—Dilatation/Time Curves for a 0·3 × 0·875-in. Specimen of Steel *A*, repeatedly quenched from 900° C. and transformed at 200° C. Curve 1 represents first treatment, curves 2, 3 and 4, succeeding ones.

to four successive treatments of steel *A* from a temperature of 900° C. in a bath at 200° C. Quite apart from the slope of the curves, the length of the induction period is clearly increased in the second and subsequent treatments. The hardness figures were, however, practically unaffected, being 702 after the first treatment and 706 after the fourth. Similar results were obtained on steel *D*, quenched from 830° C. in a bath at 240° C. The first curve obtained fell away from the remainder, which were all much the same.

PART III.—SUMMARY AND CONCLUSIONS.

(1) Believing that the fundamentals of an industrial process must be understood before it can be carried out with maximum efficiency, the authors have devoted the larger part of this paper to a consideration of the scientific background of the mechanism of the isothermal transformation process, and have allowed themselves considerable latitude in their speculations. If in a given sample of an isothermally transformed steel a fully martensitic structure can be retained, it is clearly possible to modify the conditions so as to give any softer steel that may be desired. It is, therefore, to the transformation of austenite to martensite to which the authors have devoted their main attention.

(2) For the purpose of studying this breakdown a quenching apparatus was designed capable of providing a definite cooling rate, which, judged by the uniformity of the structure or hardness of the product or by the rate of change of length of the sample, gives good reproducibility.

Along with this a dilatometer of considerable sensitivity was employed. It should be pointed out, however, that the change of length of the sample may not be exactly proportional to that of volume. This is a potential source of error in all such dilatometric work, but the authors' results have done nothing to throw doubt on the conclusion of Bain and Waring [40] that, with samples of constant size and shape, the proportionality in the changes of length and of volume is close.

(3) So far as comparison with the results of other workers goes, the main difference in the results obtained on these considerably bigger samples is in the length of the induction period, which is a far less pronounced feature of the transformations studied by the authors.

(4) When samples of very similar "ordinary" analyses, but obtained at different times, are examined, marked variations in the results are found. A possible explanation of this is to be found in the varying amounts or states of some undetermined element, possibly (probably ?) oxygen. If this shifts the S-curve to the left, *i.e.*, in the direction of increased reactivity, then, for a given rate of cooling, the normal curve will be cut at some temperature distinctly lower than that for the "abnormal" steel. In one case martensite may result, and in the other, troostite, and an explanation of "abnormality" is at once obtained. This is in complete agreement with Bain's view that, "an abnormal steel is one of the fast-reacting type," the increased speed of transformation being possibly due to alumina inclusions which act as nuclei. In a similar manner marked vertical displacement of the curve may occur.

(5) On the basis of Tammann's curve for the change of the velocity of crystallisation with temperature an explanation is offered for the shape of the S-curve itself. On exactly the same lines rather hypothetical reasons are adduced for the shape of the areas in which pro-eutectoid phases are precipitated from the austenite, and for the extra hump found in the curves for steels in which separate carbides are formed, *e.g.*, chromium and molybdenum.

(6) Alongside these curves the microstructures which can be obtained are discussed. It is shown that in some steels there is very strong evidence in support of the view that in a certain temperature range, say, 450–550° C., there is a labile shower of acicular crystals of ferrite. This hypothesis is in excellent agreement with the very rapid transformation which is found in the same temperature zone. Fig. 4, curve 5, and Fig. 12, curves 2 and 4, for instance, of the paper by Flinn, Cook and Fellows [8] show a velocity of change for a chromium-molybdenum steel which is much greater at about 700° F., the middle of the temperature interval in which acicular structures are obtained, than either above or below. That this precipitated acicular ferrite contains carbon in solid solution is shown by its behaviour on more prolonged soaking, and by the gradual change in the etching characteristics as the temperature of treatment falls. This retention of carbon is clearly a result of its diminished rate of diffusion into the surrounding austenite, a process which has been confirmed by X-ray examination. The gradual rate of the change of this ferritic structure through bainite to typical martensite is a metallographic feature worthy of note.

If some name is required for this ferrite supersaturated with carbon, might not Benedicks' term "ferronite" be revived ? It is true that it was originally coined to denote the iron component of pearlite, but this in Benedicks' view was again an iron-carbon (carbide) solid solution.

(7) The possibility of the carbon change-point being depressed by stress, so that the austenite is rendered stable at distinctly lower temperatures than the normal Ar_1 point, is considered, but the effect is shown to be so small as to be without importance from the present point of view.

(8) The period of induction is discussed at length. After reviewing the various reasons for such an effect, it is believed that the cause is to be found in the gradual relaxation of the stresses set up as a result of the sudden cooling, or of the volume change at the surface due to the formation of a skin of martensite. Further, the marked parallelism between the curve for the commencement of the breakdown and that for its completion suggests that the former corresponds to the time at which the transformation becomes measurable, or, in other words, that there is in reality no true period of induction, the change going on all the time, but at a rate so slow as to be initially undetectable.

(9) If the induction period is, as the authors suggest, a period of stress relaxation it might reasonably be expected that it should bear some relationship to the creep properties of the steel. That this is qualitatively so is quite clear from the general shift of the curves with change of composition.

Recorded below are some values of the creep limit given by Tapsell,[41] together with the times at the same temperature required to give the first detectable breakdown of the austenite. A temperature range around 500° C. has been chosen, *i.e.*, that at which the transformation is generally most rapid. All the creep values given are for material which has received no special heat treatment:

Steel.	Temp. ° C.	Time for Commencement of Change. Sec.	Creep Limit. Tons per sq. in.
C 0·25%	500	<0·5	2·4– 4·2
C 0·25%, Mn 1·7%	500	0·9	5·0
C 0·25%, Ni 3·25%	500	0·8	4·7– 5·2
C 0·17%, Ni 5·2%	450	1·5	6·5– 7·6
Ni–Cr–Mo	450	12·0	22·4–24·4

Although the compositions of the two series of steels are not the same, there can be little doubt about the general parallelism of the results.

(10) The dilatation curves obtained consist, in their most general form, of five stages, which pass imperceptibly into each other. First, there is a rapid expansion due to the formation of the martensitic skin, which dies away and leads [(2)] to a stage of temporary equilibrium analogous to the induction period. This is succeeded [(3)] by a second expansion, initially exceedingly slow, but increasing in velocity until, over a limited range,[(4)] it gives a curve which coincides with that of a pseudo-first-order reaction, *i.e.*, one in which the rate is proportional to the amount of unchanged austenite. The fifth and final portion consists of a curve indicating a diminished rate of change but which is still measurably greater than that which the first-order reaction would require. This part of the curve does not become asymptotic to the ordinate corresponding with the complete breakdown, but cuts it at a definite angle. The rates of change corresponding with these stages are shown in Fig. 26.

How far nucleation effects exert an influence on the formation of martensite is a matter which is far from certain. A nucleation curve can be fitted well on to section 2 and the earlier portion of section 3. Further, since it is difficult to deal with the restrictions imposed as a result of the high viscosity of the material, a sort of " bottleneck " may be set up, as a result of which the rate of nucleation and subsequent growth will be materially decreased. A still better approximation to section 3 of the observed curve would thereby result. Even taking this into consideration, however, the authors are not disposed on the evidence available to believe that nucleation is a process which plays any considerable part in the formation of martensite, a process which they think bears a marked similarity to the production of mechanical twins.

The complication introduced by the existence of two forms of martensite has been considered. It is shown that there is some experimental justification for the belief that at "low" quenching temperatures tetragonal martensite may first be formed and then tempered almost at once to the cubic variety.

(11) The tempering curve of a martensitic steel, transformed isothermally, differs radically from that of oil- or water-quenched material. In the former there is little or no evidence of the contraction between 100° and 200° C. owing to the martensite inversion. The samples are, therefore, already (mainly) in the cubic state. Further, the hardness figures for a given steel are markedly different when the specimen is water-quenched, or isothermally transformed to give a completely martensitic structure, and then tempered at a temperature above that at which the solid solution has broken down. The hardness of the steel quenched in a

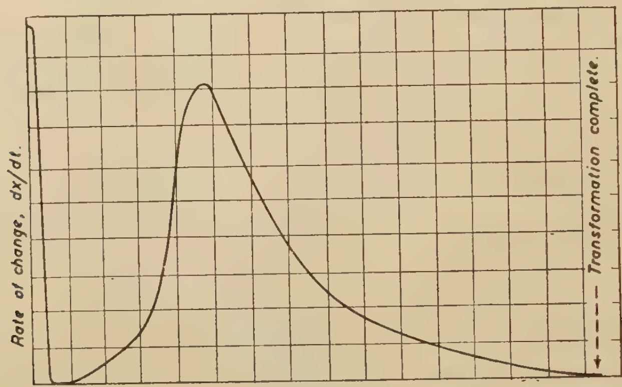

FIG. 26.—Variation in Rate of Austenite Transformation at an Isothermal Temperature with Time.

molten metal bath, when subsequently retempered, is appreciably higher than that of the water-quenched material similarly treated. This can only mean that the martensite produced by the former treatment is more stable than that of the water-quenched steel, which is exactly what would be expected if it were in a condition of much lower internal stress.

(12) The martensite in the tetragonal condition etches very slowly, and shows light needles in a darker-etching background. That from steels which show no inversion change and which, therefore, are presumably already cubic, etches far more readily, and it is the needles, then, which are dark, the background being almost unattacked.

(13) Since the possibility of treating samples of some considerable size and yet retaining a fully martensitic structure was one of the aims of the more practical aspect of this work, the maximum size of bar which could be so treated was investigated at length. The authors have succeeded in doing this in the case of a eutectoid steel in bars of up to 0·3 in. in dia., and for a 1·2% carbon steel, quenched above the cementite line, up to 0·45 in. in dia. These results are much in advance of anything hitherto recorded. It is clear that if a troostitic structure is adequate, *e.g.*, for certain types of springs, austempering might be feasible up to even bigger dimensions. The nickel-chromium steel, as would be expected, remained fully martensitic at the largest size with which the authors could deal.

(14) The effect of grain size has been discussed in some detail, and it is shown that there are strong reasons for the view that, so far as the Ar′ point is concerned, it exerts a considerable effect, the steel reacting more and more rapidly as the grain size becomes smaller. The Ar″ change, however, is apparently unaffected, neither variations of quenching temperature nor of time at that temperature having, in the authors' investigations, any effect on the subsequent transformation of the austenite.

(15) Earlier results which suggest that quenching in a metal bath shows little sign of the initial period of very slow fall of temperature so characteristic of normal quenching operations are confirmed. Rapid cooling sets in within a fraction of a second of immersion, and, as a result, the quenching is distinctly more effective than might at first sight be expected. Further, the authors' results have shown no measurable difference, other things being the same, when the composition of the bath, and hence its thermal properties, is altered.

It would be desirable to have proof that the precipitation of a pro-eutectoid phase at very high austempering temperatures has not sometimes occurred during the quenching itself. There is proof [42] that reactions can take place in the specimen prior to reaching the bath temperature, this being greatest in the case of quenchings in a medium at about 550° C. Recalescence may then take place during the cooling to an extent of nearly 50° C. The more rapid fall of temperature associated with quenchings in baths at lower temperature in which martensite can be retained do not, however, show any signs of such preliminary breakdown of the solid solution.

If the rate of cooling is not sufficiently rapid to suppress the Ar′ change entirely, it has been shown that the subsequent transformation in the bath is not materially affected by the presence of the troostite.

(16) Experiments on the effects of the pre-existing structure upon the results of isothermal transformation have indicated the rather surprising reactivity of samples which, *before this treatment*, were in the sorbitic condition. To a lesser degree the same sort of result was obtained from samples consisting of spheroidal pearlite. The effect for ordinary sorbite was checked by repeatedly treating the same (originally laminated) sample. The first curve, from the pearlitic steel, was appreciably different from the three succeeding ones, which just below the Ac_1 point would be sorbitic.

So far as the results on pearlite of varying degrees of fineness are concerned neither the dilatation curves, structures nor hardnesses showed any appreciable variation.

(17) It would appear, then, that, apart from changes of composition in the ordinary sense, the two factors which, more than any others, affect the results of isothermal treatment are, first and foremost, the degree of deoxidation, and secondly, the existence of a sorbitic structure in the material before treatment.

(18) It has been suggested that the breakdown of austenite may not be the same when quenching from a state of complete equilibrium as when quenching from a temperature range in which diffusion is proceeding.

(19) On the fundamentally important subject of the reasons for the difference of properties obtained from a normally quenched and an isothermally transformed steel, respectively, this work seems to have thrown some light. At temperatures above that of the Ar″ change there is, of course, little difficulty. The fact that the aggregate, troostite or sorbite, is formed directly from austenite without passing through the intensely stressed martensitic state will eliminate much of the danger of the formation of micro-fissures.

In those cases, however, where martensite is formed the difference

may be reasonably ascribed to the production, in the metal bath, of the cubic form directly, or at any rate through the intermediary of tetragonal martensite of the most transitory existence. The absence (or, one ought rather to say, the low degree) of internal stress in martensite formed at the temperature of a molten metal bath is evidenced by its resistance to further tempering. As a result, although such martensite is originally distinctly softer than that formed by normal quenching, the tempered product may be distinctly harder (and, of course, tougher).

(20) In conclusion, the fact must again be stressed that this contribution to the subject has appeared before finality has been attained in a number of directions, even within the limited field of this stage of the authors' investigation. Infallibility is claimed neither for the results nor for the hypotheses here offered; in fact, the more the authors are in error, the more likely it is that this will be pointed out and knowledge of the subject thereby advanced.

If there is one point in connection with the more theoretical aspect of the work, which cries aloud for elucidation, it is the condition of the carbon and/or carbide when in solution in γ-iron.

Acknowledgments.

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CORRESPONDENCE.

Mr. D. E. R. Hughes (Tube Investments, Ltd., Birmingham) wrote: The work described in this interesting paper represents a distinct advance in the technique of dilatometric measurement applied to the study of austenite decomposition. There are one or two points, however, to which I would like to refer.

First, with regard to the question of the existence of cubic martensite, the authors state that, as a result of the work of Honda and Nishiyama, it is quite clear that two forms of martensite (tetragonal and cubic) exist. I cannot agree with this, since the X-ray evidence presented by the Japanese workers is open to well-founded criticism—as pointed out by Hägg.* The merging of the tetragonal diffraction doublets into single diffuse lines which occurs on tempering quenched martensite cannot be taken as sound evidence for the formation of cubic martensite, since the degree of tetragonality of quenched martensite depends on the dissolved carbon content. Hence, a tetragonal structure with only a little carbon in solution would be difficult to distinguish, by X-ray methods, from a cubic structure with all the carbon in solution. The authors have presented no new evidence tending to confirm the existence of cubic martensite.

Secondly, in my opinion, there is a considerable weight of evidence in support of the theory that martensite cannot form isothermally, but only during cooling through an appropriate temperature range. It has been shown quite clearly that for a steel of given analysis there is a fairly definite temperature (often referred to as the Am point of the steel) above which martensite cannot form. If an austenitised steel sample is quenched into a bath held at a temperature which is below that corresponding with the particular Am point, then formation of martensite takes place only during cooling from Am to the bath temperature. After cooling to the bath temperature we get in general a period of induction followed by an isothermal transformation. I think that the authors may be right in their interpretation of the induction period as a period of stress relaxation, but I consider that the nature of the isothermal product formed at temperatures below Am is still rather obscure. If the product is cubic martensite (*i.e.*, a cubic structure with all the carbon in solution), then it is difficult to see why its formation should take a comparatively long time. On the other hand, if the transformation involves some sort of precipitation of a carbon-bearing phase (as the appearance of etched microsections suggests), then the effect of time is more understandable, since diffusion will presumably play a part.

Thirdly, in their discussion of factors which affect the process, the authors do not mention the effect of varying the quenching-bath temperature. The bath temperature of 210° C. used for the study of the martensite skin effect in steel *L* is not far from the Am point for this steel; if the Am point for the steel was, say, 220° C., then formation of martensite would occur during cooling from 220° to 210° C. Since the surface of the specimen would cool comparatively rapidly to 210° C., formation of martensite would take place almost immediately at the surface. The centre of the specimen, however, would cool comparatively slowly from 220° to 210° C.—if it ever did reach 210° C. in a reasonable time. It is possible, therefore, that the martensite skin effect would be misleading if the quenching bath temperature was near to the Am point of the steel, especially if the specimens used were comparatively large. The skin effect would also tend to be accentuated if any surface decarburisation occurred during austempering. I would be interested to know whether the authors have any information regarding

* G. Hägg, *Journal of The Iron and Steel Institute*, 1934, No. II., p. 439.

the effect of quenching-bath temperatures below 210° C. on the behaviour of steel *L*.

Finally, in their last sentence the authors refer to the need for research concerning the condition of carbon when in solution in austenite. Has not the work of Petch * thrown considerable light on this question ? As regards the nature of the low-temperature decomposition product of tetragonal martensite, I believe that the study of single crystals of martensite would yield useful information. Though I am not familiar with their work, I understand that certain Russian investigators † found new diffraction lines appearing after tempering single crystals of martensite.

Mr. D. C. Moore (Tube Investments, Ltd., Birmingham) wrote : I read this paper with no little interest and would like to congratulate the authors on their most excellent apparatus and experimental technique. The interpretation of the results is, however, in my opinion, rather odd, especially the acceptance of the cubic martensite theory of the Japanese workers so successfully disproved by the work of Hägg. The isothermally transformed material which they refer to as cubic martensite is in reality not martensite but a material consisting of precipitated carbide in a matrix that has a slightly tetragonal lattice and could easily be mistaken to be cubic in the X-ray diffraction pattern—the carbide would not show, being finely dispersed. This is borne out by the fact that the material is rapid-etching, and homogeneous cubic martensite (if it existed) would be slow-etching. Similarly, in the tempering of true martensite there is not a tetragonal cubic inversion but a diffusion of carbon and precipitation of carbide, with an accompanying progressive change in lattice dimensions towards cubic, but never quite reaching it.

The authors ask us to take the fact that compressional stress stabilises austenite to explain why there should be an induction period, but surely the volume change resulting from the transformation of the austenite at the skin would put the centre of the specimen into tension, which would favour decomposition. It then appears that the stress-release idea can be ignored. The theory that the induction period is a period of diffusion and precipitation of carbide, with subsequent facilitation of austenite breakdown due to the nucleation effect, is certainly more acceptable—*cf.* Bain's suggestion that the increased reaction speed in " abnormal " steels results from the presence of nuclei.

The authors discuss failure to reproduce results on materials of apparently similar composition. With regard to Fig. 6, have they considered the effect of nitrogen on the stability of austenite in steels of this type ?

Mr. E. H. Bucknall (Research and Development Department, The Mond Nickel Co., Ltd., Birmingham) wrote : Professor Thompson and Dr. Stanton should be congratulated on the apparatus which they have devised for recording autographically length changes and hence the progress of transformation in steels when quenched into a bath held at a fixed temperature, but the question arises whether specimens of the relatively large diameters employed are best suited to this type of work.

The methods employed for the determinations of the transformation characteristics of alloy steels in the laboratories of The Mond Nickel Co., Ltd., remain essentially as described by Allen, Pfeil and Griffiths in 1939,‡ but it has become clear that the technique needs modification for use with

* N. J. Petch, *Journal of The Iron and Steel Institute*, 1942, No. I., p. 111 P.

† M. Arbusow and G. Kurdjumow, *Journal of Physics, Academy of Science, U.S.S.R.*, 1941, vol. 8, pp. 101–108.

‡ Second Report of the Alloy Steels Research Committee, *The Iron and Steel Institute*, 1939, *Special Report No.* 24, Section XIII.

fast-transforming steels, particularly plain carbon steels, if transformation before the attainment of the bath temperature is to be avoided. In the case of the 0·3% and 0·4% carbon steels mainly studied, this seems very desirable, since separation above the bath temperature is usually of ferrite and results in carbon enrichment and change in the transformation characteristics of the remaining austenite. This factor does not enter into the work described in the present paper to the same extent, since the initial separation is probably as a rule of troostite, but would seem to militate against the wider use of the method. For the present purpose a large specimen may perhaps be regarded as useful in limiting the effective accidental decarburisation at the austenitising temperature on the form of the dilatometer curve, but it seems dangerous to rely on subsequent microscopical examination of this type of specimen. When use is made of a 0·11-in. dia. specimen, it is necessary to take considerable precautions against decarburisation, particularly with the higher-carbon steels, where the use of a thin electrodeposit of nickel, with nitrogen in the preheater, does not seem to suffice.

The main experimental evidence of premature transformation is failure of the observed contraction to match up with that expected from the temperature drop and the coefficient of expansion of the austenite. Under

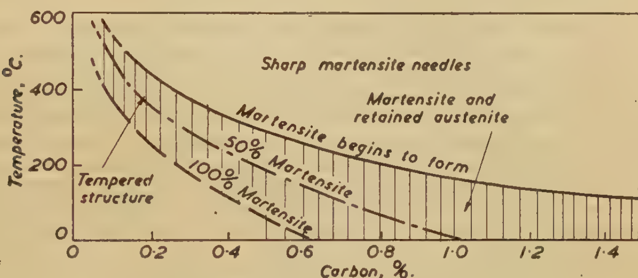

FIG. A.—Martensite Range for 3·2% Nickel, 0·7% Chromium Steels.

such circumstances the contraction curve breaks away from that normally expected at some temperature above the bath temperature and the recorded expansion is correspondingly reduced. The same type of experimental curve is usually obtained when dealing with temperatures at which the martensitic change can occur, however slow the transformation of the steel at higher temperatures, indicating that the martensitic change has a definite temperature of onset and proceeds extremely rapidly at first. Experiments which have been carried out on six types of alloy steel, each with a variety of carbon contents, have confirmed qualitatively the picture of martensite formation drawn by Carpenter and Robertson, who assume that the martensite change is a non-isothermal transformation and the temperature at which it starts decreases with carbon content along a curve which, at first, falls steeply and then flattens off in the neighbourhood of 1–1¼% of carbon.

In experiments on a 3·2% nickel, 0·7% chromium series, the relationship was as shown in Fig. A, and it did not seem to be much altered when the total alloy content was reduced to a level of 1% of nickel or 1% of chromium. With still lower alloy contents the position of the curve for the onset of the martensitic transformation could not be fixed, owing to experimental difficulties, and it was not at all clear what was the position in plain carbon steels. In these experiments the usual observation was

that a specimen could be quenched successfully without prior transformation to any temperature above the martensite onset curve, but transformation set in at the appropriate temperature on this curve when an effort was made to quench to temperatures in or below the martensitic range. Once transformation set in, it went on during cooling to a degree of completeness depending on the bath temperature. If this was below the curve indicating completion of the martensitic change, the expansion accompanying transformation was followed by a final contraction corresponding to cooling to bath temperature. With the 3·2% nickel, 0·7% chromium series it will be noted that the temperature of onset of the martensitic change falls below 200° C. when the carbon content is above about 0·8%. As the intermediate transformation in this class of steel can only run at an appreciable rate above 200° C., it follows that steels with carbon contents in the neighbourhood of 1% have a range of temperature between about 200° C. and the temperature of onset of the martensitic transformation in which austenite is apparently stable. On the other hand, when the carbon content is considerably below 0·8% it is possible to quench a specimen to a temperature within the martensitic range and so secure a partial martensitic transformation, and then have an isothermal stage of transformation. In such cases the final stage has been regarded as corresponding to the intermediate transformation mechanism rather than the transformation to true martensite.

It is felt that it should be possible to reconcile these results on alloy steels with those now presented on carbon steels, and it is, therefore, queried whether a possible construction which could be put on the experimental results is that in the plain carbon steels *A* and *L* the temperature of the onset of the martensitic reaction is raised and possibly the lowest temperature at which the intermediate reaction will proceed at an appreciable speed is lowered, giving rise to the possibility of the same sequence of changes at high carbon levels as has been observed in alloy steels only at low carbon levels. The trends of the curves shown in Figs. 3 and 21, for example, seem to support this view. The authors' views on this point would be appreciated.

Mr. B. E. Hopkins (Research and Development Department, The Mond Nickel Co., Ltd., Birmingham) wrote : There are a number of points in the explanation in terms of compressional stresses suggested by the authors to account for the induction period prior to observable transformation which need further consideration before it can be accepted. It is proposed to deal with some of these points only. An explanation for periods of induction must, quite obviously, be such that the assumptions made do not render it difficult to account for the remainder of the transformation. Quite apart from the possibility that the so-called martensitic layer formed at the surface of the dilatometer specimens during the quenching operation, or immediately afterwards, is due to decarburisation occurring during the preheating, it is necessary to account for the increasing speed with which the transformation occurs at a subsequent stage. According to the theory put forward by the authors, the transformation proper starts when the stresses imposed on the austenite have been relaxed sufficiently at the temperature of the quenching bath. The fact that this transformation—resulting, as it does, in an expansion—imposes further stresses on the remaining austenite needs consideration. To be consistent, it is essential to believe that these new stresses have to be relaxed at the temperature of the quenching bath before further transformation can occur. Once the transformation of the main mass of the austenite starts, it is difficult, in the light of the authors' theory, to see why this transformation then takes place at an increasing rate. At the most, one would

expect a balance to be set up between the rate of transformation and the rate of relaxation of stresses arising from this transformation.

Periods of induction are common to all temperatures of isothermal transformation from the Ae_1 point downwards, and from their nature it is felt that it is likely that the same, or at least a similar, explanation will account for the phenomenon at all temperatures. It does not seem that this expectation can be realised when applying the authors' theory to higher temperatures. If we take the case of a steel transforming isothermally in the pearlitic range, where long periods of induction are possible, the formation of a so-called martensitic layer at the surface of a specimen is ruled out, but a parallel formation of ferrite can occur. At such temperatures the stresses imposed upon the austenite as a result of quenching into the metal bath, and also as a result of transformation, are much less than at low temperatures, owing to the smaller temperature drop and the considerably smaller expansion, respectively. Furthermore, the rate of relief of stresses is considerably greater. Nevertheless, long periods of induction can be obtained in the pearlitic range.

It is well known that the period of induction is a complex function of temperature, and in a temperature range below the pearlitic transformation longer periods of induction are obtained than at higher or lower temperatures. This is particularly so in the case of some alloy steels, where a region of high stability of the austenite is often met with. It is difficult to see how these points can be explained in terms of a theory developed on the basis of the effect of stresses.

The results obtained in the experiments conducted to determine the effect of pre-existing structure on isothermal transformations at a temperature of 200° C. need some explanation. Some of these results, which refer to the 0·82% carbon steel *A*, are given in Fig. 24. A number of points emerge from a scrutiny of these curves. To begin with, curve *A*, representing the specimen which was pretreated to give a sorbitic structure, indicates that the specimen did not cease contracting on quenching into the metal bath until at least 10 min., and probably more like 20 min., had elapsed. This feature is also apparent in some of the other curves presented in this paper, and is due, no doubt, to the large specimens employed by the authors, such specimens taking a comparatively long time to reach the quenching-bath temperature. Presumably the experimental procedure was the same for the specimens represented by curves *B* and *C* as for specimen *A*. From this it follows that the specimens pretreated to give spheroidal pearlite and pearlitic structures did not contract to their full extent, since rapid transformation with resulting expansion occurred immediately they were quenched into the metal bath. Consequently, the total expansion was considerably less than it would have been if the specimens had reached the metal-bath temperature before transformation set in. This is a serious objection and renders the drawing of concrete conclusions more difficult. The specimen pretreated to give a sorbitic structure, and represented by curve *A*, did not exhibit an expansion at all. Thus, most of the transformation that did occur in this specimen must have taken place during the quenching operation, and all transformation must have ceased before the whole of the specimen reached the bath temperature. This means that the transformation of the specimen pretreated to give a sorbitic structure was fast, and, of all the specimens, this is the one which would be expected to have the most homogeneous austenite, owing to the fact that the carbide particles present in the sorbite before preheating to 900° C. were small. The specimens pretreated to give structures of spheroidal pearlite and of pearlite would not be expected to yield an austenite so homogeneous as the one pretreated to give sorbite, since the carbide particles present before preheating to 900° C. were considerably larger. In

fact, the austenite in these cases was probably very non-uniform as far as carbon distribution is concerned, since it is doubtful if the middle of the specimen ever reached the stated temperature of 900° C. in view of the short time allowed (10 min.). This statement is supported by the curve of Fig. 19, from which it appears that the specimen was still expanding slightly when quenched. In any case, it seems reasonable to assume that in the specimens pretreated to give spheroidal pearlite and pearlite the carbon concentration in the austenite was sufficiently non-uniform immediately before quenching to result in areas which varied widely in the facility with which transformation could take place. The faster-transforming areas probably transformed very quickly during the quenching operation, giving no indication on the dilatometer curve, but merely resulting in the contraction being less than it should have been. The slower-transforming areas would transform less readily than an austenite of the average carbon content, and it is suggested that the expansion resulting from this transformation is that given in curves B and C. The relative position of curves B and C is not inconsistent with the expected variation of carbon concentration judged from the initial structures. If the specimens had been soaked at 900° C. for longer times the difference in behaviour of the specimens during isothermal transformation would be expected to become less; in other words, the pre-existing structure should have an effect only if the austenite is not adequately homogenised before quenching into the metal bath. These remarks are offered in explanation of the different behaviour of the specimen pretreated to give a sorbitic structure as compared with the other specimens. It is realised that other factors, such as the grain-boundary effect, have to be taken into consideration in order to explain completely the structures obtained after isothermal treatment as shown in Figs. 42 to 44, but it is felt that such factors merely complicate the issue and do not affect the main trend of the argument.

Mr. A. U. Huddle (Cambridge University) wrote : In this paper the authors have revealed some interesting facts associated with the decomposition of austenite. It appears to the present writer that there are three essential factors which control the reaction :

(i) The *unhomogeneity* of the austenite as it is about to transform (*i.e.*, concentration fluctuations of carbon, &c.).

(ii) The *driving force* of the reaction at the transformation temperature (*i.e.*, the thermodynamic free-energy change).

(iii) The *resistance of the crystal lattice* to either the direct shearing process, as in the formation of martensite, or to the diffusion of solute elements, as in the formation of pearlite.

Recent work by X-ray crystallographers has shown that :

(*a*) Cementite * (Fe_3C) is essentially a 25 atom-% interstitial solid solution of carbon in austenite.

(*b*) Martensite † is a supersaturated interstitial solid solution of carbon in ferrite, the cubic lattice being slightly distorted.

(*c*) The transformation of austenite to martensite ‡ takes place by a shear on fixed lattice planes in definite directions, the movement of each adjacent plane being just less than one-third of the interatomic spacing.

* N. J. Petch, *Journal of The Iron and Steel Institute*, 1944, No. I., p. 143 P.

† H. Lipson and A. M. B. Parker, *Journal of The Iron and Steel Institute*, 1944, No. I., p. 123 P.

‡ G. Kurdjumow and G. Sachs, *Zeitschrift für Physik*, 1930, vol. 64, p. 325. *See also* (for summary) C. S. Barrett, "Structure of Metals," pp. 475–491. New York and London, 1943 : McGraw-Hill Book Co., Inc.

It is well known that pearlite is nucleated by cementite.* Now, an abnormal steel is clearly of the fast-reacting type, and it therefore seems probable that the explanation of abnormality lies in the *unhomogeneity* of the reacting austenite. It has previously been suggested † that the primary effect of aluminium additions to steel is to lower the rate of diffusion of carbon in austenite, thereby retaining the carbon concentrations during the period of heating in the γ-range. These concentrations of carbon atoms may readily nucleate the cementite with quite small "driving forces" of the reaction $\gamma \rightarrow \alpha + Fe_3C$, and therefore an abnormal structure is obtained (with its corresponding physical properties) with cooling rates which would normally give insufficient time for the formation of cementite nuclei. Is this not the reason why a specimen, initially sorbitic, should behave abnormally? A large number of the cementite precipitates present in the sorbite may not diffuse completely into the surrounding austenite during the 10 min. at 900° C. and may act as nuclei when the steel is subsequently quenched.

The other interesting point is the change from white to black martensite on heating to about 150° C. The sudden change of dimensions and etching characteristics is surely due to the initial precipitation of cementite. The precipitates will give definite cathodic points and will cause rapid etching in comparison with the slow etching (almost equipotential surface) of the white tetragonal martensite. The authors find that this change is accompanied by an increase of hardness such as is usually associated with the initial precipitation from a supersaturated solid solution.

Mr. G. Burns (Admiralty Engineering Laboratory, West Drayton) wrote: The paper by Professor Thompson and Dr. Stanton provides some interesting data and suggestions on the isothermal transformation of steels, but, unfortunately, some of the points made lose their full force by the omission of specific information.

For instance, it is not made clear whether the similarity of the two steels for which dilatation/time curves are given in Fig. 4 extends to their austenitic grain-size at the quenching temperature, 810° C.; in view of the fact that the difference of behaviour of these steels is made a basis for a hypothesis of "abnormality" there is a temptation to infer that the steels were of different grain-size at the quenching temperature, but a definite statement on the point would be of value. Again, in considering the effects of quenching temperature and of time at temperature on the transformation characteristics, the authors state that in their experiments neither factor appeared to have any effect on the Ar″ transformation, though both should result in a variation of grain-size; the force of this statement would have been increased had the authors stated over what range the austenite grain-size of their steels was actually varied by the conditions of their experiments.

In developing their suggested explanation of abnormality from the behaviour of the type shown specifically in Fig. 4 and in a generalized form in Fig. 5 the authors take the view that the right-hand curve is, in each case, the normal one for the material, whilst the left-hand curve represents the behaviour of material of the same general composition but containing, in addition, a constituent not normally estimated and which has the effect of moving the S-curve to the left. In view of the fact that this additional constituent is, for the present, of a conjectural nature and that no evidence is offered that either of the steels giving the curves of Fig.

* R. F. Mehl, "Symposium on Hardenability of Alloy Steels," pp. 1–54 ("The Physics of Hardenability"). Cleveland, Ohio, U.S.A., 1938: American Society for Metals.

† A. U. Huddle, *Journal of The Iron and Steel Institute*, 1943, No. II., p. 572 P.

4 did, in fact, contain a constituent not present in the other, it is not clear why the authors select the left-hand curve as being the one for the steel containing the addition, particularly as this selection leaves them with the difficulty of supposing that the additional constituent acts in a manner opposite to that of most additional elements in steels.

On p. 144 P, speaking of the extra peak found in the S-curves of some steels, the authors cite several steels which give such peaks and point out that the alloy elements in all the steels cited form distinct second carbides; they consider that this gives good reason for the view that the peak and the presence of a second carbide are interconnected. In the "Atlas of Isothermal Transformation Diagrams" compiled by the United States Steel Corporation there are curves showing that silicon produces an additional peak on the S-curve very similar to that produced by the carbide-forming elements; this fact throws some doubt on the connection between the additional peak and the presence of a second carbide.

In the section of the paper dealing with the effect of grain-size on the transformation rate the authors take the view that fine grain-size is likely to promote breakdown of austenite in the range above that in which martensite is formed, but they point out that it is difficult to verify this view experimentally, because there are factors other than grain-size which could produce the effect and cause confusion in interpreting the results. A case in which these disturbing factors are reduced to a minimum is that of a sample heated to the stage where a mixed grain-size is produced, very large grains and small grains being side by side; examination of such a sample at a suitable stage of transformation above the martensitic range will often show the fine-grained areas completely transformed, whilst transformation appears not to have commenced in the coarse grains immediately adjoining.

It is not clear whether the authors intend their theory of the induction period as a period of stress relaxation to apply not only to the martensitic range but also to transformation at higher temperatures, where induction periods of appreciable length can be observed, though the stresses would be expected to be lower and their rate of relaxation much greater. If the theory is to apply to all temperatures then it would be expected that at all temperatures a surface layer of transformation product analogous to that found by the authors in the martensitic range would be produced; it would be interesting to know whether the authors have observed any such layer in specimens transformed at temperatures above the martensitic range. The case quoted above of a specimen with duplex grain-size being completely transformed in the fine-grained areas whilst there is no appearance of transformation commencing in adjacent coarse grains does not appear to accord well with the theory of stress relaxation as a cause of the induction period.

AUTHORS' REPLY.

The AUTHORS wrote in reply: Two points of general interest have been raised by several contributors to the discussion, namely, (1) the existence of tetragonal and cubic martensite, and (2) the inhibitory effect of compressive stress on the austenite transformation, and it may be well to deal with these first.

It has been known for twenty years at least that when a fully hardened steel is tempered, there is an abnormality not much above 100° C. The

dilatometric experiments of Andrew * and his co-workers may be cited as typical, and the X-ray work of Honda and Nishiyama later provided an adequate explanation of this effect. More modern investigation of this subject has yielded results of a somewhat negative character, and does not as yet, in our view, disprove that of the Japanese workers. At any rate, it is evident that the position is by no means fully cleared up. From the point of view of the present work, whether the two forms of martensite belong to the tetragonal and cubic systems is more or less immaterial; what we are concerned with is the actual existence of martensite in two forms, one of which is produced by drastic quenching and changes into the second form on mild tempering, the second form being also capable of production direct from austenite by isothermal treatment. While on this subject, we may assert our complete disbelief in the theory that martensite is formed only during cooling, such a view being in absolute contradiction with our own results. In these, a steel when satisfactorily quenched and held isothermally at an intermediate temperature, ultimately begins to transform, *i.e.*, to indicate a dilatometric expansion while in the metal bath, a process which can then be retarded by raising the bath temperature, which might be expected to have exactly the opposite effect.

With regard to the criticisms of the suggested explanation of the induction period as resulting from the setting up of inhibitory stresses on the austenite, the theory goes back to 1908, when it was first propounded by Benedicks, and has not yet, so far as we know, been seriously challenged. The first formation of martensite which occurs in the skin of the sample when it has cooled sufficiently takes place without interference. Once this has occurred, however, there is a relatively rigid surface layer which imposes a restraint on any further expansion, and, insofar as expansion is inhibited, transformation of austenite is retarded, or, if the stress generated is sufficiently high, prevented entirely. For experimental proof of the hypothesis we would suggest that Benedicks' original work † would well repay reading.

Concerning transformation at higher temperatures, we would refer Mr. Hopkins to the fact that Tammann's crystallization-velocity curve is still the basis for all such reactions; the stress effects are simply superposed on this.

As regards the question of abnormality, the whole story has not yet been told, and, for anything we know at present, nitrogen may play a part. At the same time—and one of us has worked on the subject at various times during the past eighteen years—the evidence as a whole would suggest very strongly that it is oxygen, directly or indirectly, which is responsible.

In reply to Mr. Burns, the evidence that an abnormal steel has a higher transformation rate than the normal seems to us to be conclusive.

In replying to several contributors who raise the question of the effect of grain-size, all that needs to be said is that our results have clearly shown that, *so far as the change to martensite is concerned*, this factor is without any observable influence.

Mr. Bucknall's discussion would appear to centre round the question as to whether a state of stable equilibrium can be attained at a given temperature between unchanged austenite and martensite. In plain carbon steels our evidence supports the view that, once conditions for the formation of martensite have been established, the reaction will proceed to completion if sufficient time is allowed at that temperature. Since the alloy steel with which Mr. Bucknall is concerned will, for a given carbon

* J. H. Andrew, J. E. Rippon, C. P. Miller, and A. Wragg, *Journal of The Iron and Steel Institute*, 1920, No. I., p. 527.

† C. Benedicks, *Journal of The Iron and Steel Institute*, 1908, No. II., p. 153.

content, be much more sluggish, it is easy to see why a state of pseudo-equilibrium might be reached, owing to the time factor being inadequate for further observable breakdown.

Finally, may we again emphasize that this work does not claim to be more than provocative, and is put forward at the present stage in the hope that it might stimulate further discussion and work on the subject.

A STUDY OF WORK-HARDENING AND REANNEALING OF IRON.*

By M. BALICKI, Dr.-Ing. (Met.) (University College, Swansea).

Synopsis.

The paper is an account of an extensive investigation into the changes in various properties of Armco-iron wires resulting from twelve different degrees of work-hardening, and subsequent annealing in a vacuum. Particular care was taken to secure comparable sets of results. On the same samples, or on exactly similar material, the following properties were determined : Hardness, electrical resistivity, elastic limit from the bend-test, plastic limit, ultimate stress, uniform elongation, springiness, thermo-electric properties and microstructure. Attention was paid to the effect of the time of annealing and to the order of the inaccuracy caused by ageing. Some other effects of which account should be taken are also indicated in the paper. A survey of the changes induced by reannealing confirms the presence of three phenomena : Strain-ageing, recrystallisation and crystal growth.

The main change, that of recrystallisation, of a work-hardened wire is shown by all the curves as occurring in the same range of temperatures (which is contrary to the findings of Tammann); this suggests that all the changes are a result of the same fundamental cause. The recrystallisation range is lowered when the degree of work-hardening is increased.

From the change in the course of the curves which characterise work-hardened iron three regions of plasticity were distinguished.

A short description of the working hypothesis of work-hardening is given and also an account of the theory of recrystallisation developed by Krupkowski and the author. The applicability of the theory to the experimental results is demonstrated by recalculation of some of the experimental curves. The calculated and experimental values show satisfactory agreement, although not as good as was previously obtained when copper and nickel were similarly treated. Evidence is produced that irregularities are caused by strain-ageing. A point of theoretical importance is that the energy of activation of the recrystallisation process of iron is independent of the degree of reduction and is nearly the same (31,000 cal. per mol) as previously determined for copper (33,000 cal. per mol) and nickel (32,400 cal. per mol). The frequency coefficient c is found to depend upon the degree of cold-work in a similar manner to that in which certain properties increase with increasing degrees of reduction. Thus the value of c depicts the nature and the history of the metal.

Introduction.

A theory of recrystallisation which was put forward by A. Krupowski and the present author in 1937 [1] is regarded by them as applicable to all metals and alloys. This theory was deduced from an analysis of the process of reannealing of cold-worked copper [2] and yielded a good result. It was also confirmed by similar investigations carried out by the present author on nickel.[3] Subsequently it was considered important to extend the investigations to iron with the object of deciding whether or not this important metal reveals in detail the same features on annealing as the other two metals.

There is already an extensive collection of facts concerning the cold-working and reannealing of iron.[4, 5, 6, 7] These data, however, having been determined under widely varying experimental conditions, were not

* A report, received April 13, 1944, on a research carried out with the aid of a grant from the Andrew Carnegie Research Fund.

quite suitable for the author's method of approach to the problem. A systematic experimental study was therefore made of the changes in many properties with the object of obtaining a more comparable set of data. This was of importance for the subsequent theoretical work, which aimed at explaining quantitatively the changes caused by reannealing.

Material Used and Its Preparation for Testing.

Although extremely pure iron is desirable for such work, it was, unfortunately, not available in sufficient quantity and, therefore, the work was carried out on Armco iron, which is sometimes regarded as a standard ingot iron.

It was considered that if the theory of recrystallisation could be proved to be applicable to material of this composition the result would be of even greater practical value than if pure iron were used.

The material was of the following analysis:

C. %.	Mn. %.	Si. %.	P. %.	S. %.	Cu. %.
0·013	0·26	0·001	0·008	0·02	0·049

As in the author's previous investigation on copper and nickel, wire drawing was used as a means of deformation because of its many advantages

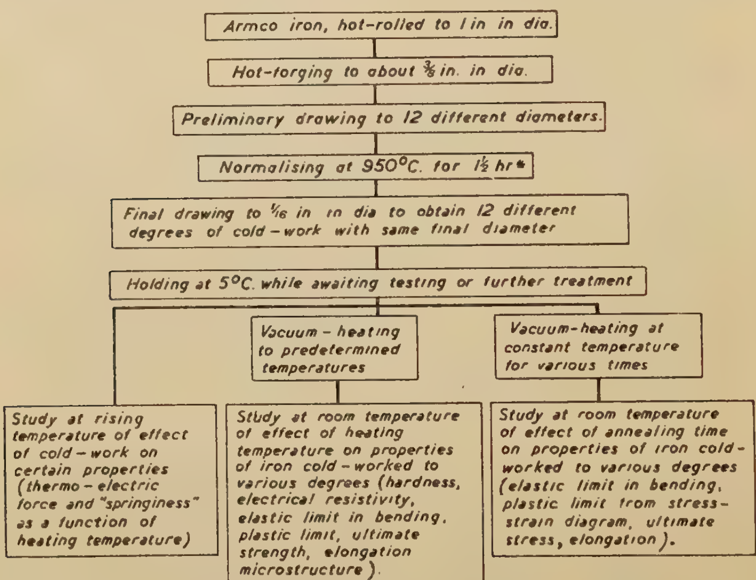

Fig. 1.—Scheme for the Preparation of Material for Testing.

The iron available was in the form of a hot-rolled bar 1 in. in dia. This diameter being too large for immediate drawing on the available draw-bench, the bar was reduced by hot forging and grinding to about ⅝ in. in dia., after which it was drawn down through a suitable set of steel dies. The sequence of these preparatory operations is clearly shown in Fig. 1, which does not need any special comments.

* *See* Balicki.[71]

Effect of Annealing Temperature on the Properties.

Method of Heating the Specimens.

In any process of annealing, time and temperature act simultaneously and this renders the study of the kinematics of reannealing somewhat complex. The method used aimed at determining the effect of temperature with the least possible influence of time. It consisted of heating the specimens in a vacuum at a standardised rate to the predetermined temperatures and immediately cooling them.

The samples, 13 cm. long and 1·59 mm. in dia., representing all degrees of cold-work, were fixed around a silica tube of small diameter. The wires were surrounded with iron wool which had been well annealed in a vacuum at 700° C. The wires were then placed in a cylindrical silica container (Fig. 2) and heated in a vacuum (0·01 mm. Hg) in a long electric furnace. When the temperature approached the intended level the container was removed from the furnace. Subsequent observations revealed the exact

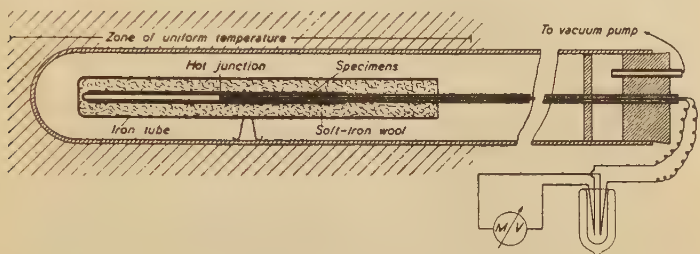

Fig. 2.—Apparatus for Annealing *in vacuo*.

maximum temperature of the samples and the short time during which this maximum was maintained, the latter being on the average 1 min. The samples were heated through the 100–650° C. range at the practically constant rate of 16° C. per min.

Specimens annealed in this manner showed no trace of oxidation and, what is important, remained straight. In comparative tests it was shown that the iron-wool wrapping had no effect on the mechanical properties of the wires.

The arrangement for annealing described above was found convenient, and was used whenever annealing was carried out.

Hardness Testing.

For hardness tests, lengths of about ½ in. were cut off the ends of the specimens. These were mounted in mild-steel plates by fitting them mechanically in grooves of exactly the same radius as that of the wires. The wires were thus kept parallel to the bottom of the plate, which facilitated polishing to the same depth. After polishing, the Vickers hardness was determined with a load of 5 kg. at four points along the axis of each wire. The mean values obtained are given in Table I. and served for the construction of the space diagram shown in Fig. 3.

The present systematic determinations of the change in hardness due to cold-working and subsequent reannealing are in good general agreement with the results of Broniewski and Krol,[8] Broniewski and Robowski,[9] O'Neil,[10] Goerens,[11] Edwards and Kuwada,[12] Thomas,[13] Swinden and Bolsover,[14] Edwards, Phillips and Pipe,[15] Sykes and Ellsworth,[16] Davenport and Bain [17] and others.

TABLE I.—*Vickers Hardness of Armco-Iron Wire after Cold-Working and Annealing.*

Annealing Tempera-ture. ° C.	Vickers D.P.H. Number after Cold-Working.											
	0%.	5·6%.	14·0%.	19·2%.	29·5%.	40·7%.	46·5%.	57·2%.	67·9%.	75·9%.	84·2%.	93·8%.
20	80	116	143	151	152	171	173	179	182	193	199	222
143	80	117	144	153	161	175	180	182	194	196	203	220
201	81·5	115	145	153	160	177	186	186	196	200	203	218
260	80·2	117	145	158	165	178	184	188	198	198	204	215
294	81	119	146	152	163	179	181	190	199	196	205	215
320	79·8	120	143	154	166	177	181	189	191	198	200	214
350	79·7	116	138	152	162	179	178	186	191	194	203	210
387	80·5	117	139	148	160	176	178	184	187	194	198	209
419	82	117	132	148	161	168	173	183	185	188	200	205
450	80·5	114	129	147	157	168	174	176	182	185	200	199
475	80	112	128	142	156	163	170	172	180	182	195	189
494	81·6	114	124	135	151	159	166	170	170	180	189	187
525	82·5	112	125	135	147	154	161	169	169	177	183	180
549	81·5	111	121	130	145	154	159	159	166	153	161	146
575	81·2	...	115	125	143	143	143	139	150	123	120	118
590	...	110	118	123	136	133	128	126	116	99	100	105
607	81·8	103	112	118	130	118	115	103	100	98	101	108
630	...	102	115	113	122	112	98	90	95	92	99	105
650	83·5	102	107	112	114	99	95	93	92	93	99	105
675	...	100	104	105	109	93	93	88	93	88	96	101
701	...	102	102	106	100	90	93	86	89	91	95	97
729	...	100	99	104	96	88	85	89	89	92	95	97
749	85·6	94·5	94	96	93	86	87	87	90	90	95	98
850	...	97	97	93	94	85	81	84	86	87	90	84
950	63·3	66	74	73	69	66	72	70	77	76	80	74

Summary of Results of Hardness Tests.

(1) Practically no change in hardness is found in fully annealed iron that has not been cold-worked, even after heating up to 750° C.

(2) The deviation of the points from the average curve is small for this non-cold-worked iron (± 3 D.P.H. units) and for low degrees of cold-work except where there is a sudden drop in hardness. This shows that all secondary factors affecting the accuracy were kept under adequate control.

(3) The hardness increases as the degree of cold-work increases in accordance with the well-known curve.

(4) All the wires cold-worked up to 70% reveal a more or less distinct

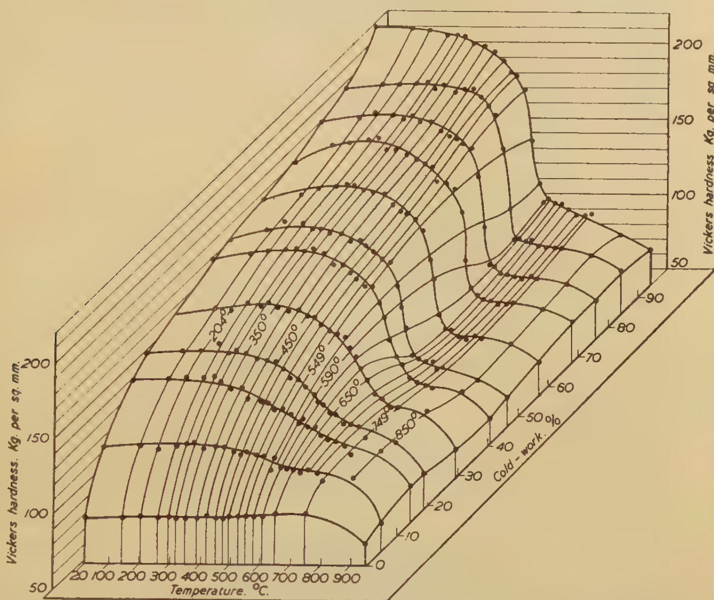

FIG. 3.—Vickers Hardness of Armco Iron Wire in relation to cold-work and annealing temperature.

increase in hardness after heating to moderate temperatures; this is ascribed to strain-ageing. Wires with a higher degree of cold-work than 70% do not reveal any increase of hardness.

(5) Heating to successively higher temperatures causes a steady diminution of hardness in slightly cold-worked wires, whilst, for the medium and heavily cold-worked series, a sudden drop occurs.

(6) The higher the degree of cold-work the lower is the temperature at which this main drop in hardness takes place.

(7) At still higher temperatures of heating (up to 850° C.) for all degrees of cold-work the hardness is not much affected and the curves are more or less flat.

(8) Heating the wires to 950° C. causes a second drop in the hardness.

(9) The hardness of iron annealed at the higher temperatures is influenced, but not to a great degree, by mechanical treatment prior to the annealing. In other words, after annealing, specimens previously subjected

TABLE II.—*Changes in the Electrical Resistivity at 20° C. of Armco-Iron Wire with Increasing Cold-Work and Annealing Temperature.*

Annealing Temperature. ° C.	Specific Electrical Resistivity at 20° C. in Microhms-cm. after Cold-Working.											
	0%.	5·6%.	14·0%.	19·2%.	29·5%.	40·7%.	46·5%.	57·2%.	67·9%.	75·9%.	84·2%.	93·8%.
20	11·20	11·17	11·15	11·19	11·24	11·23	11·24	11·27	11·30	11·27	11·30	11·32
143	11·21	11·16	11·16	11·19	11·23	11·23	11·26	11·26	11·29	11·27	11·30	11·33
201	11·20	11·16	11·17	11·19	11·24	11·20	11·22	11·24	11·28	11·26	11·29	11·30
260	11·20	11·17	11·16	11·20	11·24	11·19	11·23	11·26	11·26	11·29	11·26	11·29
294	11·19	11·15	11·15	11·16	11·23	11·18	11·22	11·26	11·26	11·29	11·27	11·26
320	11·18	11·14	11·15	11·16	11·21	11·16	11·17	11·24	11·27	11·25	11·25	11·31
350	11·19	11·17	11·13	11·17	11·19	11·15	11·21	11·26	11·27	11·22	11·24	11·23
387	11·20	11·18	11·15	11·16	11·19	11·19	11·19	11·21	11·22	11·25	11·23	11·23
419	11·19	11·16	11·13	11·17	11·20	11·17	11·15	11·23	11·22	11·22	11·18	11·20
450	11·21	11·14	11·14	11·16	11·20	11·15	11·17	11·20	11·20	11·18	11·22	11·19
475	11·19	11·15	11·15	11·16	11·19	11·17	11·15	11·17	11·21	11·18	11·19	11·19
494	11·21	11·16	11·15	11·14	11·18	11·14	11·13	11·18	11·18	11·16	11·16	11·17
525	11·20	11·15	11·15	11·13	11·15	11·12	11·13	11·16	11·15	11·14	11·14	11·15
549	11·20	11·14	11·15	11·16	11·13	11·12	11·14	11·14	11·16	11·15	11·15	11·16
578	11·21	11·15	11·13	11·15	11·15	11·15	11·15	11·17	11·15	11·14	11·16	11·16
607	11·22	11·13	11·14	11·14	11·17	11·17	11·15	11·15	11·12	11·15	11·17	11·16
650	11·21	11·17	11·14	11·19	11·18	11·19	11·19	11·20	11·17	11·17	11·18	11·16
749	11·20	11·19	11·18	11·20	11·19	11·21	11·19	11·18	11·19	11·21	11·26	11·26
950	11·20	11·20	11·20	11·18	11·20	11·19	11·18	11·19	11·22	11·24	11·25	11·25

to low and high degrees of cold-work have slightly higher hardness values than medium cold-worked specimens. The minimum hardness is found in wires subjected to about 50% cold-work.

Electrical Resistivity.

Electrical resistivity determinations were carried out by the compensation method. The annealed samples were kept at a constant temperature in a stirred paraffin oil bath. To prevent heating of the specimen by the testing currents, the current density in the sample was kept at 80 milliamp. per sq. mm. of section. As an additional precaution the current from a large-capacity battery was passed through a standard resistance of 0·02 ohm and through the specimen during the actual measurement only, and was by-passed at other times through an equivalent circuit. The changes in temperature of the bath during the actual tests were ± 0·01° C. The accuracy of the determination of the potential was ± 1μV., while that of the determination of length was ± 0·01 mm. Although the wires were

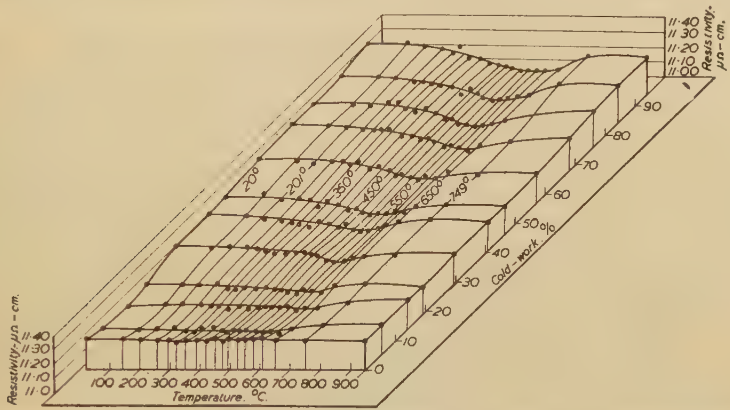

Fig. 4.—Specific Electrical Resistivity of Armco-Iron Wire at 20° C. in relation to cold-work and annealing temperature.

finished in the same die, the diameter varied slighly and was carefully checked at several places, because of the known effect which variations in section have on the accuracy of electrical resistivity determinations.

All the determinations were carried out at 20° ± 0·5° C.

From the data obtained the resistances at the actual temperature were determined and subsequently calculated for 20° C. Because the deviation from the temperature of 20° C. in every determination was small, it was sufficiently accurate to use the linear formula for extrapolation to 20° C., and to assume that the temperature coefficient was independent of the amount of cold-work and annealing.

The results of tests and calculations, both repeated twice on the same specimens, are to be found in Table II. Fig. 4 was constructed from these values.

Summary of Results of Resistivity Tests.

(1) The specific electrical resistivity at 20° C. of normalised Armco iron of the above composition is found to be 11·20 microhms-cm.; this value does not vary with subsequent short-time annealing.

(2) Work-hardening affects this property, but not to such a degree as it

does the hardness. The maximum increase in resistivity was observed on wires with 93·8% of cold-work, and is found to be only 0·9%. Small amounts of cold-work first lower the electrical resistivity slightly; an increase then begins at about 30% of cold-work.

(3) Annealing a cold-worked wire lowers its resistivity. This decrease for higher degrees of cold-work is noticeable at as low a temperature as 100° C. Moderately cold-worked wires reveal this beginning of the drop at slightly higher annealing temperatures.

(4) The diminution in resistivity continues with increasing annealing temperature up to 550–650° C. The minimum value of specific electrical resistance lies somewhere in this region.

(5) This minimum value is dependent upon the degree of cold-work imposed upon the wires before annealing. The lowest values are obtained for wires with 40·7% and 46·5% of cold-work.

(6) Still higher annealing temperatures bring about an increase in the specific electrical resistivity. Here, too, the wires with medium degrees of cold-work are not so much affected as wires which were either slightly or heavily deformed before annealing.

(7) The shifting of the temperature range of annealing with varying degrees of cold-work is not distinctly revealed by these determinations. The main reasons for this are the very small order of the changes which occur in this property and insufficient accuracy of the determinations.

The results obtained in the present work corroborate those of the following investigators: Credner,[18] Goerens[11] (Fig. 21), Uêda,[19] Tâkahasi,[27] Broniewski and Glotz,[29] Köster and Tiemann,[30] Bardenhauer and Schmidt[31] and others.[4, 28] The much quoted[4, 21, 50] paper by Tammann and Moritz[20] reveals a rather lower range of recrystallisation change than those observed in the present work and by the other investigators. This, however, might be due to the use of an active atmosphere of hydrogen during annealing,[22–27] to the accumulation of errors connected with the measurement of electrical resistivity at elevated temperatures (including the method of reduction of the values to normal conditions) and finally, to the fact that the same wire was used for successive determinations at all temperatures of testing. These differences do not permit of a quantitative comparison with the results of the present author or of other investigators quoted above, nor with other cases.[21, 50]

Determination of Elastic Properties by the Bend Test.

The use of the bend-test for the determination of the elastic properties of metals has been developed by a number of workers.[2, 32–35, 38, 70] It should be noted, however, that with the exception of the work of Dreyer and Tammann[51] the literature seldom refers to bend tests on iron in the cold-worked and annealed states.

Fig. 5 shows the arrangement for the bend test. The samples, of the same length and diameter previously used for electrical resistivity determinations, were fixed at one end in a grip B held in a suitable jaw J and directly loaded at the other end. To obtain the best possible union between the grip and the wire, the former was made of two hard brass plates fitted with grooves with a radius of $\frac{1}{16}$ in. The other flattened end of the wire was notched to ensure the application of the load at the same initial distance L (70 mm.) from the fixed end. Changes in the position of the free end, after removal of the loads, were observed with an accuracy of $\pm$ 0·001 mm. by means of a travelling microscope. All samples were retested after reversing the ends.

It should be noted that the readings of the position of the free ends were made immediately after the sample was unloaded because there was a

slight tendency to recovery. As is well known, the cantilever bend test involves simultaneous tension of the upper fibres and compression of the lower fibres along the whole sample with maximum stresses at the support. If the maximum tensile or compressive stresses are lower than the elastic limit of the tested material there will be no plastic deformation, and after releasing the load the end of the specimen should regain its position. If

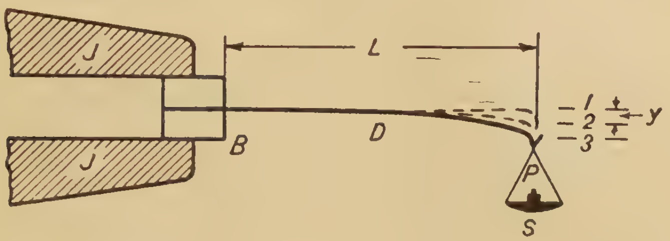

FIG. 5.—Arrangement for Bend Test.

the stresses exceed the elastic limit a permanent deflection of the appropriate outer fibres will be revealed by a permanent deflection of the end. In view of the results of Merrils [36] and Rees [37] on iron, both stretched and compressed sides should deform simultaneously.

For certain reasons it was decided to study the stress which produces a given standard deformation, not merely the commencement of the deformation.

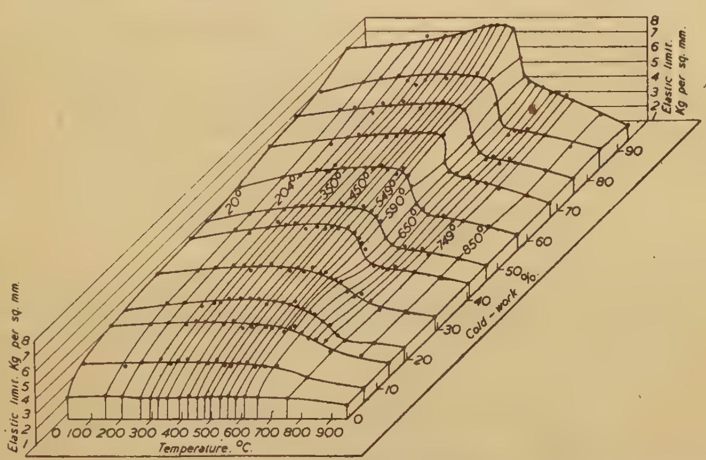

FIG. 6.—Elastic Limit (calculated from bend-test data) of Armco-Iron Wire in relation to cold-work and annealing temperature.

formation. A deflecticn of 0·07 mm. was adopted for this value. This set can be more strictly given as the angle of permanent deflection; it amounted to 5° 45′.

To evaluate the critical load, the load-deflection curves for all samples were drawn. In all cases the adopted standard deflection indicated the commencement of permanent set. From the values of these critical loads

TABLE III.—*Elastic Limit of Armco-Iron Wire in Relation to Cold-Work and Annealing Temperature.*

Annealing Tempera-ture. °C.	Elastic Limit in Kg. per Sq. Mm. after Cold-Working.											
	0%.	5·6%.	14·0%.	19·2%.	29·5%.	40·7%.	46·5%.	57·2%.	67·9%.	75·9%.	84·2%.	93·8%.
20	1·5	2·6	3·5	3·4	3·8	4·0	3·9	4·0	4·6	5·3	5·1	6·2
143	1·6	2·4	3·5	3·9	4·0	4·5	4·5	4·7	5·0	5·5	5·8	6·4
201	1·5	2·7	3·7	4·0	4·2	4·3	4·7	4·9	4·8	5·7	6·1	6·4
260	1·4	2·9	3·5	3·7	4·4	4·6	4·8	5·3	5·4	5·3	6·2	6·9
294	1·4	2·6	3·8	4·0	4·4	5·1	5·1	5·2	5·5	6·0	5·9	6·7
320	1·3	2·9	3·6	4·0	4·1	4·8	5·1	5·5	5·4	5·8	5·6	7·0
350	1·4	2·9	3·6	4·2	4·4	4·9	5·1	5·4	5·5	6·1	5·9	7·0
387	1·3	2·6	3·5	4·2	4·5	4·6	5·1	5·4	5·4	5·9	6·4	7·2
419	1·5	2·7	3·8	4·2	4·3	4·7	5·0	5·5	5·4	5·6	5·8	7·3
450	1·4	2·9	3·3	4·2	4·0	4·6	5·2	5·4	5·7	5·9	6·4	7·5
475	1·5	2·5	3·4	3·8	4·3	4·8	5·0	5·6	5·6	6·0	6·3	7·6
494	1·4	2·4	3·5	4·0	4·0	4·6	4·9	5·3	5·6	6·1	6·2	7·7
525	1·4	2·7	3·3	4·0	4·3	4·7	5·1	5·0	5·6	5·9	6·4	7·7
549	1·5	2·7	3·3	3·8	3·9	4·3	4·7	5·4	5·6	5·8	6·1	7·5
578	1·4	2·4	3·3	4·1	4·0	3·9	4·4	4·9	5·5	5·6	6·0	5·4
590	...	...	2·9	3·8	3·6	3·7	3·9	4·2	3·5	3·8	4·4	4·3
607	1·5	2·4	3·3	3·6	3·5	3·5	2·9	3·0	3·0	3·4	3·3	3·9
630	...	...	3·4	3·4	3·3	2·1	2·4	2·7	2·9	2·7	2·4	3·6
650	1·5	2·4	2·9	3·2	2·7	2·0	2·0	2·3	2·7	2·2	2·6	3·4
675	...	...	2·4	2·7	2·3	1·9	2·3	2·3	2·7	2·4	2·4	3·2
701	...	...	2·4	2·4	2·4	2·0	2·1	2·3	2·3	1·9	2·7	3·2
729	...	...	2·4	1·6	2·0	1·9	2·1	1·9	2·3	2·3	2·4	3·0
749	1·4	1·7	1·6	1·4	1·4	1·7	1·7	1·9	1·9	2·0	2·3	2·6
850	...	...	1·4	1·7	1·3	1·3	1·5	1·5	1·5	1·5	1·5	1·5
950	1·0	1·0	0·9	0·9	0·9	0·9	1·0	0·9	0·9	0·9	1·3	0·9

and from the weight of the projecting part of the samples the critical moment M was calculated and, together with the appropriate value of the section modulus W, was substituted in the known formula :

$$\sigma_b = M_b/W$$

to calculate the value of the elastic limit from the bend test, σ_b. Table III. and Fig. 6 contain the data obtained in this way.

Summary and Discussion of the Bend-Test Results.

(1) The curves of the load-deflection relationship, as is well known from the work of Meyer,[38] represent the first portion of the stress-strain and compression-strain diagrams.

(2) A load as small as 20 g. was, as a rule, sufficient to produce a clearly observable set in nearly all the wires tested. The corresponding stress, which is of the order of 1·5 kg. per sq. mm., is astonishingly low. It must be pointed out, however, that any imperfections of the surface or the presence of traces of any soft microstructural component, and also the diminution in length of the true arm of the moment during bending, might have a bearing on this result.

(3) Cold-worked wires annealed at moderate temperatures revealed this initial set, but subsequently resisted the action of the bending moment better than wires cold-worked but not annealed.

(4) Higher temperatures (above 550° C.) caused a further increase in the tendency to set which usually ended in the collapse of the wire. This sudden drop occurred in samples annealed at high temperatures and sometimes occurred after a short lapse in the time under load.

(5) The curves for elastic limit from the bend test as a function of the degree of cold-work had a similar trend to the corresponding curve for hardness.

(6) Disregarding the curve in Fig. 6 for non-cold-worked iron, all the other curves show an increase with the rise in annealing temperature, followed by a sudden drop at higher temperatures. In well annealed wires there is a further decrease in the elastic limit with increasing temperature.

(7) The temperature at which the sudden drop occurs depends upon the degree of cold-work in very much the same manner as in the case of hardness.

Tensile Tests.

Tensile tests were carried out with a 5-ton Amsler machine equipped with autographic recording. The lowest scale range was not very suitable and therefore a new one of 150 kg. was arranged and carefully calibrated. As the extensometers available were too heavy, it was considered advisable to dispense with them despite the risk that the diagrams might be affected by any slipping in the jaws. A distance gauge assured that a constant length (10 cm.) of sample was always secured between the jaws, and a constant rate of straining was applied.

Fig. 7 contains some of the curves recorded by the machine and these illustrate the effects which the cold-working and annealing have on the shape of the strain-stress diagrams.

As it was not possible under the given conditions of testing to determine either of the recognised elastic limits (including proof stress), the determinations of the "plastic limit" were carried out to secure a numerical indication of the stress which characterised the change from the elastic to the plastic region. This was calculated from the values of the loads at which the pointer of the dynamometer slowed, stopped or dropped with a constant rate of strain. As such a load depicts the commencement of

spontaneous ductility and not the end of the elastic region in the cases where there is an upper and lower yield point, the lower yield-point load was taken for the calculation. The data are presented in Table IV. and their dependence upon the degree of cold-work and annealing temperature is shown in Fig. 8.

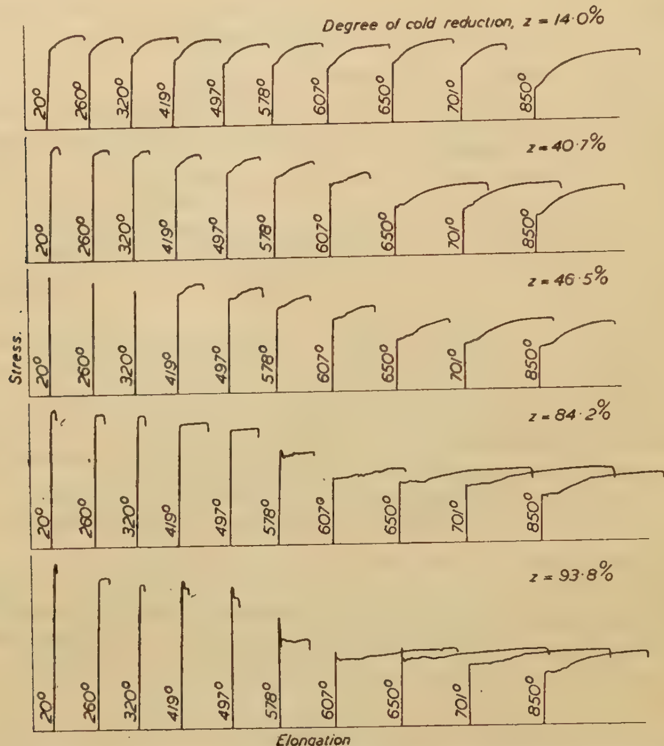

FIG. 7.—Stress-Strain Diagrams for Armco-Iron Wire after different degrees of cold-work (z) and annealing at the temperatures indicated.

Summary of Results Relating to the Transition from the Elastic to the Plastic Stage.

(1) The transition from the elastic to the plastic region of the specimen differed greatly according to mechanical and thermal pretreatments and it can be classified as follows:

(*a*) Specimens giving a normal stress-strain diagram with a normal yield point. This type is common to wires normalised and annealed at above 675° C. Annealing at low temperatures restores this type of curve in slightly cold-worked samples.

(*b*) Specimens with smooth curves without any " step " before the ascending branch were met with when testing samples representing low degrees of cold-work, whether annealed at moderate temperature or not annealed.

(*c*) Specimens practically without a plastic region, *i.e.*, those for

which the plastic limit approaches the ultimate stress. These are the specimens with more than 46·5% of cold-work even after low-temperature annealing.

(*d*) Specimens with medium degrees of cold-work when heated to moderate temperatures and those of high degrees heated at low temperatures frequently have a type of stress-strain diagram the plastic region of which consists of elastic and yield-elongation portions only, the

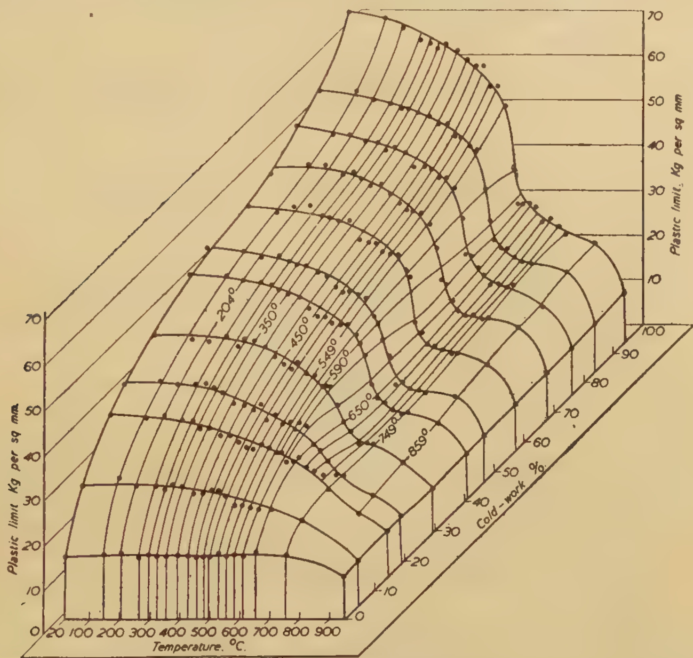

FIG. 8.—" Plastic Limit " of Armco-Iron Wire in relation to cold-work and annealing temperature.

ascending portion being absent. The characteristic yielding features were present during the tests ; very fine stretcher lines could be observed moving along the wires in exactly the same manner as in the cases of the usual or pronounced yield point and the whole elongation took place at a constant stress.

(*e*) The last group of specimens revealed the stress-strain characteristics which can be described as having exaggerated yield points, for they showed great discrepancies between the upper and lower yield stresses and extensive yield-elongation. In some cases the upper yield stress exceeded the stress at which the reduction of area began (*see* Fig. 7). In other cases two or three stages of yielding were present. The samples behaving in this manner were those of higher degrees of cold-work than 70%, annealed at temperatures between 480° and 600° C.

(2) The plastic limit gives a good, though approximate, numerical measure of the transition from the elastic to the plastic range of the material.

TABLE IV.—"*Plastic Limit*" *of Armco-Iron Wire in Relation to Cold-Work and Annealing Temperature.*

Annealing Temperature. ° C.	"Plastic Limit" in Kg. per Sq. Mm. after Cold-Working.											
	0%.	5·6%.	14·0%.	19·2%.	29·5%.	40·7%.	46·5%.	57·2%.	67·9%.	75·9%.	84·2%.	93·8%.
20	14·0	26·3	36·3	40·1	46·0	50·2	52·2	54·7	55·0	59·5	62·3	73·6
143	14·8	28·2	35·5	40·5	43·9	49·7	51·0	50·8	56·7	58·2	62·3	72·2
201	15·1	26·2	36·0	39·8	42·5	48·7	50·3	54·1	56·4	58·3	60·5	68·8
260	14·0	26·3	35·1	39·7	44·0	48·3	50·5	54·1	54·5	56·9	59·6	67·5
294	14·7	25·4	34·8	39·7	41·5	47·9	49·2	52·1	55·0	56·4	58·7	66·6
320	14·2	26·6	33·0	39·0	43·2	46·7	48·4	51·7	53·5	54·5	58·5	65·5
350	14·2	26·2	33·3	38·0	42·8	45·8	48·3	50·9	52·1	55·3	57·8	66·5
387	14·5	24·6	31·6	35·5	41·6	44·9	47·0	51·0	52·2	54·2	56·8	65·2
419	14·3	24·7	31·4	34·9	39·3	42·5	45·8	46·7	49·8	50·8	55·1	63·0
450	14·0	25·3	29·9	33·8	37·6	40·8	43·5	46·2	47·8	51·0	55·3	61·5
475	14·2	25·2	28·9	35·6	36·4	40·7	41·9	46·0	46·5	50·2	52·2	61·7
494	14·2	24·0	28·5	31·4	38·7	39·3	42·4	44·1	46·2	48·6	51·9	60·9
525	14·9	21·8	29·4	33·0	35·7	38·9	41·8	43·8	45·8	48·7	50·2	58·9
549	14·2	21·8	28·8	30·6	35·1	38·9	41·1	43·5	43·3	47·0	49·8	54·7
578	14·7	21·3	27·7	31·7	32·5	36·6	36·9	40·2	35·3	40·0	41·4	43·1
590	...	...	27·8	29·5	32·5	34·7	35·4	38·5	31·2	31·3	34·3	35·0
607	14·3	20·9	25·4	31·3	32·5	33·4	32·8	28·3	30·0	30·7	29·2	30·8
630	...	...	25·7	30·5	28·3	25·5	28·7	26·1	26·5	28·3	27·3	30·9
650	15·0	21·1	26·8	26·3	24·5	22·5	23·0	23·5	24·0	26·7	26·9	29·9
675	...	...	24·3	25·1	22·6	22·3	23·5	23·0	23·1	24·4	25·2	26·8
701	...	...	23·5	22·6	19·8	19·8	21·0	22·2	23·2	24·7	24·6	27·3
729	...	...	22·5	19·8	20·1	18·8	20·5	21·3	22·3	23·5	24·0	25·7
749	14·5	18·5	19·3	19·3	19·8	19·3	19·8	21·2	22·7	24·5	24·8	24·2
850	...	...	14·1	15·1	15·5	16·7	19·1	18·8	20·3	19·7	19·8	22·0
950	9·7	9·7	10·3	10·8	9·8	9·2	10·5	10·0	10·2	10·3	10·7	10·9

(3) This limit for normalised wires is 14·5 kg. per sq. mm., and it does not change with subsequent annealing below 650° C.

(4) Raising the annealing temperature reduced the plastic limit in all cases.

(5) The plastic limit increases with the degree of cold-working in a manner similar to the hardness. The first portion of the curve, however, rises more rapidly.

(6) Iron cold-worked to less than 75·9% reveals an increase in the plastic limit after low-temperature annealing. This initial increase is followed by

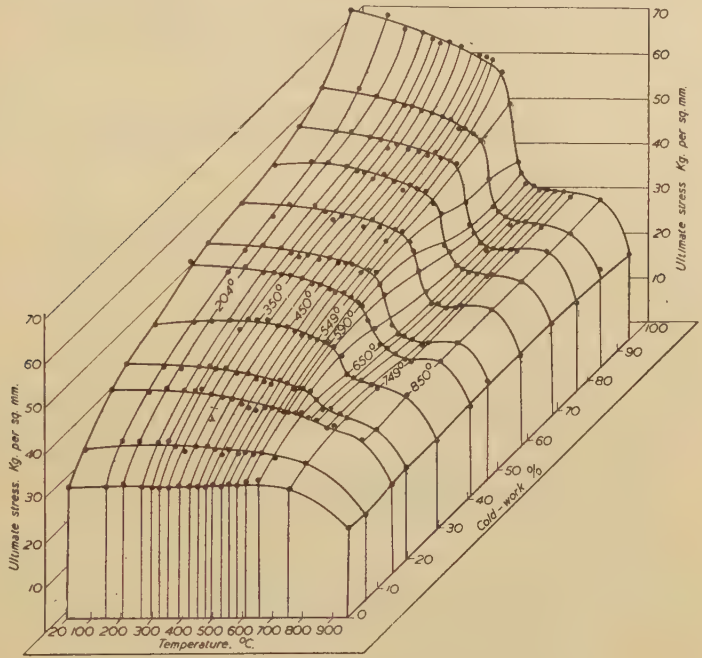

FIG. 9.—Ultimate Stress of Armco-Iron Wire in relation to cold-work and annealing temperature.

a decrease, slight at the beginning but becoming a sudden drop somewhere between 590° and 650° C. In wires more heavily worked no increase is noted and the drop begins from the lowest temperatures of annealing.

(7) The commencement and the main drop depend upon the degree of cold-work. The dependence is similar to that met with in the hardness determinations.

Ultimate Stress.

The ultimate stress was calculated as usual from the maximum load registered by the dynamometer during the test. The values are given in Table V. and are graphically represented in Fig. 9. The relationships are self-explanatory and it is unnecessary to give a summary which would not differ from that for the hardness and plastic limit.

TABLE V.—*Ultimate Stress of Armco-Iron Wire in Relation to Cold-Work and Annealing Temperature.*

Annealing Temperature. °C.	Ultimate Stress in Kg. per Sq. Mm after Cold-Working.											
	0%.	5·6%.	14·0%.	19·2%.	29·2%.	40·7%.	46·5%.	57·6%.	67·9%.	75·9%.	84·2%.	93·8%.
20	29·4	34·2	41·7	44·1	47·4	51·9	53·0	54·7	55·1	59·5	62·3	73·6
143	29·4	36·2	41·0	43·3	45·9	50·3	51·3	52·8	57·0	58·2	62·3	72·2
201	30·2	35·9	42·0	42·2	46·2	51·5	52·4	54·2	56·7	58·3	60·5	69·1
260	29·4	35·5	41·3	43·3	46·5	50·2	51·8	54·1	55·6	56·9	59·6	68·2
294	29·4	36·3	41·3	43·5	43·2	50·3	50·7	52·8	55·8	56·4	58·7	67·0
320	29·4	34·7	41·5	43·3	46·5	49·7	49·4	51·4	54·2	54·5	58·5	66·2
350	29·4	33·4	40·6	43·1	46·4	49·5	50·4	51·7	53·5	55·3	57·9	66·7
387	29·4	34·2	39·7	41·9	46·5	49·0	49·2	51·7	54·2	54·2	56·9	65·5
419	29·4	34·9	39·4	41·7	45·0	47·5	49·7	49·3	52·5	53·8	56·0	66·2
450	29·5	34·2	38·3	40·4	45·0	47·5	48·5	51·0	50·6	53·3	55·5	64·2
475	29·5	32·5	37·6	40·0	43·8	46·4	48·5	49·4	51·9	53·8	53·5	63·0
494	30·0	34·2	37·0	39·3	42·9	45·8	48·5	46·2	51·0	52·5	53·8	62·8
525	29·5	33·2	37·4	40·6	42·4	45·7	46·5	48·8	50·3	51·8	52·2	59·8
549	30·3	33·5	37·4	39·2	42·3	44·8	47·1	48·3	47·5	50·2	50·8	52·6
578	30·1	32·9	37·0	38·5	41·3	43·5	46·4	44·5	45·2	42·3	42·2	39·8
590	...	...	36·3	38·6	41·5	42·8	43·8	45·8	38·2	37·1	36·0	37·4
607	30·7	33·5	36·2	38·6	40·4	39·6	39·8	37·0	36·7	35·7	35·1	35·1
630	...	...	36·5	38·0	38·5	38·0	33·2	35·1	33·8	33·7	33·9	34·7
650	31·4	32·2	36·5	36·0	34·3	35·4	32·9	32·3	33·0	31·8	31·9	33·8
675	...	...	35·1	33·2	33·6	33·3	32·2	32·3	32·2	32·2	32·7	33·8
701	...	...	34·8	34·1	33·0	32·1	31·4	31·2	31·8	32·0	32·8	33·2
729	...	...	31·1	33·0	32·2	31·0	30·3	31·5	31·3	32·2	32·2	33·3
749	29·2	31·2	32·2	32·2	31·7	30·6	31·1	31·8	31·8	31·6	31·6	31·7
850	...	...	30·7	29·5	29·8	30·6	31·0	31·0	30·2	31·8	30·0	31·3
950	20·6	19·5	20·5	21·0	19·8	20·6	22·5	20·4	21·0	20·2	22·3	19·3

Elongation.

Apart from the information about ductility from the stress-strain diagrams, a numerical indication of the change in this property was secured despite the unfavourable conditions rendering determinations of the total elongation difficult. The uniform elongation was measured from the change in diameter and interpreted with the help of an analysis of the correlation between the total and uniform elongations carried out by Krupkowski.[39]

The diameters of the fractured wires were measured at a distance five times the initial diameter d_0 from the fracture and the mean diameter substituted for d in the formula :

$$a = \left[\left(\frac{d_0}{d}\right)^2 - 1\right] \times 100.$$

The values obtained for the uniform elongation a are shown in Table VI. and are graphically represented in Fig. 10.

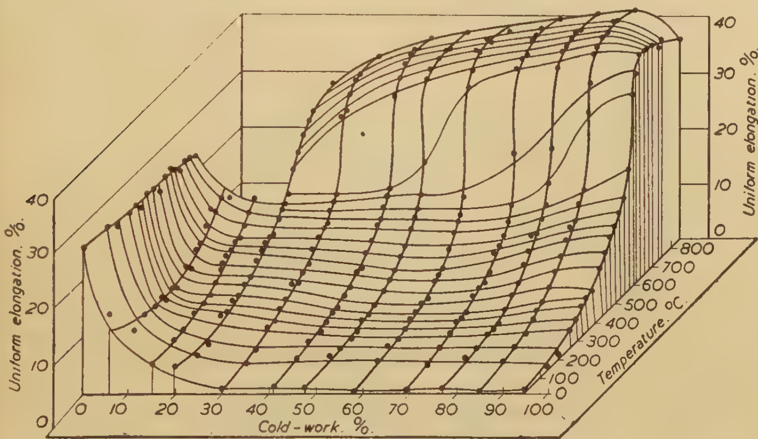

FIG. 10.—Elongation of Armco-Iron Wire in relation to cold-work and annealing temperature.

Summary of Results with Regard to Elongation.

(1) The uniform elongation produced by fracture of the normalised specimens is about 25% and is not affected by subsequent annealing at temperatures below 600° C. Annealing at higher temperatures causes the specimens to have a very coarse surface after fracture and this makes the determinations uncertain. These values have therefore been omitted. The same applies to specimens with lower degrees of cold-work annealed at high temperatures.

(2) With an increase of cold-work up to 30% there is initially a very rapid drop in the uniform elongation. The elongation of the more heavily cold-worked wires is negligible.

(3) Low-temperature annealing of slightly and moderately cold-worked wires does not seem to affect the elongation, whilst those with more than 30% of cold-work show a noticeable increase in ductility after a similar treatment. The higher the degree of cold-work and temperature, the greater is the increase in elongation.

TABLE VI.—*Uniform Elongation of Armco-Iron Wire in Relation to Cold-Work and Annealing Temperature.*

Annealing Tempera-ture. ° C.	Uniform Elongation, %, after Cold-Working.											
	0%.	5·6%.	14·0%.	19·2%.	29·5%.	40·7%.	46·5%.	57·2%.	67·9%.	75·9%.	84·2%.	93·8%.
20	26·0	14·2	5·3	5·0	1·0	1·5	1·2	0·3	0·2	1·0	0·0	0·8
143	25·0	7·8	5·8	2·8	1·5	2·5	2·6	1·8	1·0	2·2	1·5	0·3
201	23·5	8·2	5·5	2·8	2·0	3·3	2·3	2·2	2·5	2·2	2·3	1·0
260	24·2	7·7	6·4	6·2	3·8	3·6	2·1	2·5	2·3	2·8	2·8	3·8
294	24·2	8·0	6·4	5·5	3·9	3·8	2·1	2·8	3·8	2·2	4·2	4·0
320	23·0	6·0	6·3	6·5	5·4	4·6	3·6	3·5	3·6	2·6	4·6	5·2
350	24·8	8·0	7·5	6·6	6·0	4·6	3·0	3·7	3·7	4·0	5·8	6·0
387	24·2	6·4	7·6	7·3	7·3	5·8	3·8	4·0	5·3	5·3	6·7	6·0
419	22·8	9·5	9·2	7·3	6·4	5·8	5·2	3·9	5·6	6·3	8·6	9·0
450	24·5	8·0	10·2	9·5	6·7	7·3	4·2	5·6	5·6	8·8	10·3	10·8
475	25·0	10·8	12·1	10·3	8·0	8·0	5·0	6·2	6·4	10·5	12·3	12·8
494	24·2	11·0	11·5	11·3	10·0	9·3	9·2	6·4	9·2	12·3	14·0	13·8
525	23·0	13·8	10·8	10·8	11·0	10·6	9·3	8·8	9·4	11·8	14·0	15·3
549	24·0	15·3	12·2	10·5	12·0	12·5	12·0	11·7	10·4	13·5	16·4	17·2
578	23·5	14·0	13·9	12·5	12·0	12·8	13·0	12·0	12·2	15·2	17·8	21·2
590	...	...	...	13·8	13·8	14·0	14·2	12·6	14·7	18·6	26·0	34·2
607	22·8	15·4	17·5	14·5	14·7	16·8	16·2	15·6	23·8	24·2	36·8	37·8
630	...	...	...	15·3	15·4	17·0	21·5	34·2	38·3	38·0	40·5	40·8
650	...	...	...	18·4	28·6	31·8	35·3	36·8	39·0	39·0	41·2	40·0
675	...	...	...	20·7	28·5	34·6	36·5	39·6	39·0	40·3	42·0	40·0
701	...	...	...	23·2	28·0	36·2	37·8	38·5	41·2	40·8	42·0	39·6
729	...	...	...	25·2	22·0	37·9	37·8	40·6	40·5	41·5	43·0	38·0
749	...	...	...	26·0	32·8	37·2	37·8	38·5	37·0	40·8	42·2	39·0
850	...	...	...	27·6	32·5	36·1	37·0	37·0	39·6	40·3	41·0	35·8
950	...	...	...	8·0	5·3	7·1	7·0	6·8	8·0	6·8	9·5	6·0

(4) Annealing of the cold-worked wires at successively higher temperatures causes this increase to grow until at certain temperatures there is a sudden jump to the high values, which are thereafter not much affected by further heating to higher temperatures.

(5) Wires which were heavily worked prior to annealing show, after heating above 600° C., almost double the values of the ductility of normalised iron. The 93·8% cold-worked wires appear, however, to be less ductile than the others of the group.

(6) Heating to very high temperatures, *i.e.*, normalising for a second time even for a very short period, had a striking lowering effect on the ductility. These features may be connected with the formation of large crystals and an intercrystalline type of fracture.

The change in the tensile properties of iron and mild steel as a function of cold-working and of subsequent annealing has been frequently studied under various experimental conditions by numerous investigators. It is regretted that many interesting results of Kenyon and Burns,[40] Jeffries,[41] Pomp,[42] Comstock,[43] Pfeil,[44] Köster and Tiemann,[30] Freeman,[45] McAdam, jun.,[46] Francis,[47] Elam,[48] Broniewski and Krol,[8] Broniewski and Robowski,[9] Goerens,[11] Swinden and Bolsover,[14] Edwards, Phillips and Pipe,[15] and many others cannot be referred to more fully.

Microscopical Examination.

Although the structural changes which accompany the processes of cold-working and subsequent annealing have very frequently been studied, it was considered worth while to supplement the other methods of testing by microscopical examination. Under this examination there was no observable difference between the wires normalised and 5·6% cold-drawn and those cold-worked and annealed at temperatures lower than 525° C. At annealing temperatures above this value there is a change in the directional structure of the heavily cold-worked wires. Heating of the wires to still higher temperatures causes only quantitative changes, described as crystal growth. Wires with a low degree of cold-working are specially susceptible to this growth and their structural transformation takes place at temperatures higher than 525° C. Annealing at 950° C. affected all the wires in very much the same manner and produced large crystals plainly visible without any magnification. It was also evident that crystal growth depends upon the degree of cold-working before annealing. This variation in the crystal size can easily explain the fluctuations in the values of the properties above the recrystallisation range.

Effect of Increasing the Annealing Time.

It was of some interest to know whether such a short annealing time as was adopted in the experiments was long enough to obtain uniformity of temperature across the specimens, for it was feared that this lack of uniformity might be responsible for the peculiar behaviour of some of the wires, especially in the tensile tests which exhibited an exaggerated yield point. To answer this and to learn how the results might change should a longer annealing time be adopted, a series of wires with 93·8% of cold-work were annealed at different temperatures for one hour. To secure comparable conditions the same arrangements for heating were used as in the former case. On approaching the required temperature, the container with the specimens was quickly transferred to a similar furnace maintained at the desired temperature level. To eliminate the effect of possible change of composition two samples were annealed simultaneously. The mean results are presented in Fig. 11, which, for comparison, contains

corresponding curves determined earlier. The curves for both series of experiments are parallel and run close to each other. This confirms the view that the time of annealing is less important than the temperature. As for the peculiarities in the appearance of the stress-strain diagrams, full confirmation of their existence has been obtained despite the change in the experimental conditions. From this it is concluded that this occurrence is not accidental, but is due to the structural changes in heavily cold-worked iron developed by the action of heat.

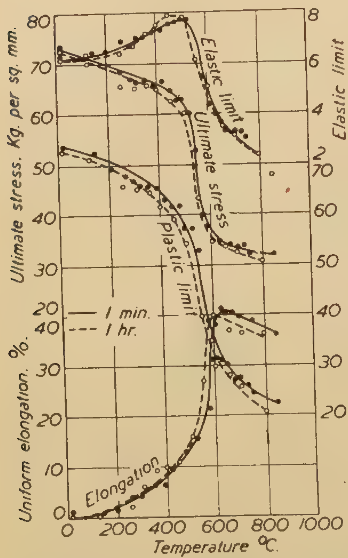

FIG. 11.—Effect of Prolonging the Annealing Time on the Properties of Armco-Iron Wire.

It ought to be noted that the wires for these last-mentioned experiments were newly prepared, immediately annealed and tested, so that they were free from the effect of any ageing during storing. From the negligible difference in the initial values and the immediate lowering of the tensile properties by low-temperature annealing, it follows that very heavily cold-worked iron wire does not age after the completion of drawing.

The wire specimens used for further experiments undertaken to examine the effect of annealing times ranging from 1 min. to 61 hr. consisted of a set with 19·6%, 46·6%, 67·8% and 93·4% of cold-work. Again, to avoid the effect of ageing, fresh wires were drawn and were immediately seasoned for 1 min. at 575° C., the temperature chosen as the annealing temperature. It was appreciated that by so doing one introduced an additional heating stage to the annealing temperature, but this was in fact common to all the samples.

All specimens with varying degrees of cold-work, but intended to be annealed for the same time, were treated simultaneously in order to obtain better conditions for comparison. The fluctuations in the temperature of the specimens was negligible (± 0·5° C.) during all the annealing periods.

After annealing, the specimens were subjected to a tensile test, the results of which are shown in Table VII. with the appropriate values for cold-worked wires. Fig. 12, illustrating the changes in the ultimate stress, also serves to illustrate the changes in the plastic limit, for they are very similar. In conformity with the previous results, most of the change takes place at short times of annealing. Annealing for more than one hour produces further diminution in the values, but this reduction is comparatively small. Even after holding for 61 hr. at 575° C. the wires were not fully annealed and it appeared that a very much longer time would be necessary to attain the fully annealed state. The greater the degree of cold-work the quicker was the drop in values obtained at the beginning of annealing, as can be seen from the drop to more or less the same level. If allowance is made for the opposite direction of the values of the changes in uniform elongation, the above remarks apply for this property also.

TABLE VII.—*Properties of Cold-Worked Armco-Iron Wire after Annealing at 575° C. for Different Times.*

Q = plastic limit; R = ultimate stress; a = uniform elongation.

Annealing Time at 575° C.	19·6% Cold-Work.			46·6% Cold-Work.			67·8% Cold-Work.			93·4% Cold-Work.		
	Q. Kg. per sq. mm.	*R.* Kg. per sq. mm.	*a.* %.	*Q.* Kg. per sq. mm.	*R.* Kg. per sq. mm.	*a.* %.	*Q.* Kg. per sq. mm.	*R.* Kg. per sq. mm.	*a.* %.	*Q.* Kg. per sq. mm.	*R.* Kg. per sq. mm.	*a.* %.
0	41·7	44·1	5·0	53·0	53·0	1·2	54·4	56·1	0·2	73·6	73·6	0·8
1 Min.	31·2	39·9	25·8	50·2	51·3	24·5	39·5	46·6	32·8	48·8	52·2	29·8
4 Min.	...	38·2	...	36·4	42·2	27·7	31·7	39·5	33·8	39·4	40·3	30·0
7 Min.	29·6	38·2	26·4	32·2	39·7	28·7	28·3	34·6	31·8	32·5	36·0	34·5
15 Min.	29·6	38·2	27·3	31·7	41·7	27·0	24·5	33·6	34·0	32·4	34·5	36·0
30 Min.	28·7	38·3	27·3	31·7	38·9	29·5	24·5	34·3	33·7	32·3	35·0	34·7
1 Hr.	28·6	38·0	27·7	27·5	36·0	29·6	27·0	34·2	34·3	28·7	37·3	35·6
3 Hr.	28·2	37·8	28·6	29·3	36·5	32·0	28·7	34·4	34·3	30·9	35·2	35·9
6 Hr.	27·2	36·8	27·3	27·4	35·0	33·0	26·0	34·5	34·3	30·0	35·0	36·4
16 Hr.	27·2	35·9	28·2	24·3	33·2	33·8	26·2	33·5	35·2	31·7	34·6	37·6
35 Hr.	27·0	35·5	29·2	23·4	33·6	34·8	25·7	33·0	35·0	27·8	35·0	36·0
61 Hr.	27·2	35·6	28·6	23·1	34·6	34·7	25·0	33·8	35·9	27·3	34·6	37·2

It should be noted that the curves are, in general, exponential and in this they resemble those observed with copper.[3]

An exact comparison of the results with the published data is not possible because of the changes in the annealing conditions. The results of Goerens,[11] however, are similar to those obtained here.

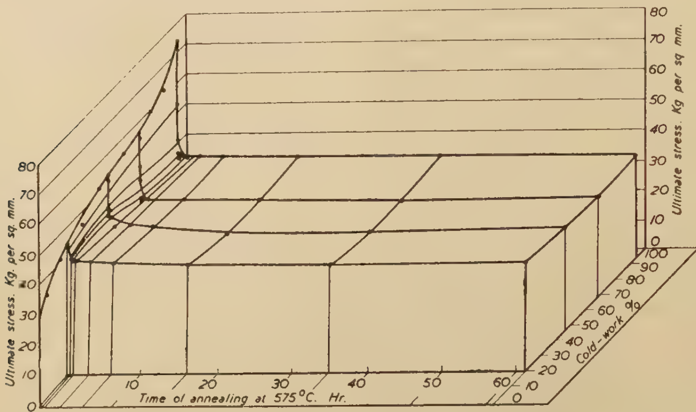

FIG. 12.—Ultimate Stress of Armco-Iron Wire in relation to cold-work and annealing time at 575° C.

EFFECT OF STRAIN-AGE-HARDENING DURING STORING UPON THE MECHANICAL PROPERTIES.

It is stated in the scheme for preparing the specimens (Fig. 1) that the wires were kept in a refrigerator at 5° C. in order to slow down the ageing as much as possible. Despite this, however, some ageing might have taken place in the refrigerator as well as outside during the tests. To check the magnitude of this influence appropriate tests were carried out on wires with 67·8% of cold-work.

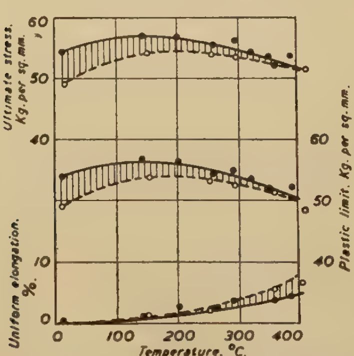

FIG. 13.—Effect of Ageing on the Properties of Armco-Iron Wire Cold-Worked 67·8%. Full line, aged specimens; dotted line, unaged specimens.

The normalised wire of suitable diameter was drawn to the standard diameter in exactly the same manner as previously. A sample cut off immediately after the completion of the drawing was at once tested for tensile properties. Other samples similarly prepared were immediately annealed for one minute at different temperatures and then tested. The results of the tests are shown in Fig. 13 together with the corresponding points from the main set of tests. The shaded areas indicate the change in the properties due to ageing. The possible ageing during the actual drawing operation, it will be remembered, is not accounted for. The effect is greatest, and amounts to about 10%, in the case of material which has not been annealed; it diminishes with increasing annealing temperatures.

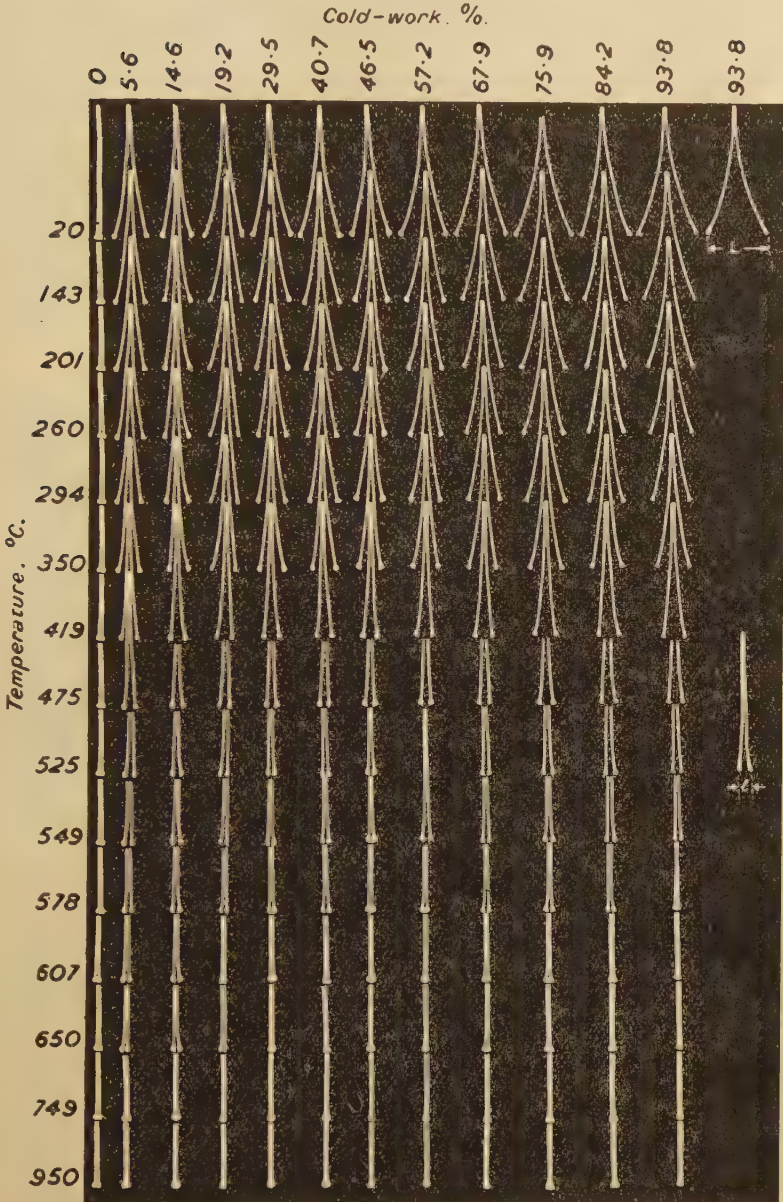

Fig. 14.—Silhouettes of Specimens of Cold-Worked Armco-Iron Wire after "straightening" for 1 min. at various temperatures.

Similar experiments were performed with material of other degrees of cold-work and a similar behaviour was revealed, although for the preparation of these wires material normalised in a vacuum, not in nitrogen, was used.

In conclusion, it may be pointed out that this interesting effect has a bearing upon the main subject of this investigation. With regard to the future theoretical considerations, it is essential to remember that, despite the precautions taken, the wires aged during storing and presumably also during the actual drawing operation. Consequently, both the initial values of the properties and also the beginning of the curves for various properties in relation to the annealing temperature of iron may differ according to the progress of ageing. There is as yet no criterion by which it is possible to decide which value or curve represents most accurately a given degree of cold-work in iron.

Tests at Elevated Temperatures.

Springiness.

The wires were not absolutely straight after drawing, but were slightly curved as a result of the uneven flow of material in the die and of uneven distribution of the residual stresses. This fact suggested a method of studying the action of a straightening force at elevated temperatures.

For this purpose pieces of wires representing each degree of cold-work were cut off and bent to 180° in the middle in the direction opposite to the curvature to produce V-shaped specimens. The ends of the wire were kept against each other by means of a tight ring for a period of one minute. After removing the ring, each specimen was placed on the surface of a photographic paper on which its shape was printed. Such a silhouette is clearly shown on the right in Fig. 14.

Afterwards the rings were replaced on the specimens and the whole set put in a furnace together with the specimens destined to be tested for the effect of temperature upon the various properties. The former specimens, when quite cold, were freed from the straightening force and their new outlines registered. Subsequently, the same specimens were heated to successively higher temperatures. The results showed the changes produced in specimens of various degrees of cold-work after annealing at different temperatures.

Some explanation of the two silhouettes on the right of Fig. 14 is necessary. As the same specimen representing a single degree of cold-work was successively heated to higher temperatures, some cumulative action from each heating might be anticipated. To check this, an additional sample with 93·8% of cold-work was prepared, photographed after straightening at room temperature and annealed with other specimens at a temperature of 525° C. only. Comparing the two silhouettes—one of the specimen annealed once only at 525° C. and the other of that heated several times until this temperature was attained—some of the expected differences can be observed. The difference, however, is small and from this it may be deduced that the temperature to which the specimens were heated had in this instance also a greater influence than repeated heating, *i.e.*, than a longer heating time.

In order to give a numerical value to these changes it was decided to express the distance between the ends after heating to a particular temperature as a percentage of the initial spread after straightening at room temperature. The changes in this ratio are plotted in Fig. 15.

The test itself as outlined above gives some direct information about the resistance which iron offers to the action of external forces while at high

temperatures. Less directly it reflects the internal transformations occurring in the material when subjected to the action of stress and temperature.

Analysing the springiness test from this point of view, the problem resolves itself into a consideration of the bending at elevated temperature of two built-up beams by the moment developed by the straightening action of the ring. This bears some similarity to the bend test.

A strict comparison of the results obtained from both methods of testing cannot be made unless suitable corrections are introduced for the following important features in the springiness test :

(1) The bending occurs at elevated temperatures.

(2) The bending moment is approximately the same along the whole length of the wire, hence the total deflection is greater than in the case of cantilever bending.

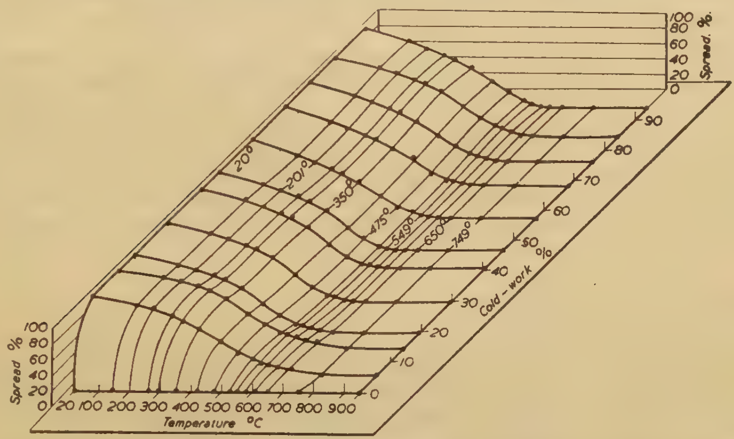

FIG. 15.—Spread of the Ends of Bent Specimens of Cold-Worked Armco-Iron Wire after releasing the clamping ring.

(3) The value of the moment diminishes not only from one heating stage to another but even during the same heating stage.

(4) The same sample was repeatedly heated.

These differences, particularly the first, explain why curves from the springiness test follow different courses from those obtained from elastic-limit determinations in the bend test, and why the commencement and finish of the drop occur earlier. The duplex structure of a partly annealed metal, as explained later in the theoretical section, may be an additional reason for the observed changes.

Summary of the Results of the Springiness Test.

(1) The curvature of the arms of the samples decreases on heating and the change most convenient for observation is that of the distance between the ends. The normalised wire being straight at the time of preparation, V-specimens would naturally not reveal any spread of the ends. It has been included merely to complete the series.

(2) As was anticipated there is no marked difference in the spread of the specimens representing different degrees of cold-work. The small differences were eliminated by taking the initial spread as the basis.

(3) The distance between the ends of all the specimens begins to decrease at very low annealing temperatures, which is significant when compared with the results obtained for the plastic limit in the tensile tests and with the elastic limit in bend tests.

(4) In all specimens the decrease continues with increasing annealing temperature up to roughly the same region of temperature where distinct changes in microstructure occur and where the other properties approximately regain the value they had prior to cold-working.

(5) There is only a slight indication of the effect of increasing degree of cold-work upon the temperature at which the spread of the ends becomes zero.

Apart from a short note by the author about the results obtained using the same method with nickel [3] there are no actual data for direct comparison. The nearest analogy to the test itself is the method developed in the Woolwich Research Department,[52–55] where bending on formers was applied.

The method of Dreyer and Tammann [51] has the same mechanical characteristics as that described here, although there are some superficial differences.

The curves obtained by the above-mentioned authors for iron and other metals reveal in each case the same characteristic descent in the curve, commencing at very low annealing temperatures, as was found in the present work. It was shown in the present investigation that this diminution continues up to much higher annealing temperatures than that found by others.

Thermo-electric Properties.

Another series of experiments in which the principle was the determination of the change of a property with increasing temperature was the determination of the functional relationship between the thermal e.m.f. and temperature of a thermo-element constructed of a cold-worked metal and the same metal annealed. The apparatus shown diagrammatically in Fig. 16 was used for the experiments. It was developed from an arrange-

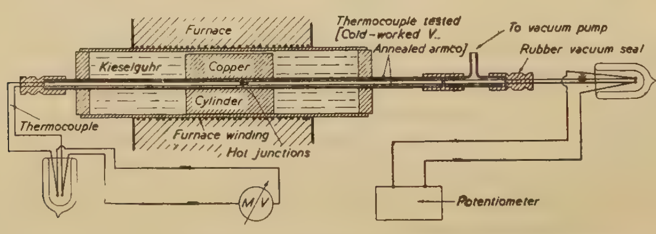

FIG. 16.—Apparatus for Determining Thermo-electric Properties.

ment previously used [2] in order to overcome certain experimental difficulties. As can be seen from the figure, the apparatus consists of a copper block placed slightly to one side of the middle of an electric furnace so as to secure uniformity of temperature of the hot junctions and also to produce a temperature gradient along the couple. Around the block was placed a thin earthed iron tube to insulate the experimental iron thermocouple from the magnetic field of the furnace and to contain the insulating material. In the axis of the copper block and the furnace there was a

narrow silica tube connected with a vacuum pump. Through one end of this tube a chromel-alumel thermocouple was inserted and through the other the thermocouple to be tested. This consisted of one wire in the cold-worked state and a second wire of standard iron annealed at 675° C. for 1 hr. The hot junctions of the thermocouples overlapped each other so as to measure, as accurately as possible, the temperature of the bead of the iron

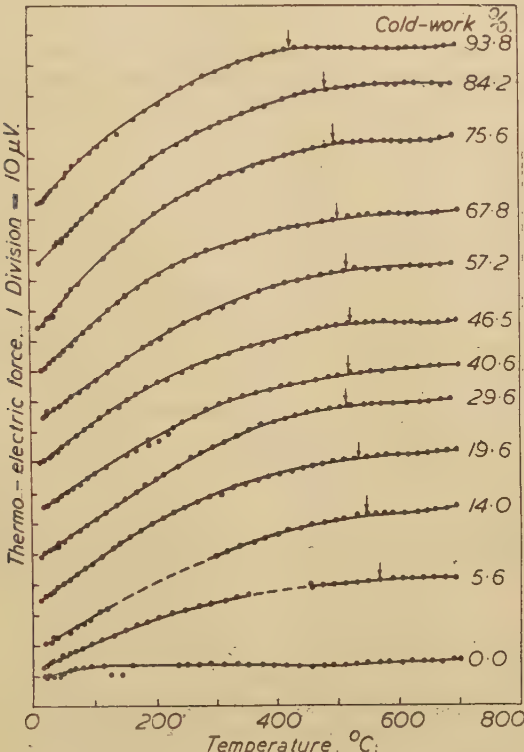

Fig. 17.—Effect of Cold-Work on the Thermal E.m.f. produced by Cold-Worked/Annealed Iron Thermocouples at Different Temperatures.

couple. It need hardly be pointed out that the wires were carefully insulated. The cold junction of the iron couple was the most troublesome part of the apparatus and it was considered that this was the main source of inaccuracy of the determinations. This is caused by the comparatively large thermal e.m.f. developed between the iron and copper leads. Thus any difference in temperature between the two cold junctions would have a great bearing upon the whole determination. To suppress this effect, great care was taken to ensure that both cold ends were at a constant temperature.

The results of the determinations are shown in Fig. 17 by a set of curves characterising each degree of cold-work. The slope of these curves in the 20–100° C. range served for the determination of the values of the thermo-electric power (de/dt) and the first flat portion of the curves for the

determination of the maximum e.m.f. ($e_{max.}$). The values are shown in Fig. 18 as a function of the degree of cold-work. Both the curves are similar in trend to the curves obtained for the strength properties. In the early stages of cold-work there is a rapid increase followed by a more or less flat portion in the region between 30% and 70% of cold-work. For higher degrees of cold-work there is a tendency for a subsequent increase less clearly demonstrated in the case of $e_{max.}$.

Summary of Results.

(1) Cold-worked iron and annealed iron forming a thermocouple develop a very slight e.m.f. potential and therefore special care is necessary to secure the desired degree of accuracy. From a consideration of the results

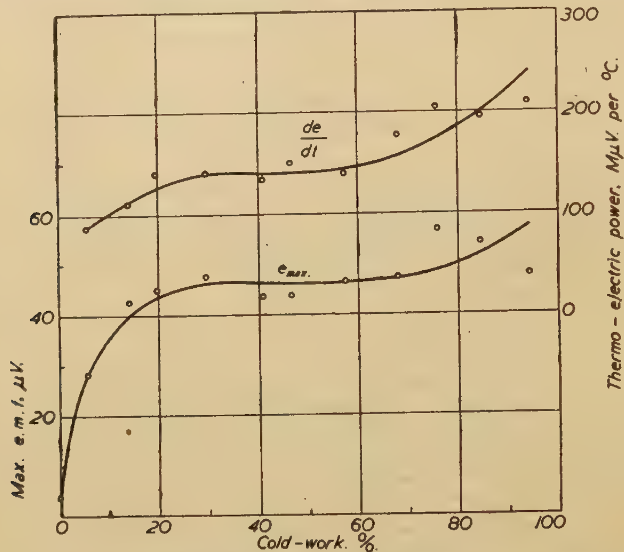

FIG. 18.—The Thermo-electric Power and the Maximum e.m.f. of an Iron Thermocouple in Relation to Cold-Work.

obtained when two annealed wires were tested it was found that unfortunately factors beyond control were operating and slightly marred the results.

(2) A study of the thermal e.m.f. of a couple made of annealed and cold-worked iron wires in relation to the temperature of the hot junction reveals characteristic changes which mainly reflect the internal transformation caused by annealing. Some contribution to the change may be caused by : (*a*) chemical heterogeneity of the wires; (*b*) variation in temperature of both cold ends; and (*c*) the fact that the thermo-electric-power/temperature relationship may not be constant.

Testing wires of varying degrees of cold-work revealed purely quantitative differences.

(3) Changes due to cold-working can be most easily estimated when the characteristic coefficients deduced from the experimental curves are compared. These are: (1) The thermo-electric power in the 20–100° C. range; and (2) the maximum e.m.f.

(4) Both the thermo-electric power and the maximum e.m.f. increase

as the degree of cold-work increases. The course of the change resembles that of the increase in the strength properties.

(5) From the commencement of the bend in the curves it again appears that the beginning of recrystallisation of iron occurs at very low temperatures.

Comparison of the Results with those of other Investigators.

The thermal e.m.f. of cold-worked metals and that of metals under external stress during the determination have been studied by numerous investigators;[(56-58)] particularly the work of Borelius[(59)] and of Thiele[(60)] threw much light upon this property of metals. Goerens[(11)] was not very successful in his determinations of the e.m.f. of iron.

The method of Tammann and Bandel[(61)] was similar in many respects to that described herein with the exception that they heated the hot junction in a gaseous atmosphere and at a less rapid rate, namely, at 2·5° C. per min., compared with 17·8° C. per min. in the present work. They presented the results not as electromotive force but as thermoelectric power. Such a curve shows an increase in the region between 70° and 320° C. According to Thiele's work an ascending type of curve, instead of a descending one as in the case of other metals, might be explained as due to accidental magnetisation either during the period of preparation or during the actual testing of the specimens.

General Survey of the Results of Reannealing.

From the detailed descriptions given in the preceding sections it is somewhat difficult to form a clear general view of the changes in properties resulting from reannealing. For this purpose it is convenient to plot in one diagram all the experimental curves determined for one degree of reduction and then compare them. This has been done in Fig. 19, from which it is seen that the curves are of a variety of types, because they illustrate the changes either in various properties or in the same basic property as determined by different methods.

Conditions of Comparison.

At this stage it may be pointed out that the conditions for comparison are good so far as the composition and the structural uniformity are concerned. The curves were determined in many cases on one and the same specimen, whilst in other instances they were determined on the same material with the same pretreatment. The direction of the tests, except those for hardness, was always in the direction of the axis of the wires. To emphasise the difference the hardness curve has been dotted and so have the curves for springiness and the thermal e.m.f. Apart from the fact that separate portions of the wire were used for the determinations, it must be remembered that there is present, in the case of springiness, the cumulative effect of repeated heating of the same samples and the action of the testing stresses at elevated instead of at room temperature. Additional reasons for dotting the curve for the thermal e.m.f. have been given in the appropriate section. The expected deviation in the tendencies of these curves calls for some corrections; the individuality of each method of testing should also be appreciated. To illustrate the extreme magnitude of these effects and the allowances necessary, it will be remembered that the plastic limit, springiness and elastic limit deduced from the bend test reflect the same basic property, namely, the transition from the elastic to the plastic state. Consequently, one might expect the curves to be similar and parallel. As can be seen from Fig. 19, this is not the case, and this is due to the numerous operating factors.

Survey of the Curves.

All the curves in Fig. 19 reveal the main change in the region of medium temperatures. This change could be identified with the aid of the microscope as being due to recrystallisation; the changes in the high-temperature region were found by the same means to be due to crystal growth. In the low-temperature region the curves for hardness, ultimate stress, plastic limit and elastic limit from the bend test reveal increases which are signs of

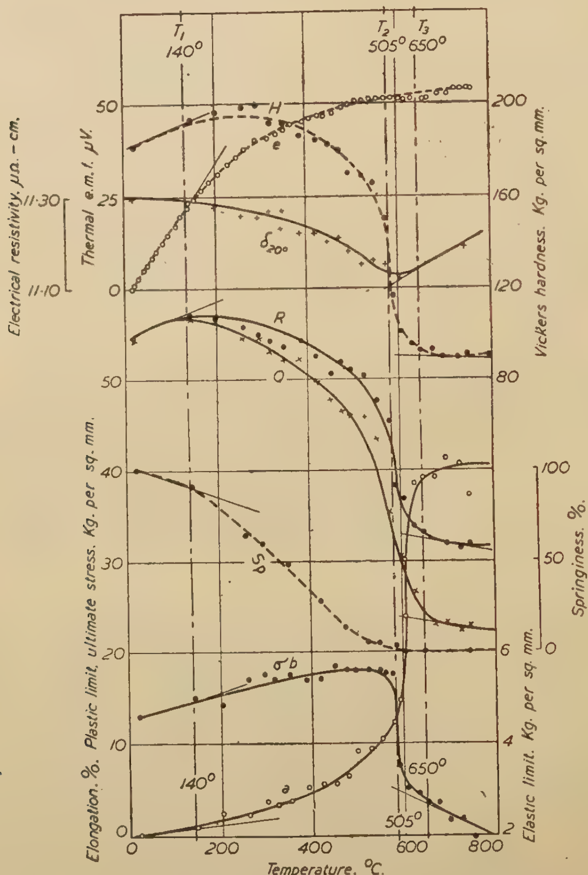

FIG. 19.—The Effect of the Annealing Temperature on the Properties of 67·9% Cold-Worked Armco-Iron Wire.

the presence of ageing or, more correctly, seasoning. As recrystallisation changes are the most prominent of all, they will be the main topic of the present consideration; crystal growth and strain-age-hardening will be considered as influencing factors. It will also be appreciated that there are no distinct boundaries between the regions of occurrence of these phenomena and that they possibly overlap the region of recrystallisation.

From both microscopical studies and the change in the properties it

follows that recrystallisation does not take place at a definite temperature but in a temperature range. To define this range it is necessary to define two temperatures, that of the commencement T_1 and that of the end T_3 of recrystallisation. Direct microscopical observations, even when very high magnifications are used, are not sufficiently sensitive to determine the temperature of the commencement of recrystallisation. Even X-ray methods do not appear to have been developed enough to reveal clearly the very beginning of the changes in the structure. An indirect approach to the problem can be made by an analysis of the changes in the physical properties. As there are masking effects present, it is only by studying a number of properties extensively and accurately that an objective result can be secured.

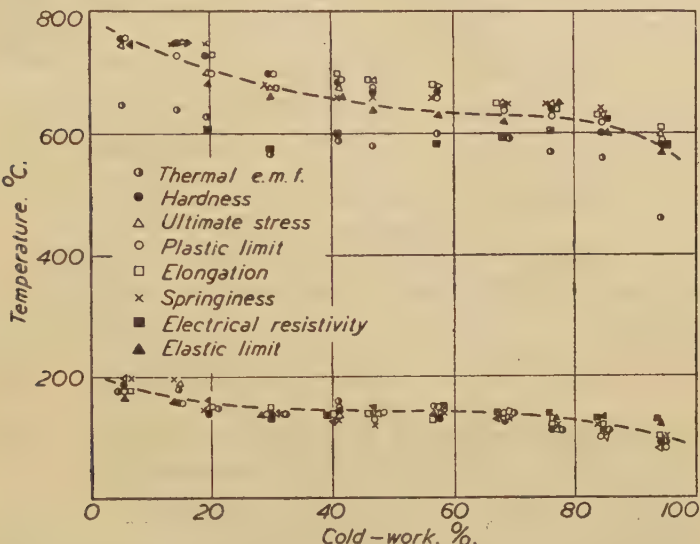

FIG. 20.—The Temperatures at which the Recrystallisation of Armco-Iron Wire Begins and Ends, as indicated by changes in various properties.

There are two methods of determining the temperature of the commencement or end of a change from the course of a curve; both yield the same result in the case of a single change with no overlapping and both methods will be employed for the determination of the recrystallisation range.

In the first method of arriving at the value of T_1, the practical commencement and the critical temperature of the main drop are estimated by studying the course of the curve from the high to the low temperatures of annealing. For most of the curves in Fig. 19 the mean value thus obtained is 275° C. The closely related curves for the elastic limit from the bend test and for springiness only yield different and contradictory values.

The other way of determining T_1 is to draw a tangent to that portion of the curve where the recrystallisation change cannot be far advanced, and observe the departure of the curve from the initial slope. By applying this method to Fig. 19 a value of 140°, which is practically common to all the curves, is obtained. The discrepancy between the two results for T_1 can be linked with the presence in this region of strain-age-hardening and

with the practical difficulties of adopting a rigid treatment to a case which requires some latitude.

Owing to the presence of the effects caused by crystal growth, the application of the same methods to determine the temperature of the end of the processes was difficult. Nevertheless, an average value of about 640° C. for T_3 was arrived at. As can be seen in the diagram (Fig. 19) all the main changes cease quite close to that temperature.

Another characteristic temperature is T_2, which is the mid-point of the

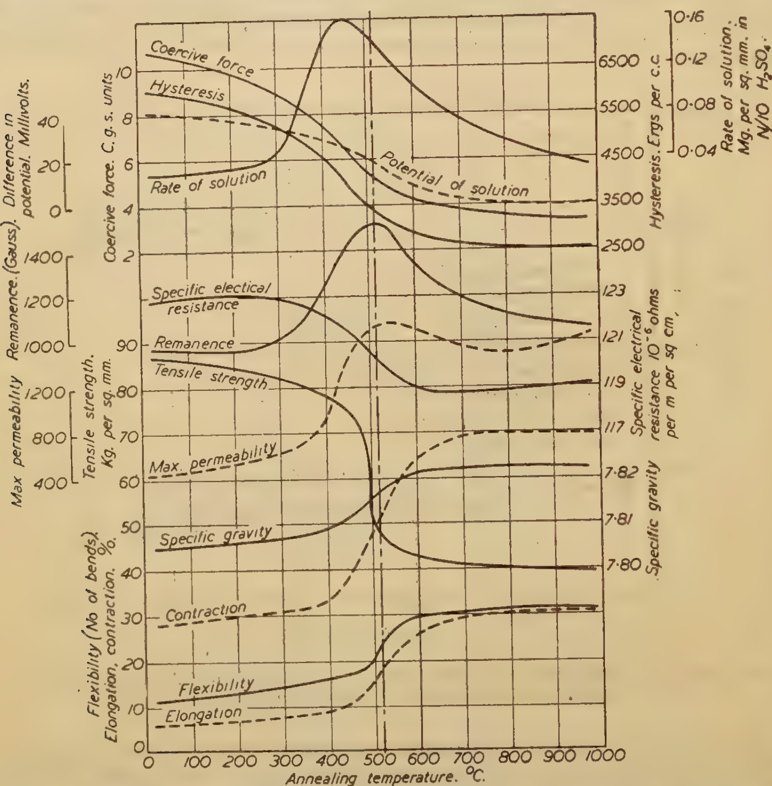

FIG. 21.—The Changes in the Properties of Iron due to Recrystallisation (according to Goerens [11]).

recrystallisation range, where the process first appears to slow down. It is therefore convenient to adopt the temperature of the point of inflection in the curve as T_2. For the majority of the curves in Fig. 19 this temperature is practically the same and is about 585° C. The curves for springiness, thermal e.m.f. and electrical resistivity yield somewhat lower values, which is not very surprising.

It was considered instructive to carry out the same determination for all other reductions. For this purpose all of the values of T_1 and T_3 from all the curves were determined and plotted in one diagram (Fig. 20). It can be readily seen in this diagram that, on the whole, the points fall sufficiently

close together for each degree of cold-work and that the agreement of the results is better with the heavily cold-worked specimens. This can possibly be explained by the somewhat lesser effect of seasoning, crystal growth and other influences in that region.

An additional result from this treatment is the dependence of the recrystallisation range upon the degree of reduction as defined by the two curves in Fig. 20.

Comparison of the Results with those of Others.

It is possible to compare this result with that found by Goerens[11]; for this purpose Fig. 21 has been compiled from his tables, in which some of the curves are dotted for clarity. A careful analysis of the experimental conditions reveals that the curves are comparable with those in the present work. Allowances for experimental error have to be appreciable in some cases, however, as can be deduced from the order of scatter of the points determined experimentally and from the remarks in the paper.

A survey of the curves confirms the fact that the recrystallisation changes take place in the same range of temperature, the commencement of the changes again being somewhere between 100° and 200° C. and the end between 600° and 700° C. On this point Goerens observed that all the main changes in all the curves, within the experimental error, take place at a temperature of 510° C. From this it follows that he considered the recrystallisation change to be an isothermal process.

Other papers which deal with the subject are those of G. Tammann and his co-workers.[20, 50, 51] Unfortunately, as can be seen from Table VIII., the experimental evidence of these workers was not collected methodically enough to be used for comparative purposes. This may explain why Tammann's deduction with respect to the simultaneous occurrence of the recrystallisation changes in iron differs from that made by the present author.

Survey of the Results Relating to the Effect of Work-Hardening on the Properties.

In the preceding section the last of the properties to be studied as a function of the reduction was the recrystallisation range. This permits a detailed survey of the observed effects to be made. To facilitate this review all curves relating to the worked metal have been collected in Fig. 22. The figure includes also the expression $P_m = R(1 - z)$ proposed by Krupkowski,[39] where R is the ultimate stress of the material reduced by the amount z by cold-work. The different curves in Fig. 22 can be compared especially when certain minor allowances for the factors previously indicated are made. As can be seen from the diagram, the curves for ultimate stress, plastic limit and elastic limit from the bend test all rise as the degree of reduction increases. The same can be said with regard to the curves for the thermo-electric properties and that for the electrical resistivity, although the last curve shows a pronounced decrease in the region of small reductions. Most probably this is a result of the breaking up of the crystal boundaries. The curves describing the approximate boundaries of the recrystallisation range decrease with the amount of reduction in a manner which confirms an earlier observation made about the linear relationship of some of the properties of copper [1] to the temperature of recrystallisation. As can be seen in the curve for uniform elongation, the ductility decreases rapidly up to about 24% reduction and subsequently retains its negligible value. The P_m curve rises slowly to a maximum, after which a steady decrease occurs.

An examination of this set of curves in a manner similar to that out-

TABLE VIII.—*The Testing Conditions in the Investigation by Tammann and His Co-Workers of the Reannealing of Cold-Worked Iron.*

Property.	Composition of Specimens.	Method of Cold-Working.	Degree of Cold-Work. %.	How Annealed.	Temperature of Testing.	Remarks and References.
Hardness	Not stated	Rolling	90·0	Not stated	Room temperature	Thin strip tested on steel anvil.(20, 63)
Solubility in H_2SO_4	Not stated	Rolling	84·7	Not stated	Not stated, probably at room temperature	Tammann and Rocha.(63)
Electrical resistivity	Not stated	Drawing	98·7	Tested at increasing temperatures in hydrogen		Tammann and Moritz.(20)
Change in internal energy	Armco iron	Torsion	Not stated	In nitrogen or hydrogen	Elevated	Satô.(64)
Hardness	Electrolytic iron	Rolling	80–90	Not stated	Room temperature	Thin strip, not clear whether same load used.(51)
Change in electrical resistivity	Probably electrolytic iron	Rolling	Not stated	Not stated	Elevated temperature, probably in hydrogen	Dreyer and Tammann.(51)
Broadness of X-ray interference lines	Probably electrolytic iron	Rolling	Not stated, probably 95–98	Annealed under stress for 1 hr. (atmosphere not stated)	Room temperature	Dreyer and Tammann.(51)
Springiness of a coiled strip	Probably electrolytic iron	Rolling	95–98	Annealed under stress for 1 hr. (atmosphere not stated	Actual testing at elevated temperature. Determination of the effect at room temperature	Dreyer and Tammann.(51)
Number of bends	Probably electrolytic iron	Rolling	Probably 95–98	0·5–1·0 hr.	Room temperature	Dreyer and Tammann.(51)
Deflection under constant load	Probably electrolytic iron	Rolling	Probably 95–98	0·5–1·0 hr.	Room temperature	Dreyer and Tammann.(51)
Magnetic induction	Electrolytic iron	Drawing	98·0	For 45 min. *in vacuo*. (One specimen for all the tests)	Room temperature	Tammann and Neubert.(62)

lined by Krupkowski [39] reveals the presence of three regions of plasticity. The end of the first region is indicated by loss of ductility, the maximum value of P_m and by the coincidence of the curves for the ultimate stress and the plastic limit. Above this boundary, at about 24% reduction, that is, in the second region of plasticity, the curves fluctuate less and are

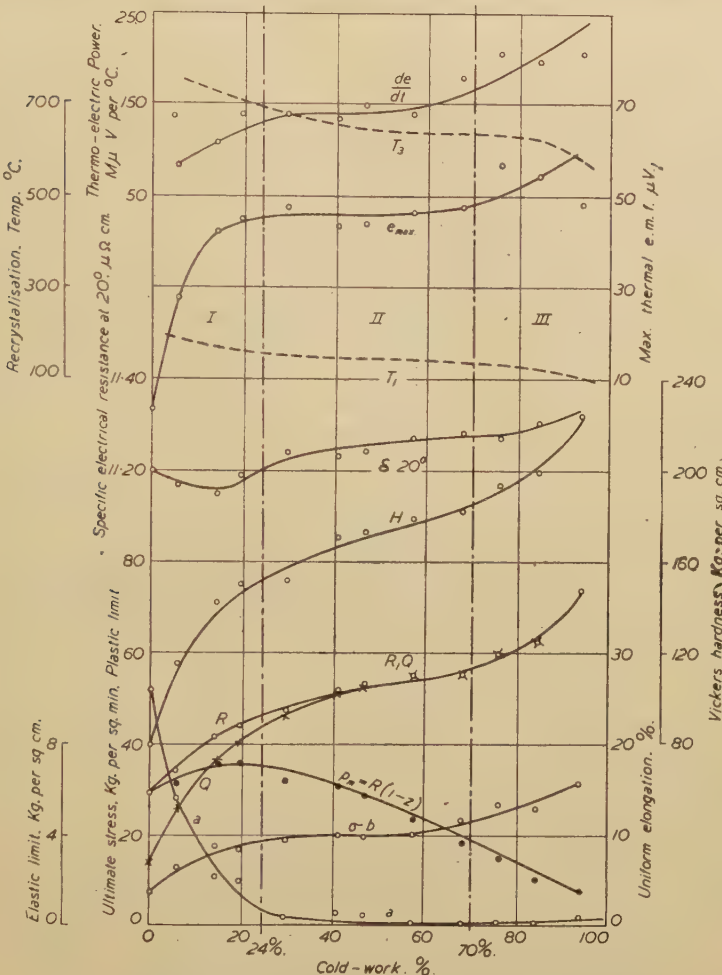

FIG. 22.—Effect of Cold-Work on the Properties of Armco-Iron Wire.

flat. In the third region, that is, above 70% reduction, most of the curves rise or fall more rapidly.

In addition to the reasons for this classification of Krupkowski,[39] investigation of the reannealing changes supplies some further evidence for its acceptance. It will be remembered that the wires from the first region were very susceptible to seasoning and to crystal growth; the highest

ductility after annealing was observed in the wires belonging to the second region, and samples from the third region revealed peculiar stress-strain diagrams. It should be appreciated that this division into the three regions can serve as a more rigid definition of the terms low, medium and heavy reductions and that it supplies two values which characterise a metal and its pretreatment. Broniewski and his co-workers [8, 29] used this method for technically and chemically pure iron.

Thermodynamical Consideration of Work-Hardening.

Any discussion of recrystallisation necessitates a reference to work-hardening, for these two phenomena are closely related. Structural changes resulting from cold-working have been the subject of numerous investigations which have resulted in several theories. A convenient aspect of work-hardening, sufficiently general to be in agreement with all sound theories, is arrived at when work-hardening is considered from the thermodynamical point of view.

A study of the flow of a metal during plastic deformation which causes work-hardening to take place, reveals, when full consideration is given to the mosaic structure of crystals and to the mechanism of plastic flow, that a slightly cold-worked metal cannot be regarded as homogeneously work-hardened. Some groups of atoms, contiguous to the gliding and rotating planes, should be considered as actually work-hardened, whilst the interiors of the crystal blocks remain unchanged and retain fully the properties of the annealed state. There is some change in the orientation of these crystal blocks, but this alone cannot be the sole cause of the changes in view of the changes in the properties which are known to be independent of orientation (*e.g.*, electrical resistivity of metals crystallising in the cubic system) and because the metal is affected in all directions.

As the interiors of the crystallites retain their structure unchanged, they will be considered as being still in exactly the same thermodynamical state as that prior to deformation, the change in orientation or position being of no influence in this respect. By adopting the formal thermodynamical methods this annealed state [66, 67] can be exactly described by a set of fundamental quantities, such as the free energy F, the volume V, the chemical content μ_i* and the temperature T taken for each system-cell † belonging to systems under consideration. In view of the heterogeneity in such a system-cell which results from the variations in the chemical content μ_i, especially at the crystal boundaries, such a description of the system-cells will, in general, vary slightly from point to point. For that reason, the interiors of the crystal blocks will be considered as an aggregate of similar but not entirely identical system-cells.

It will be remembered from experimental determinations that the bulk specific volume and the level of energy of a metal attain different values after cold deformation, and this, by a change in mathematical description, indicates that the actually work-hardened places in the metal to which all the changes should be related are in a different thermodynamic state from that prior to deformation. In view of the chemical heterogeneity this state should be understood as another aggregation of system-cells in systems contiguous to the gliding and rotational planes.

The difference between the two states indicated by thermodynamic treatment has its counterpart in the structure and it is very convenient

* As it is possible that system-cells which differ in chemical composition may be described by the same set of values T, V and F, it is necessary to include a brief account of their chemical constitution.

† A small part of the whole system selected for consideration is called the system-cell. It has to be thermodynamically homogeneous.

indeed to picture the work-hardened state as portions of the crystallites with some distortion in the lattice. It may be mentioned, however, that other published conceptions of the nature of the changes resulting from cold-deformation, such as those of Beilby,(65) Tammann (50) and others, can supplement the thermodynamical view outlined above. This is of some value because consideration of recrystallisation will not depend on the changing views on the nature of work-hardening.

The progress of work-hardening can be interpreted from the adopted point of view as an increase in the amount of material in the work-hardened state at the expense of that in the annealed state. Thus the scale of reduction z and the regions of plasticity can be interpreted as indicating in a somewhat complicated manner the proportions of the two states. A metal which is cold-reduced to the highest possible degree should be in one state only, the annealed phase being completely absent. Unfortunately, however, such a desirable specimen cannot be easily produced, owing to the lowering of the recrystallisation range and to the increase in the temperature of the metal during working. Some idea of the properties of this state can be derived from a study of very heavily or medium cold-worked metals.

As the partly work-hardened metal consists of a mixture of two states, the bulk properties of it should be a function, in the case of so-called extensive * properties, of the sum of the products of quantities and of the mean properties of each state.

An Outline of the Theory of Recrystallisation of A. Krupkowski and M. Balicki.

From the moment when the process of plastic cold-deformation ceases, the cold-worked and annealed states are not in equilibrium with each other and the acting forces, but are under the action of a potential which tends to establish a new equilibrium. Under the given conditions, the thermodynamic equilibrium can be established only by a complete conversion of the less stable state, characterised by a higher level of internal energy into a more stable annealed state. This process of establishing a new equilibrium—the thermodynamic equivalent of recrystallisation—is governed not only by the magnitude of the potential, but also by the opposing resistances.

From the experimental findings (2) it has been deduced that, with the assistance of heat energy, the recrystallisation of a sufficiently small system-cell dV, chosen at random in a volume of the fully work-hardened metal, occurs in a somewhat abrupt manner and is complete. That is, the product is in the fully annealed state and not in one of the possible intermediate stages as suggested by the concept of "recovery." It has also been observed that recrystallisation does not occur simultaneously at all points in the metal, so that a partly annealed, previously heavily cold-worked metal forms a mixture, the components of which are the unchanged work-hardened state and the fully annealed phase of the same metal. These deductions simplify the problem to such an extent that it can be expressed as a formula. To explain the process, including the rate at which it proceeds, the theory (1) utilises the concept of activation (Arrhenius) and the concept of repartition of heat energy (Maxwell).

* As the term "additive properties" holds for three properties only, namely, for mass, for radioactivity and for volume in the case of dilute solutions, the alternative term "extensive properties" has been adopted (from J. S. Marsh (67)) to denote those of the constitutive properties which change linearly or almost linearly as a function of a constituent of a mixture. Properties changing in a manner other than linear will be called "intensive properties" and are not suitable for the quantitative determination of the constituents of a mixture.

The main formula which correlates the rate of recrystallisation, that is, the amount of metal which recrystallises in unit volume, with the temperature T and time t is:

$$\mathrm{Log}_e \frac{1}{1-x} = c \,.\, t \,.\, e^{-A/RT} \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad (1)$$

where A denotes the energy of activation, c the frequency coefficient and e the base of natural logarithms.

Where the temperature in the region of recrystallisation increases progressively, that is, when $T = n \times t + m$, equation (1) takes the form:

$$\mathrm{Log}\,\log \frac{1}{1-x} =$$
$$\frac{c \,.\, A}{R \,.\, n}\, e^{RT/A}\,(RT/A)^2\{1 - 2(RT/A) + 6(RT/A)^2 - 24(RT/A)^3 + \ldots .\} \quad . \quad (2)$$

where n denotes the rate of heating.

To apply the formula to a practical case it is necessary to know either the amount of work-hardened state which has recrystallised, x, or the values of A and c. As yet it is impossible to measure the value of x directly and it is only possible to obtain an approximate figure for it from the change in the "extensive" or "additive" properties. This can be done on the basis that the extensive properties of a mixture of states or phases is proportional to the amounts of and values for each constituent. The more practically extensive is the property chosen for the determination of x, *i.e.*, the less it is affected by factors operating during testing (such as temperature, work-hardening, &c.) and the less the constituents are affected by other phenomena (strain-age-hardening, crystal growth), the greater is the accuracy of the treatment. The formula for the calculation of x is a very simple one:

$$x = \frac{P_z - P_x}{P_z - P_a} \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad . \quad (3)$$

where P_z and P_a denote the values of an extensive property of work-hardened and fully annealed metal, and P_x is the value of that property of a partly recrystallised specimen.

On this basis, it has been possible in the case of copper and nickel to calculate the values of A and c and to reconstruct the change in some of the properties. The degree of agreement between the reconstructed curves and the experimental ones was very high and this is regarded as a proof of the soundness of every step in the treatment and of the theory.

Application of the Theory to Iron.

As iron on annealing reveals the same features as copper and nickel, namely, a common recrystallisation range, the theory can be applied to this metal, for this indicates that the changes in all properties are a reflection of the same transformation in the structure as that postulated by the theory. The application of the treatment to impure iron, however, cannot be very accurately done, because it is impossible to make corrections for the pronounced influence of strain-ageing and crystal growth which affect most of the extensive properties and thus cause the determination of x to be inaccurate. Therefore, the treatment in this case should be considered as an approximation only.

It has been decided to base the approximate calculations of x on the values of ultimate stress and plastic limit as given in Tables IV. and V. The values of work-hardened metal at 20° C. were substituted for P_z in equation (3), those corresponding to the temperature of the end of recrystallisation for P_a and the values determined after annealing to various

temperatures for P_x. With these derived values of x it has been possible to calculate the values of:

$$\text{Log}\log\frac{1}{1-x}+\log n-2\log T,$$

that is, all the members which depend upon the time and temperature in the slightly modified equation (2). By plotting these values against the inverse of the corresponding absolute temperatures a set of straight lines was obtained, the slope of which served for the calculation of both the energy of activation and of the frequency coefficients. The results are given in Fig. 23. It can be seen that the energy of activation appears to be independent of the reduction, just as in the case of nickel, and to be of the order of 31,000 cal. per mol. The probability coefficient depends upon the reduction in the characteristic manner resembling the trend in many

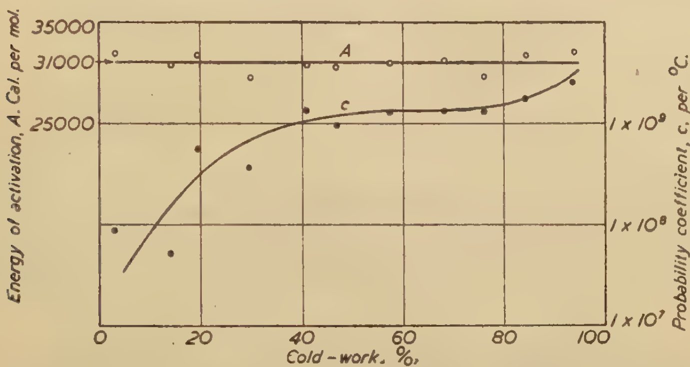

FIG. 23.—Energy of Activation (A) and Probability Coefficient (c) of Cold-Worked Armco-Iron Wire.

properties shown in Fig. 22. It is of the order of $10^7 - 10^9$ per degree C. Thus the numerical value of this coefficient mathematically depicts the metal and its past history.

Both these results support the adopted view on work-hardening. The constancy of the energy necessary to restore the annealed state indicates that the work-hardened state is essentially the same in a metal work-hardened to different degrees. The variation in the frequency coefficient, and the similarity between the course of this coefficient and the other properties should obviously be considered as an indication of the change in the amount of the work-hardened state. The fact that the energy of activation has been found to be practically the same for cold-worked copper (33,000 cal. per mol.), nickel (32,400 cal. per mol.) and iron (31,000 cal. per mol.) is highly interesting and indicates that the resistances which must be overcome during recrystallisation appear to be independent of the metal. Such a result, if confirmed on other metals, might be of vital importance for a strict theory of work-hardening.

To show that the recrystallisation of iron is also mathematically depicted by equation (2) and to show that the method of arriving at the values A and c was algebraically correct, an inverse operation has been performed. In this the values of x were calculated for any temperature by using the newly determined values of A and c. With x known it is easy to calculate by means of equation (3) the value of a property after annealing at any

assumed temperature. Some of the results obtained are shown in Fig. 24. The black dots in this figure represent the calculated data for the ultimate stress and hardness of the wires for the degrees of cold-work indicated, assuming the wires to be heated up at a rate of 960° C. per hr. The points marked by other signs represent experimental data from Tables II. and V. The agreement between the experimental and the calculated data is

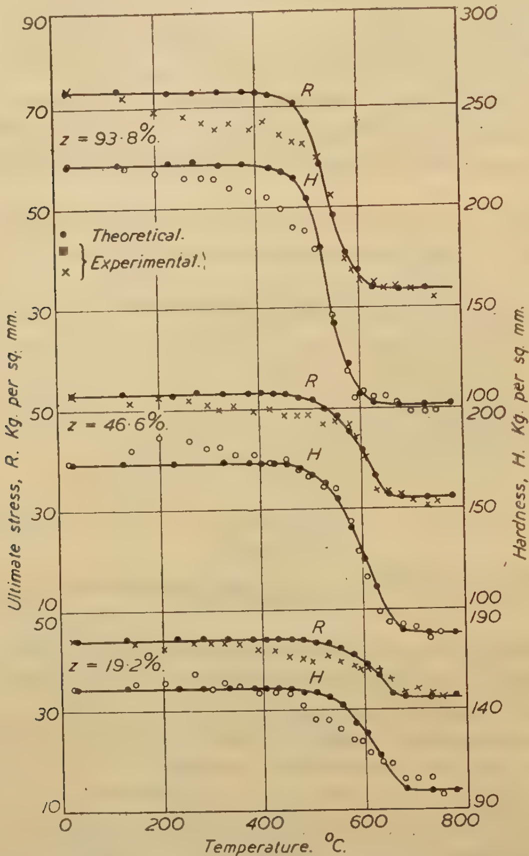

FIG. 24.—Ultimate-Stress and Hardness Curves for Cold-Worked Armco-Iron Wire from Calculated and Experimental Data.

satisfactory, although not as good as that found for copper and nickel, the maximum discrepancy being in the region preceding the main change, *i.e.*, it coincides with the region of red-shortness. From the hardness curve for 46·6% reduction, which shows quite good agreement even in this region, it may, however, be deduced that the discrepancy is of accidental character and that it may be one of the effects of strain-age-hardening.

The latter possibility has been investigated and the results are given in Fig. 25. For 67·9% reduction two values for the ultimate stress of unannealed work-hardened metal were available, namely, 55·0 kg. per

sq. mm. and 49·0 kg. per sq. mm., obtained on unaged material. Using both these values for calculation, two curves R' and R'' were obtained, the experimental data falling in between. This shows adequately that the discrepancy and inaccuracy of the treatment is highly affected by strain-ageing, which in turn is a result of using an impure metal for the work.[69]

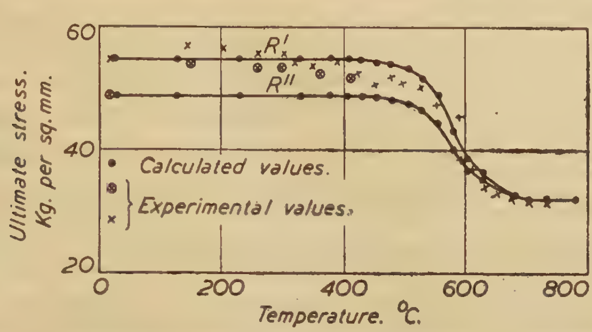

FIG. 25.—Effect of Strain-Ageing 67·9% Cold-Worked Armco-Iron Wire on the Calculated Ultimate-Stress Curve.

In view of the numerous difficulties frequently discussed elsewhere in the paper and clearly illustrated by the last treatment, it may be said that depicting the recrystallisation changes in iron with the aid of the theory of recrystallisation is not unsatisfactory and suggests that under more favourable conditions, or with appropriate corrections, still better accuracy might be obtained.

Acknowledgments.

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CORRESPONDENCE.

Mr. Roosevelt Griffiths (University College, Swansea) wrote: Having been associated with Dr. Balicki during the course of the work described in the paper, the writer would like to make a few comments.

In the first place tribute should be paid to the great amount of work done by the author and the way it has been presented. Whilst a fair picture of the experimental procedure and reasons for the conclusions are given, the whole of the experimental work has not been described, partly owing to lack of space.

Attention is drawn to valuable work carried out on the subject by various investigators in Poland. It is perhaps unfortunate that a complete account of the theory of Professor Krupkowski and Dr. Balicki has not been published in the English language. It may be added that Dr. Balicki has, with the writer's assistance, prepared a complete translation of the paper.

Considering the question of work-hardening of metals in general, there appear to be similarities in the effects with different metals, and yet with the same metal differences arise which are apparently associated with the treatment previous to working. It would seem that any theory and formula that aim at accounting for the complex phenomena of hardening must contain a common unifying expression as well as a factor to account for the differences of behaviour. The author has tackled the problem from

the angle of thermal stability with some success. The unifying expression that he has used is the "energy of activation" and the varying factor relating to the nature of the metal and its previous treatment is provided by the variation of the "frequency coefficient." This would seem a likely basis for a quantitative theory of work-hardening.

The author has put forward a simple conception of annealing, and his mathematical treatment of recrystallization is useful as a step in the right direction towards a complete understanding of this process.

AUTHOR'S REPLY.

The AUTHOR replied: Investigators of the problems of work-hardening and recrystallization will, no doubt, appreciate and adopt the critical view expressed by Mr. Griffiths. His postulate that "any theory and formula that aim at accounting for the complex phenomena of hardening must contain a common unifying expression as well as a factor to account for the differences of behaviour" may, without question, be accepted as axiomatic. Any working hypothesis, therefore, should and a quantitative theory must conform with this criterion. As Mr. Griffiths has pointed out, this is the case with the working hypothesis adopted for the purpose of this work. Before an attempt is made to develop this, or for that matter any other, hypothesis into a formal theory, however, it would be advisable to follow Mr. Griffiths's lead and examine it further by more such postulates. Doubtless this would help in the search for the much wanted synthesis of the maze of observations concerning hardening in general and work-hardening in particular. As one of those interested, the author would welcome a development of these difficult questions by other scientists as well acquainted with the subject as Mr. Griffiths, and possessing his rare ability of finding and formulating the essential postulates and requirements.

THE APPLICATION OF RADIOGRAPHY TO THE IMPROVEMENT OF FOUNDRY TECHNIQUE.*

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Paper No. 13/1945 *of the Steel Castings Research Committee (submitted by Mr. W. J. Dawson).*

(Figs. 30 to 61 = Plates XVII. to XXVIII.)

SYNOPSIS.

The principles involved in the taking and interpretation of a radiograph of a casting are given and the nature and extent of the defects which are revealed are discussed.

Examples are given of the application of radiography to the examination of steel castings and of the methods adopted to improve the quality of castings thus examined. Tests carried out on steel castings improved by these means have shown their strength to be much greater than had been previously recognised.

SECTION I.—INTRODUCTION.

WITH the increase of scientific control in the foundry, castings have become more uniform in character and, in these circumstances, alterations in practice can more easily be seen to have their effect. The direction in which these alterations should be made depends, however, upon the purpose for which the casting is required; some difficulty arises in defining a good casting and differentiating it from one which is not so good. The skin and general appearance may be improved, but until comparatively recently the internal soundness was difficult to assess and improve, mainly because of difficulty in examining for this feature. The fracture or cutting-up of a casting is expensive, destructive of the casting, does not permit the examination of the whole of the casting and does not guarantee that the next casting of similar design and manufacture will be of exactly the same standard. Using this method, however, certain types of defects may be found and techniques developed to overcome them. The advent of radiography has, however, enabled the examination of the internal soundness of a casting to be undertaken non-destructively and, usually, much more easily and quickly and with less expense.

The present paper is a summary of work carried out on the improvement of castings in a large foundry, the development being checked by radiography at each stage until a sound or serviceable casting was obtained. The variations in manufacturing technique were not novel, using as they did the established principles of moulding and other practices, but the rapid means of assessing the results by radiography enabled the working of the various alterations to be followed without the alternative expensive and cumbersome method of cutting-up. In addition, it was possible, by showing the moulder a radiograph of the results of his work, to enable

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him to realise perhaps more fully than he otherwise would the part which all his different operations played in arriving at a satisfactory casting.

A further point of value which has arisen from the work is the increase in the confidence with which castings can be regarded. It was formerly considered as axiomatic by the engineer that all forgings are sound and all castings are unsound, or at least are strongly suspect. It has now been demonstrated that castings can be made sound, that they can have an effective strength approaching that to be expected from a forging and that for certain purposes a casting may be better, cheaper, and more easily made than a corresponding forging fulfilling the same purpose.

SECTION II.—THE USE OF RADIOGRAPHY IN THE EXAMINATION OF A CASTING.

(1) *Basis of the Interpretation of the Radiograph.*

When X-rays are passed through an article they are partially absorbed by the material, and the intensity of the X-ray beam is reduced by an amount which depends upon the thickness of the article and the absorptive power of the material. The latter is related to the density of the material and also depends upon the type of X-rays used.

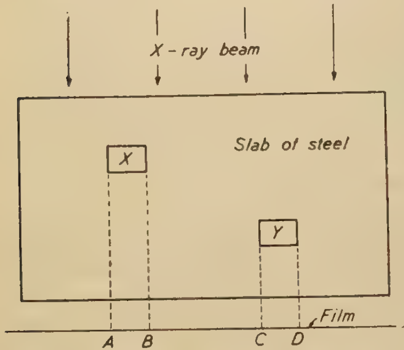

FIG. 1.—Method of Detecting Faults in a Block of Steel by Radiography.

The method of using X-rays to detect faults in a block of steel is shown diagrammatically in Fig. 1. The two holes X and Y are part of a section of the specimen through which the X-ray beam is passing to the photographic film. In each case the thickness of the steel to be penetrated by the X-rays at these points is less than that to be traversed in the remainder of the slab. As a result, the intensity of the rays leaving the slab below these positions is greater than that from other points of the slab, the places AB and CD on the film being more affected by the rays and thus appearing blacker when the film is developed. If the conditions were such that the specimen could be examined by means of a fluorescent screen, then these two places would appear brighter on the screen. Clearly, the greater the depth of the hole along the path of the rays the greater would be the blackening of the film or the brightness of the viewing screen. For steel specimens it is more usual to use the film method of examination, and, in this paper, only this technique is referred to, except for particular purposes, although, in general, the remarks made apply to both methods when used in their appropriate spheres.

If the two holes X and Y are now considered to be filled by a substance such as slag, the density of which is less than that of steel, and the absorptive power (more strictly, absorption coefficient, as defined in Appendix I.) of which is, therefore, less than that of the surrounding metal,* the intensity of the rays reaching the film at positions AB and CD is greater

* In certain cases the absorption coefficient may not be less if the wave-length of the X-rays is near one of the wave-lengths characteristic of the material of the slag. This does not, however, apply to the wave-lengths normally used in radiographing steel.

than that at other points, although still less than that reaching AB and CD in the case in which there are cavities at X and Y.

From these two illustrations the general basis underlying the interpretation of the radiograph is apparent. If the X-ray beam passes through a part of the specimen which has a total thickness less than that of other parts, the film below this part is blacker than that exposed under the remainder of the specimen, and the same holds true if the X-rays pass through a less dense part of the steel. From these arguments corresponding deductions may also be made that, if part of the radiograph is blacker than the rest, it must have been exposed below a section of the steel which was either thinner or which contained cavities or light inclusions.*

It would seem from the foregoing explanation that it would be an extremely simple procedure to determine exactly the variations in thickness or density which have caused a certain radiograph to be produced. That such is not the case, however, is soon found by practical trials, although the exact determination of these variations should be the aim of the radiologist. A few of the complications which cause this difficulty in interpretation are discussed later in Appendix I. Because of these complications and possible variations in technique it is possible to produce radiographs of unsound castings from which the castings appear to be sound; accordingly, a knowledge of the appropriate technique is essential in interpreting a radiograph.

(2) *Information Given by the Radiograph.*

It has been shown in the previous sub-section that a variation in the density of the radiograph is caused by an alteration in the thickness or density of the part of the specimen which is above the film. From the standpoint of castings these alterations in thickness or density can be assigned to several causes. It is the purpose of the present section to show the kind of deductions which can be made about the casting from an examination of the radiograph. The reproductions of the radiographs in this paper are of the same form as the originals. For example, a hole in the casting would appear in the radiograph as a dark area against a light background, and it would be shown in the same way in the reproduction. For clarity of reproduction it has been necessary, in some cases, to limit the area of the radiograph reproduced, in order to render the important parts more distinct; it must not be assumed, therefore, that the reproduction represents the whole area radiographed.

(a) *Shape of the Casting.*

The radiologist is not usually required to examine the internal shape of a casting unless this is part of a structure received in the built-up condition. Occasionally, however, the radiograph may give readily visible evidence of a shift or lack of alignment of cores. Normally, the radiologist knows the shape of the part, and has to devise a technique to examine it satisfactorily, and this may require specialised devices if the form of the casting is complicated.

(b) *Types of Internal Defects.*

Leaving out of consideration the variations in the thickness of the casting due to its shape, the defects in the casting which can be found and differentiated by radiography may be classified under the following headings:

(i) Threadlike shrinkage.
(ii) Shrinkage cavities.

* Certain anomalies to this statement have been found and have been ascribed to diffraction effects occurring in the material, owing to the orientation of the grains.[1, 2]

(iii) Blow-holes and gas-holes.
(iv) Slag inclusions.
(v) Sand inclusions.
(vi) Alien-metal inclusions.
(vii) Surface defects.
(viii) Cracks (including pulls).

This list covers the main types of defect which are found, but is not intended to be exhaustive, nor are the titles intended to define rigidly the limits of each type of defect. For example, the term "sand inclusion" is to be understood to cover not only inclusions composed of sand, but also all inclusions of the solid-mineral class. In general the vast majority of such inclusions consist of sand, and the class has, therefore, been given this title. The general features of each type of defect are discussed below and the defects illustrated when discussing means of eliminating them.

(i) *Threadlike Shrinkage.*—This is the form of trouble which is met with in specimens of approximately uniform section, when the skin of the casting has solidified and the contraction which takes place when the molten inner parts of the casting become solid has to be taken up by the whole of the casting. The last part of the casting to solidify is generally near the centre-line of the specimen, unless steps have been taken to remove it to some other part. The whole of the contraction must therefore be found adjacent to this centre-line, and the spaces which are devoid of metal will be spread at random throughout the area involved. For this reason the cavities may be of small cross-section, but may cover a considerable area, and are usually linked together to give an appearance of a string or thread network. Hence the name given to this type of defect. The depth of the streamers is not usually great, but there is considerable linkage between the various parts of the defect, which may cover an appreciable area and lead to leakage if pressure is put on the specimen.

(ii) *Shrinkage Cavities.*—This type of defect is concentrated in a few isolated holes and is usually found at a change in section; particularly is it troublesome at the centre of a T-section or X-section. This is caused by the thin sections solidifying first and feeding themselves from the thicker parts, which remain liquid for a longer time.

(iii) *Blow-holes and Gas-holes.*—Blow-holes occur on or just below the surface, and have a rounded appearance. Gas-holes have a somewhat similar form, but are also present in the inner parts of the casting. The differentiation by radiography between these two types of defect is by no means clear. If the distinction be drawn that blow-holes are due to some factors outside the steel, for example, a wet mould, the question is not really solved, because, according to Zapffe and Sims,[3] the defect is due to hydrogen from the moisture in the mould, and if this is small in amount it causes trouble only if there is already an almost critical amount of hydrogen present in the steel. For this reason, therefore, the two types of defect are dealt with as one, though if the cause of their occurrence is apparent, the method of overcoming the trouble becomes more obvious.

(iv) *Slag Inclusions.*—Although these may have a similar form to blow-holes or gas-holes they contain slag trapped in the metal in the molten state. They have a globular appearance, and may usually be discerned by their disposition throughout the casting.

(v) *Sand Inclusions.*—As already stated, this term refers to solid mineral particles which are trapped in the casting. They show as angular defects, unlike molten slag particles, but they may occur in similar parts of the casting.

(vi) *Alien-Metal Inclusions.*—The difference in density of these inclusions may be less than for the slag or sand inclusions and they may occur as light patches in the radiograph if a heavy metal has been entrapped. The history of the manufacture of the casting usually furnishes information regarding their identity.

(vii) *Surface Defects.*—The presence of surface defects does not require the use of radiography for their detection, but these defects, by their very nature, may have almost any shape and appearance in the radiograph, and they must first be excluded from consideration before conclusions are drawn as to the form of the internal defects. Should they occur on the face of the casting which is in contact with the film, there is the danger that for various reasons they may become exaggerated, and care must be especially exercised in radiographing specimens of rough or peculiar shapes in order to avoid the masking of internal faults by what are possibly minor surface defects. Whilst it is usually possible to say from experience whether a defect is internal or due to a surface mark, this is not always the case. The danger of confusing internal and surface defects is, however, readily overcome by always interpreting the radiograph in the presence of the casting from which it was taken. An example is given in Fig. 31 of a casting apparently containing a defect of considerable length. Closer examination of the radiograph showed that there was a slight gap at the mid-point of the defect. Examination of the casting itself revealed that the apparently most penetrating part of the defect, *A*, actually occurred at the surface, and would be machined away. The real internal defect, *B*, was only half the length of that which would be expected from a rapid glance at the radiograph. In practice, this internal defect was also within the machining allowance, and the casting was satisfactory. Had the whole mark on the radiograph been due to internal defects, the casting would have been scrapped.

(viii) *Cracks* (*Including Pulls*).—When a casting is cooling after being cast, the various parts contract, and stresses may be set up of sufficient intensity to cause rupture to take place. These ruptures, or discontinuities, can be divided into two classes, those which occur at high temperatures, when the material is in a semi-plastic state with low tensile strength, and those which take place at a lower temperature, during the later stages of cooling. In this latter case the strength of the material is higher, but the stresses are also increased by the additional contraction. The former type of rupture is called a "pull" or "hot tear"; the latter has been termed a "clink" or "crack." The pull is usually a more easily visible break and is also noticed more readily on the radiograph. In this discussion, however, the term crack is used to cover both forms of discontinuity, as it is not possible from the radiograph to distinguish between the two forms.

The recording of a defect on the photographic film depends, as has already been pointed out, on the difference in the thickness or density of the specimen at that place. In the case of a deep fine crack, the direction in which it is viewed will make a marked difference to the radiograph. It may be that, in the case of a fine crack which is normal to the X-ray beam, the alteration in total thickness of the specimen, due to the presence of the crack, is below the critical amount required to make the defect visible. For example, a crack 0·001 in. wide running in a plane at right angles to the X-ray beam, if occurring in a specimen 1 in. thick, would cause a change in density of the film less than that to which the eye is sensitive. In such a case the crack will not be seen. As the crack becomes more parallel to the beam, the alteration it causes in the thickness of the specimen increases, reaching a maximum when the crack is parallel to the direction of the beam. Even in this position, however, other considera-

tions, such as general scatter from other parts of the casting, come into play, and if the crack is fine and not deep such scatter may completely obscure the effect on the film. A survey of the subject has been made by Warren,[4] who found that the angle through which a deep crack can be seen depends not on the depth of the crack but only on the width. With very fine cracks the defect is seen only if the crack is in the direction of the beam, but with wider cracks there is a linear relation between the width of the crack and the angle through which it is visible. There have been cases in which hair-line cracks have been shown to be present by X-rays, but this is unusual. If the crack is shallow, however, the visibility depends not on the width or the depth, but on the volume of unit length. Other factors come into the problem when the usual type of irregular crack is viewed. The visibility then depends on the shape of the crack, as discontinuities are more easily seen than are gradual variations in density and a wavy crack is more readily revealed than a straight one. In this case also the changes in direction of the crack alter the angle which it makes with the surface and hence with the direction of the beam. There is, therefore, more chance of seeing the usual type of crack than a straight one of the same cross-section such as may be artificially produced for experimental purposes.

The conclusion must be drawn that if a radiograph shows the presence of a crack, then the crack exists, but if the radiograph fails to detect a crack, the inference must not be drawn that there is no crack in the specimen. In addition, the shape of the cracks is so irregular and the sizes are usually so near the critical range that it is inadvisable in general to attempt to draw definite conclusions as to their size or extent.

(c) *Severity of Defects.*

From the variation in the density of the radiograph which is produced under different conditions by the same specimen, as set out in Appendix I., it is apparent that a straight-forward interpretation of the radiograph in terms of the size of the defects, by simple reference to a calibration chart, is not yet possible, though such a conclusion should be the ultimate aim. The interpretation remains at the moment a matter for the expert, and the only way to acquire the necessary skill is to examine many radiographs in conjunction with the corresponding sectioned castings. (Several examples of such a comparison of radiograph and section are given in Section III.) These comparisons enable standards to be set up which are of value, although they are essentially subjective and allow at best only a qualitative estimate to be made. A method has been devised, however, which enables the expert to obtain a more objective and therefore satisfactory set of standards, and also to pass on the information he has obtained to others who have not had his experience. The method is based on the comparison of the radiograph with a series of sample radiographs showing a range of severity of the defects; the details are discussed in part (3) of this Section.

(d) *Area of Defects in Casting.*

The amount of a casting which can be examined by radiography is limited only by the capabilities of the apparatus and the time available for the examination. This extensive examination is one of the great advantages which the X-ray examination has over the older techniques for examining castings, for whereas these other devices, such as sectioning, enable the soundness of the specimen to be seen at one point, the X-ray method enables the whole of the casting to be examined, and the distribution of the defects may be observed.

This is information of value which can only be obtained at length and much expense by the other procedures, even where these are possible.

If a casting contains only a small number of defects, or the defects themselves are small, or if they occur in places where they would not normally be expected, it is possible to fail completely to find the unsoundness if resort is made to sectioning methods. The function of radiography in giving an overall picture of the soundness of a casting is seen in Fig. 30.* The radiograph was taken of a wheel-shaped casting which was afterwards sectioned to reveal the size of the defects which had been shown to exist in the outer flange. A first cut was made along the line AA', and, as shown in the photograph, no defect was revealed. Closer examination of the radiograph showed that the defect did not cross the line AA', although it was present in places throughout most of the circumference of the casting. After careful reference to the radiograph further sectioning at AB showed the unsoundness to be present, as shown by the arrow in the photograph of this section. On several occasions the sectioning of a casting has revealed no defects until, by the use of radiography, the positions of the defects have been found. The danger is obvious of relying upon the results of the sectioning of what must be a limited number of castings to give information as to the soundness of large numbers of similar castings. In addition, each casting which is sectioned, even if it were sound when cast, is made a waster by the method of examination.

(e) *Limitations of the Interpretation.*

A restriction to the interpretation of the radiograph should be mentioned here. It is possible at the present time to give an idea of the soundness of a casting from a reading of the radiograph of it, and in the future it may be possible to explain in strict quantitative terms the size of the defects in a casting. Such an interpretation, however, does not of itself enable a decision to be made as to the suitability of a certain casting for a particular function. The answer to such a question may only be given when the engineering requirements are specified, and is a problem for the engineer and designer to decide upon; the subject is beyond the scope of the radiologist as such. With the assistance of the metallurgist he can give guidance upon the subject, and some experiments in this direction are described later (Section IV.).

(3) *Standards of Reference for Defects.*

It has been emphasised in (2), above, that the interpretation of a radiograph is a matter for the expert, and that at present an accurate quantitative evaluation is not possible.

In the determination of the quality of welds the idea of standards was previously applied in the following manner: Radiographs were taken of a series of welds ranged in classes defined as good, borderline and unsatisfactory.[5] Interpretation was made by reference to these standard radiographs. The same method may be applied to the evaluation of the quality of castings, but for this purpose there are several difficulties, not the least being the fact that, for the method to be of general application, the techniques applied by the various operators using the standards must be exactly the same.

When used for reporting upon castings, however, the method has many advantages. It may be held that the best way in which to show the result of radiographing a particular casting is to present the radiograph with comments. The difficulties in making use of this method are

* The radiographs reproduced in this paper are printed to imitate the original radiograph, *i.e.*, the defects are shown black against a lighter background. The technique of printing the radiograph is discussed in Appendix II.

that the radiograph must be viewed by means of a special illuminator, which may not be available or which may be inconvenient to use at the place at which the report is being read. Secondly, if several people are interested in the results, then several radiographs must be prepared, as printing from a radiograph cannot reproduce the detail as it appears in the original. Further, the taking of several radiographs is obviously expensive of tube time, material and effort. Finally, the radiograph must be interpreted by a skilled person, since the unassisted layman may read an entirely false meaning into the film.

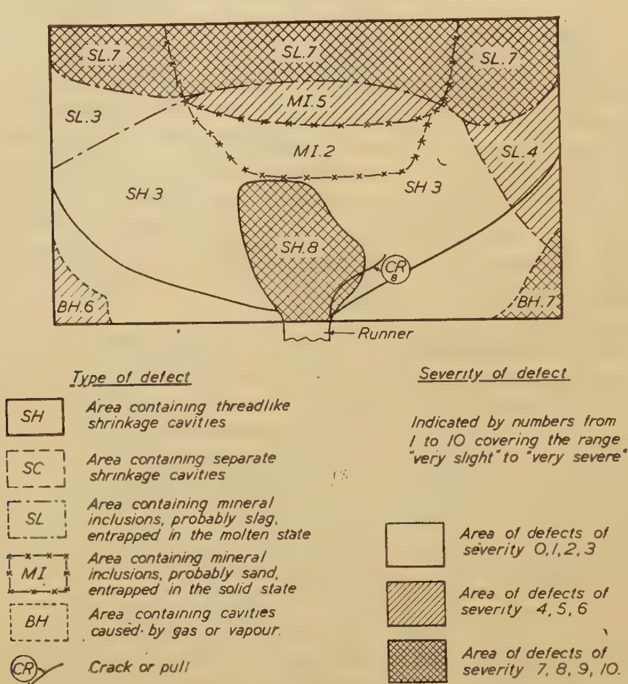

FIG. 2.—Method of Representing Typical Defects as Found in a Plate.

For these reasons a different method has been adopted. At present it is only applicable to castings of approximately uniform thickness, but it is hoped to extend the method later to other cases. It is based on a series of radiographs illustrating a range of severity of the type of defect being studied. Each type of defect has its own series of standards, but for convenience only one defect, threadlike shrinkage, is illustrated here. Fig. 32 shows the range of severity of defect from sound (*SH*.0) to unsound (*SH*.8) for plates which are $\frac{5}{16}$ in. thick. The last two places in the series, *SH*.9 and *SH*.10, are omitted to allow for defects of a severity greater than any at present encountered.

With the aid of these standards the problem of reporting the results of the radiographing of a plate is much simplified. Instead of producing a radiograph of the plate, with the attendant disadvantages already pointed out, or endeavouring to reproduce the appearance of the radiograph by a sketch, which is also unsatisfactory, it is a quick and easy process to draw

a map of the plate with lines encircling the areas of approximately uniform severity of a particular defect. The methods used to distinguish the different types of defect, and the appearance of a map of a plate containing various defects, are given in Fig. 2. Such a map conveys a good impression of the arrangement of the defects in relation to the relevant data, such as positions of runners and heads. By means of a number of these maps, drawn for plates made by different methods of moulding, it is possible to gather at a glance sufficient information to enable comparison to be made of them in relation to the specific methods applied in each case. In addition, however, if anyone is interested in the appearance of the radiograph, he can, by reference to his own set of standards, obtain an idea of the actual appearance of the radiograph without the difficulty of obtaining extra radiographs from the original casting.

The difficulty, mentioned earlier, of applying standards to all X-ray sets does not arise in their present application, as the standards obviously refer only to the one apparatus on which the work is being done. The order of severity of the defects, however, should be the same with all sets. It should be possible to relate standards from different sets of equipment by radiographing the same series of castings, but the radiographs obtained on one set may not, in general, be strictly interchangeable with those from other sets, because of the variation in the type of X-ray tube and equipment. This lack of interchangeability of different types of equipment is a drawback to the formation of the standards of rejection or acceptance mentioned earlier, but this trouble does not arise in the present use of standards for description of defects.

Section III.—The Application of Radiography to the Improvement of Foundry Technique.

In Section II. (2) the defects which are found in castings were classified according to their appearance in the radiograph. In considering the improvement of the foundry technique the same classification is adhered to, and each type of defect is dealt with in turn. Examples are given of the way in which difficulties due to each type of fault have been overcome.

(1) *Threadlike Shrinkage.*

This type of defect is caused by the contraction of the molten steel producing threadlike cavities, usually along the centre section of the casting. An example of severe shrinkage of this type is shown in Fig. 33, in which the radiograph is compared with sections cut from the positions marked. Trouble due to shrinkage may be avoided in two ways; either the shrinkage may be reduced, or it may be allowed to take place in some less important part of the casting. The first method may be accomplished in either of two ways: By casting at a low temperature, in which case other defects, such as cold laps or short-run castings, may be introduced, or by using, if possible, a steel which contracts less. The second method is to alter the effective shape of the casting, so that the important part of the specimen solidifies before the rest and is able to feed from the remaining molten steel.

The following examples are of castings which suffered from threadlike shrinkage, and the methods used to overcome its incidence are given.

Plates, $12\frac{1}{2} \times 8\frac{1}{2} \times \frac{5}{16}$ *in.*—It was found that these thin plates, when cast in a steel which behaved in a sluggish manner, were extremely unsound. Experiments were carried out by altering the heads and runners with the object of improving the soundness of the plate.

The plates were first cast in the usual economical way—by running the plate from a single runner in the middle of the long side, as shown in

Fig. 3, the plate sloping up from the runner at an angle of approximately 10° to the horizontal. No other precautions were taken to ensure sound-

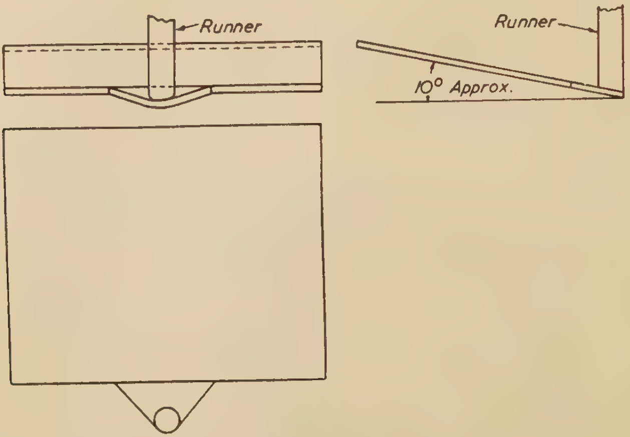

FIG. 3.—First Method of Casting Thin Plates.

ness; the result of the X-ray examination is given in Fig. 34. A schematic representation of this radiograph is given in Fig. 4, using the method of

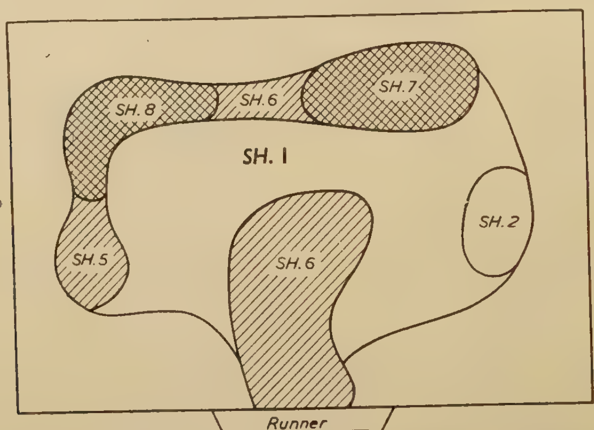

FIG. 4.—Schematic Representation of Radiograph shown in Fig. 34.

standards of reference already discussed. For convenience in reproduction, and to demonstrate more clearly the effects of the alterations of technique, this method of representation is adhered to in this discussion, as variations in radiographs may easily be altered in value, owing to the processes of reproduction. For greater clarity, only the effects on shrinkage are considered, other incidental defects being ignored. There was a considerable amount of shrinkage in the plates produced by the first method.

An adaptation of the technique was made by adding a "flow-off" on the side of the plate opposite to the runner, as shown in Fig. 5. The effect

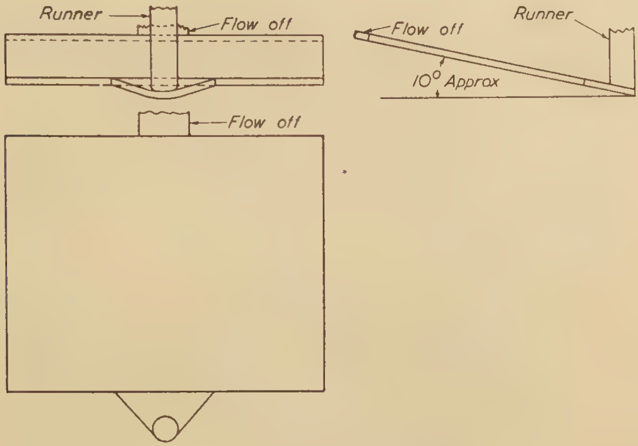

Fig. 5.—Second Method of Casting Thin Plates.

of this was slightly to reduce the amount of shrinkage, but not to alter its shape. The result is illustrated in Fig. 6.

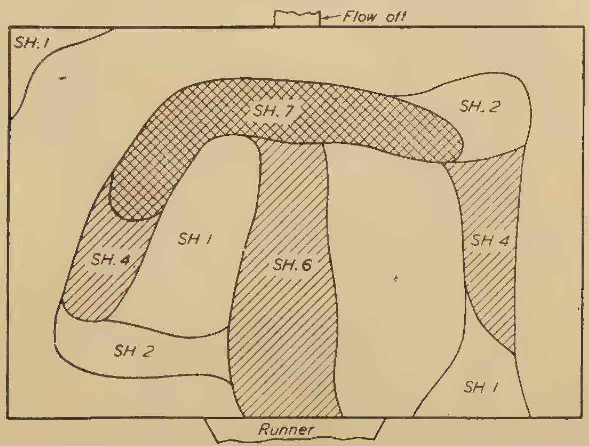

Fig. 6.—Schematic Representation of Radiograph of Plate Cast by Second Method.

It was apparent from these trials that feeding from the runner through such a thin plate was insufficient, and the addition of heads was next tried. A single head was placed on the plate, as shown in Fig. 7, and the number of ingates was increased to two, the casting still being run uphill at an angle of 10°. The effect of this change was a reduction of the amount of shrinkage and an alteration in the lines of flow, as seen in Fig. 8.

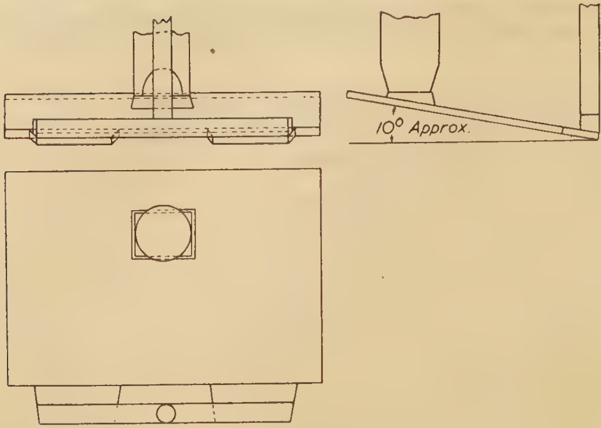

FIG. 7.—Third Method of Casting Thin Plates.

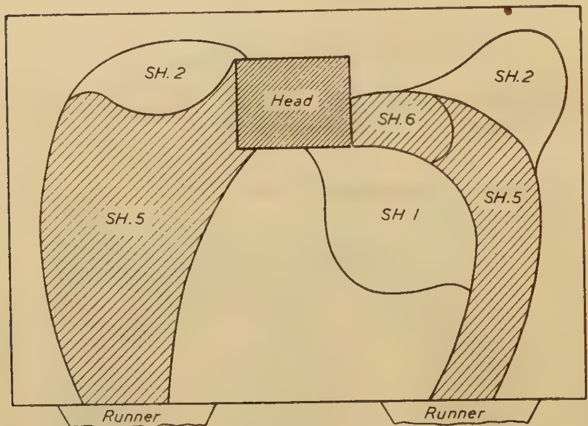

FIG. 8.—Schematic Representation of Radiograph of Plate Cast by Third Method.

An obvious extension of the same procedure was to put two heads on the plate instead of the one already used. This was done, as shown in Fig. 9; the form of the shrinkage was again altered and the amount reduced, as is seen in Fig. 10.

These plates were now reasonably satisfactory, and it appeared that an extension of the method to the use of four heads would still further have increased the soundness of the plates. From the practical point of view, however, an increase in the number of heads means an increase in the difficulties of fettling and might involve machining of the casting, so further experiments were made with the object of reducing the amount of fettling required. The means used was to put the heads as far as possible on the edge of the plate, so that a simple burning or shaping operation would suffice for their removal.

The first of these methods tried was that of casting the plate vertically with two heads on the top edge, these heads being used both for running

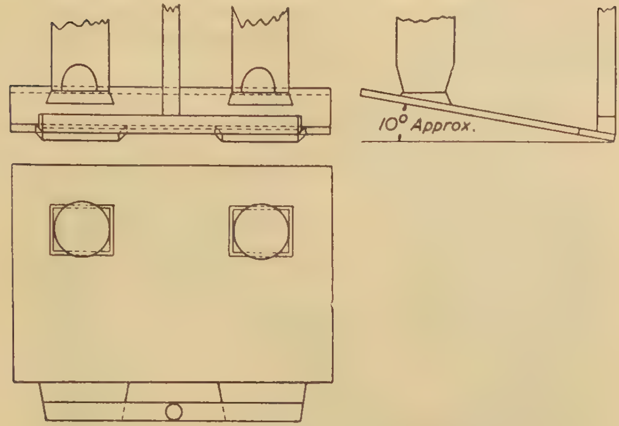

FIG. 9.—Fourth Method of Casting Thin Plates.

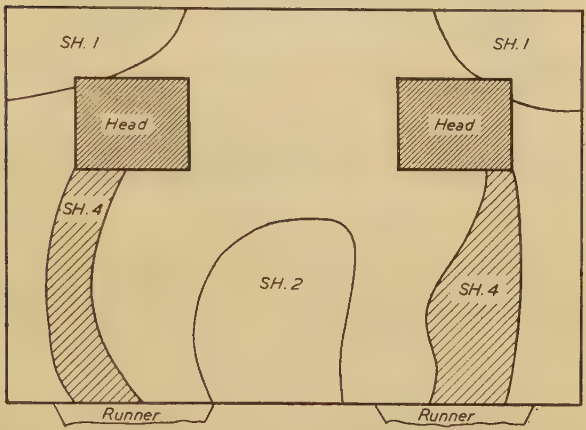

FIG. 10.—Schematic Representation of Radiograph of Plate Cast by Fourth Method.

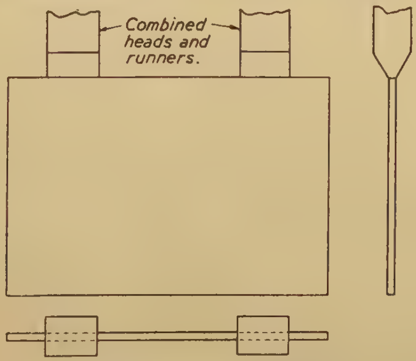

FIG. 11.—Fifth Method of Casting Thin Plates.

and for feeding. The method is illustrated in Fig. 11, and the position of the considerable amount of shrinkage is shown in Fig. 12. Feeding had again obviously been unsatisfactory.

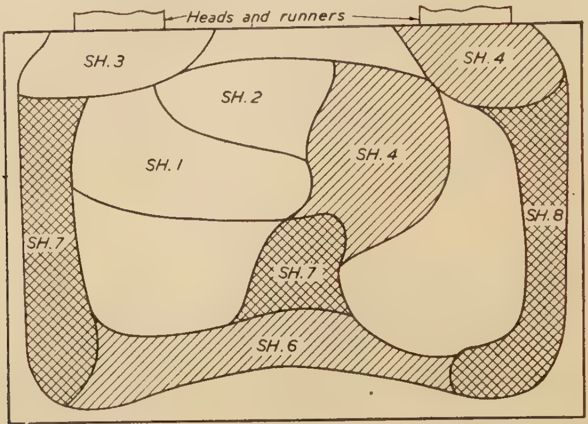

FIG 12.—Schematic Representation of Radiograph of Plate Cast by Fifth Method.

The second attempt was made using two heads offset from the edge of the plate—which was cast uphill at an angle of approximately 10° to the horizontal—the steel being run from two ingates at the opposite side

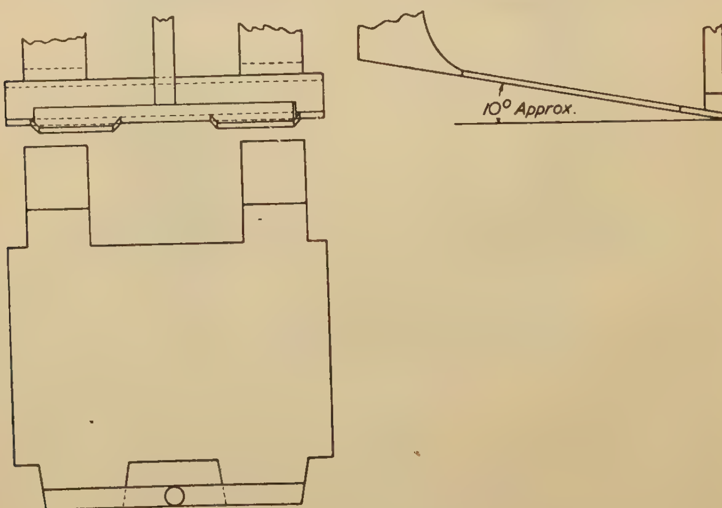

FIG. 13.—Sixth Method of Casting Thin Plates.

of the plate. This method was a slight improvement on the previous one, but much less satisfactory than the methods with the heads placed on the plate itself. The method of casting is given in Fig. 13, and the form of the shrinkage in Fig. 14.

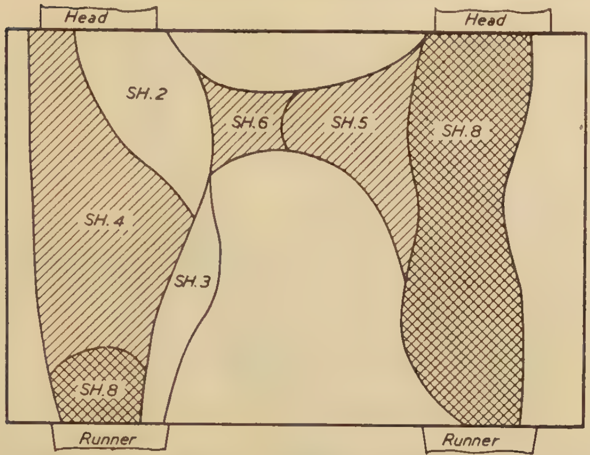

Fig. 14.—Schematic Representation of Radiograph of Plate Cast by Sixth Method.

A further trial of vertical casting was made with the plate run from the bottom, and with a flow-off on the top edge. This method is shown in Fig. 15. The amount of shrinkage which resulted was only a slight improvement on the previously vertically cast plates; the character of the shrinkage is shown in Fig. 16.

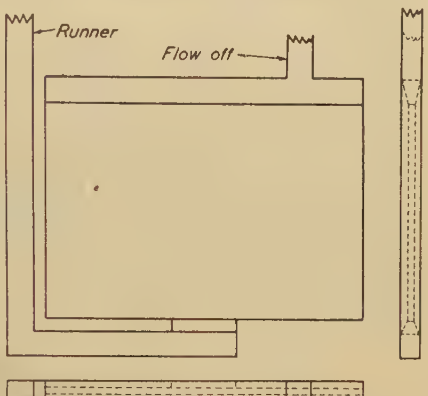

Fig. 15.—Seventh Method of Casting Thin Plates.

In this series of trials the plates were cast over a range of temperature by each method, and the temperature was found to be of minor effect in producing alterations in the position and severity of the defects. It was not possible to control the temperature within fine limits, because of the practical difficulties involved in fitting these trials into the normal running of the shop, but the results given are typical of each method of casting and serve to indicate the improvements to be obtained by a change of moulding practice.

The steel on which these experiments were carried out was a highly alloyed nickel-chromium type which is rather difficult to cast, and for this reason it enables the effects of the different casting techniques to be more easily observed. Other steels were tried, using the most satisfactory moulding method, *i.e.*, casting uphill from two runners with two heads on the plate, and all of these gave sounder plates than those previously obtained. There was no obvious relation between the composition of the steels tried and the soundness of the resulting plate. The possibility of correlating such properties as the fluidity to the castability of the steel, as evidenced by the soundness of plates cast under these conditions and examined by radiography, should, however, be borne in mind.

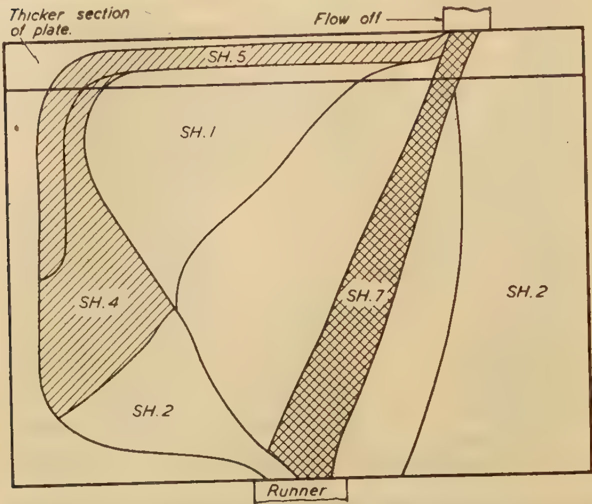

FIG. 16.—Schematic Representation of Radiograph of Plate Cast by Seventh Method.

(2) *Shrinkage Cavities.*

As this type of defect is usually found at a change of section, two possible methods of reducing the cavities are available. The first is to ensure that the thick sections solidify at the same time as the thin sections, and so cannot be used for feeding; the second is to assist the thick parts to be fed in turn from some other liquid reservoir, for example, a head. Some examples are given of castings which suffered from these defects and which were cured by the application of these principles after the cavities had been located by radiography.

(a) *Fan Blades.*—These blades were approximately ½ in. thick at the root, which formed an X-section and which tapered away to thicknesses of ¼ in. and zero at the tip of the blade. The type of casting is shown in Fig. 17. To produce a defective casting for experimental purposes, two

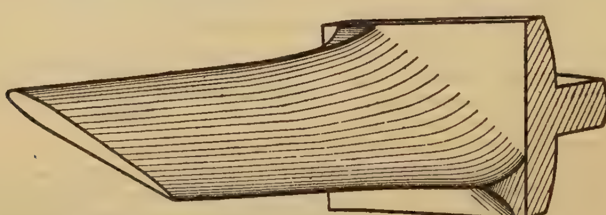

FIG. 17.—Fan Blade.

blades were cast with no heads, and, as expected, a cavity was found in the centre of the X-section as shown in the radiograph, Fig. 35, and the photograph of the corresponding sectional casting, Fig. 36.

With the thickness of the upper part of the X-section increased and a head placed over it, as in Figs. 18 and 19, the section was able to feed from the head and the cavity did not appear.

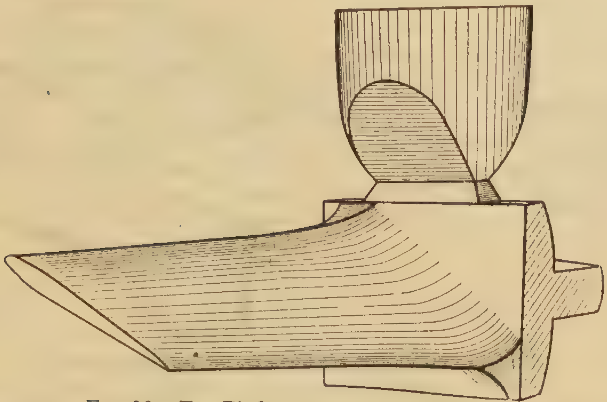

FIG. 18.—Fan Blade, showing position of head.

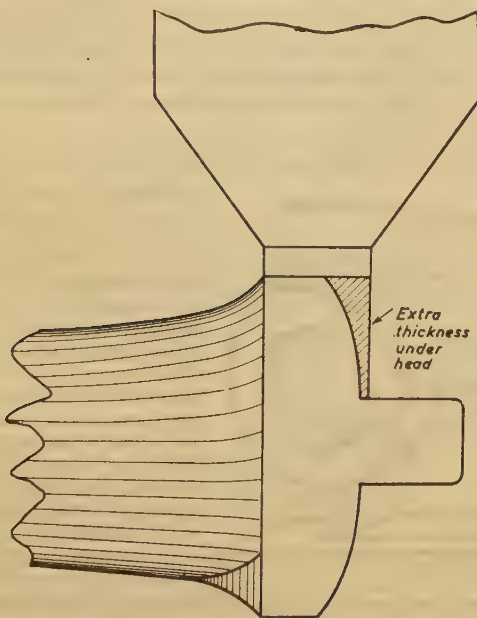

FIG. 19.—Fan Blade, showing section thickened to assist soundness.

(b) *Small Brackets.*—The same method of attack was used to eliminate cavities from this type of small casting, a drawing of which is given in Fig. 20. Although only 3 in. long and weighing about 4 oz., great difficulty was found in obtaining the required degree of soundness. Fig. 37 is a radiograph of one of these early attempts, and it will be seen that there are cavities in the centre of the little boss and unsoundness in the semi-circular yoke.

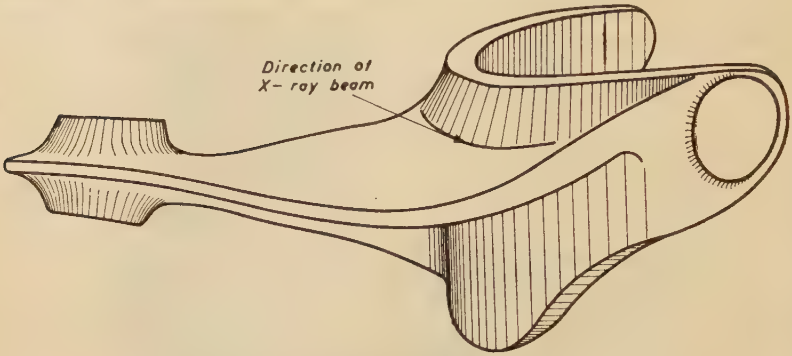

FIG. 20.—Small Bracket Casting, in position as cast.

The casting was made in the position as shown in Fig. 20, the metal being run into the rear side of the central connecting web. The cavity in the boss was eventually overcome by increasing the height of the boss so that unsoundness was removed from this region, and this was further assisted by placing a small flat chill under the boss. The unsoundness in the yoke was removed by increasing the thickness of its section and by placing a head above it. These precautions resulted in the sound castings shown in the radiograph, Fig. 38.

(c) *Pivot-Bracket Castings*.—Part of this casting consists of a boss of approximately $2\frac{1}{2}$ in. in dia. and 2 in. long, which has a $\frac{1}{4}$-in. thick web meeting it along one side and a $\frac{1}{4}$-in. thick rib crossing both boss and web at right angles to the web, and occupying over half the circumference of the boss. Fig. 21 is a sketch of this part of the casting. To reduce machining as much as possible, the part was first made to the approximate finished dimensions, with a core in the bore of the boss. Fig. 39 is a radiograph taken through the boss from an angle when this coring method was used in making the casting. It will be seen that there is a definite cavity or unsound area at the junction of the boss with the web and rib. The next attempts were made with chills placed round this position, but the results were not too successful, and eventually the boss was made solid with an L-head to ensure feeding. The radiograph, Fig. 40, shows the appearance of a bracket cast with the boss solid and then rough-bored to facilitate the taking of the radiograph. The previously unsound portion is seen to be quite solid and no further trouble was encountered owing to this defect.

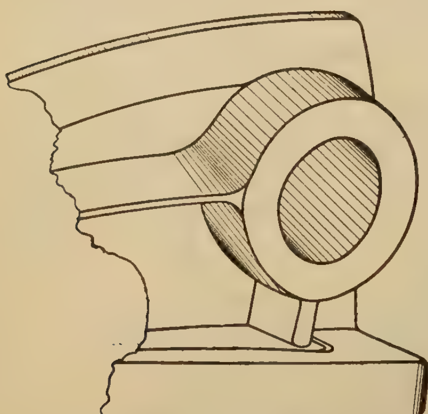

FIG. 21.—Boss of Pivot-Bracket Casting.

Although, in this case, the method of chilling was unsuccessful, it has been used to advantage in other places on this same casting. This same boss was separated from the base plate by an X-section $1\frac{1}{4}$ in. long, formed

by two intersecting ribs, each ½ in. thick, and for a considerable time a frequent fault was the appearance of a cavity running through the greater part of the centre of this X-section. It was impossible to increase the thickness of the ribs, owing to difficulties in machining the excess metal away. Neither was it possible to add an extra head to the base plate to assist feeding, as this would have required an entirely new arrangement for the casting of the bracket. The use of chills was first tried by placing them against the ribs, under the boss at *A*, as shown in Fig. 22. The

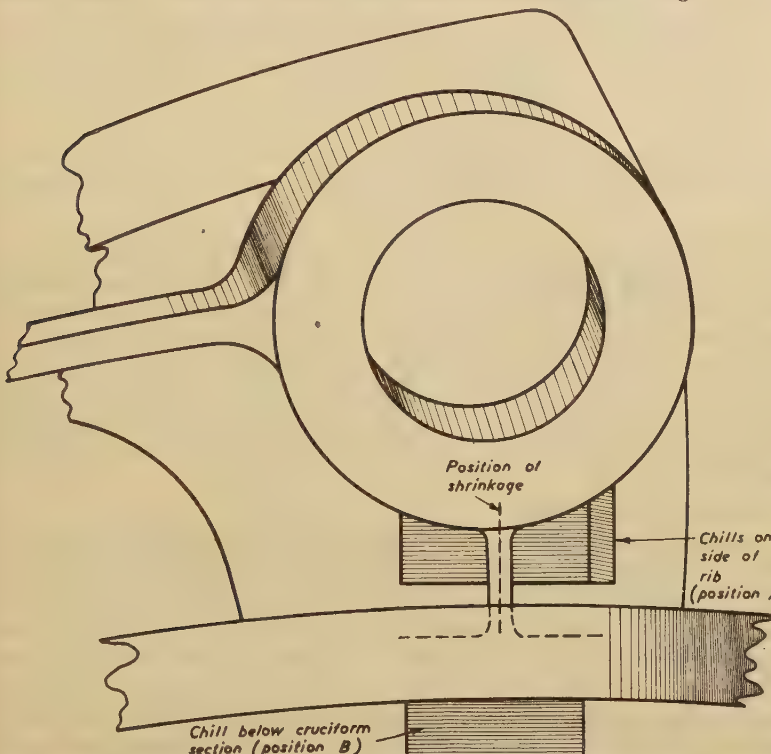

Fig. 22.—Boss of Pivot Bracket, showing positions of chills. Chills at *A* failed to cure unsoundness; one at *B* was successful.

method proved unsuccessful. A further attempt, also shown in Fig. 22, was then made by applying the chill to the underside of the base plate, at the root of the X-section (position *B*), when an immediate improvement was found. The use of the chills along the section apparently sealed off the X-section both from the base plate and the boss, so that, while solidification was assisted, it was not possible to feed the section from either end, and the cavity persisted. When chilling took place from the base-plate end of the section, solidification was speeded up, but it was possible for the section to be fed from the liquid steel in the boss and its adjacent head, with a resulting improvement in quality. The appearance of the section as radiographed before the use of the chill on the base plate is shown in Fig. 41. This defect disappeared completely when the chill was introduced.

A further way in which chills were used on this bracket can be seen from Fig. 23, which shows a section through a second boss and the plate

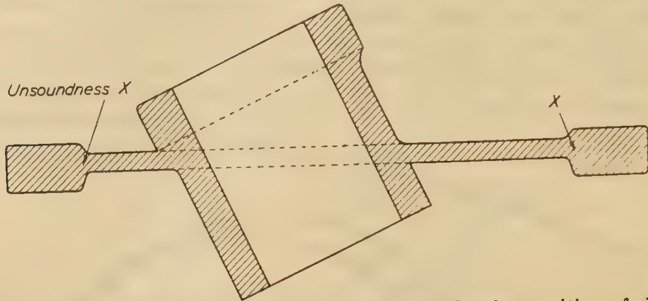

FIG. 23.—Section through Boss and Surrounding Plate, showing position of shrinkage near beading surrounding plate.

through which it projects. The bulk of the plate has a thickness of $\frac{1}{4}$ in., while the beading round the edge is $\frac{5}{8}$ in. in thickness; the whole plate had to be run through the thin section and fed from the boss, which again was made solid. The result, as would be expected, was unsoundness in the junction between the thin and the thick parts of the plate; this is shown in the radiograph, Fig. 42. A chill was placed on the thick beading of the plate, and this was successful in solidifying the thick section while, at the same time, allowing feeding to take place through the hotter thin section from the reservoir of liquid steel in the boss. The efficiency of the method is clear from the radiograph, Fig. 43, taken after this modification had been introduced.

(3) *Blow-holes and Gas-holes.*

During the manufacture of a number of experimental rectangular plates it was necessary in one case to complete the moulding of two of the corners by hand-tooling. The sand in these corners was made rather damp and it was suspected that defects might occur. The blow-holes produced in the plate are shown by the radiograph, Fig. 44, and the cause of this trouble is clear from the history of manufacture of the plate.

One heat of fan blades was suspected of being unsound, though no surface defects were visible, and a radiograph was taken of the first and last castings made. The result is shown in Fig. 46. A photograph of a section of the last casting, the more unsound one, is given in Fig. 45, for comparison. An investigation revealed the cause of the trouble, and showed that the moulding practice was not at fault. The error was corrected and no further defects of a similar type were found.

Cap Castings.—These small plain carbon steel castings were giving trouble when they came to be machined, owing to general porosity. The castings were made in groups of three from one runner, and a typical defective casting is shown in the radiograph, Fig. 47. These castings were apparently quite sound on the surface, and only radiographing or machining disclosed any defects. Greater care was taken with the steel for the next batch of castings, and, using the same moulding technique, the castings were much sounder, as shown by the radiograph, Fig. 48. The elimination of the former defects, however, led to shrinkage cavities in the lugs. By the addition of a chill below the lugs, the cavities were removed. At the same time, brackets were added to the lugs to allow any air trapped there to escape to the central head.

(4) *Slag Inclusions and Sand Inclusions.*

With high-quality castings (and these are the type most frequently radiographed) sufficient care is taken in the preparation of the moulds to ensure that little incidence of slag and sand inclusions is likely to occur. The value of efficient skimming, or, preferably, of bottom-pouring from the ladle, is well known, as also is the use of slag traps for preventing slags from entering the casting itself. On the other hand, faulty moulding may result in the trapping of loose sand in the casting. Although the character of the mould may apparently remain the same, it is possible to have castings made from it of widely different quality. This is clearly shown in two examples taken from examinations made at the request of the Foundry Practice Sub-Committee. Two plates, each 2 ft. square and 10 mm. thick, were cast from the same heat of plain carbon steel made in a 10-ton basic electric-arc furnace, and were bottom-poured into identical moulds of the pattern shown in Fig. 24. The bottom 6 in. of

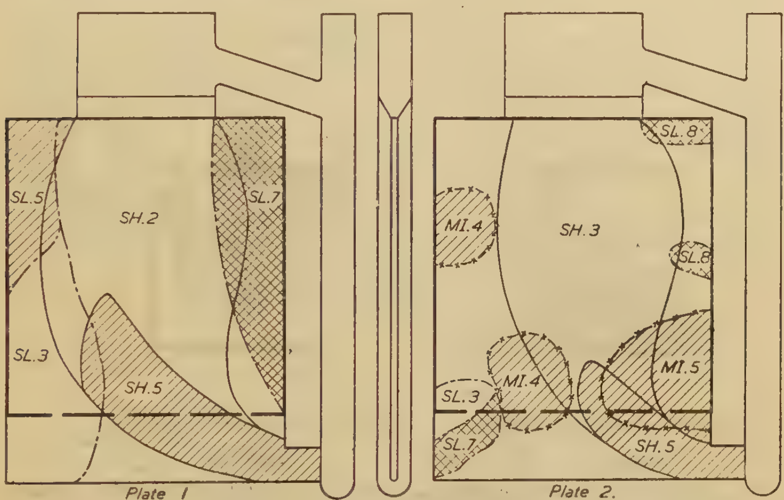

FIG. 24.—Method of Moulding and Diagrammatic Representation of Results of Radiographic Examination of Two Plates 2 ft. sq. × 10 mm. thick.

the plate "as cast" is intended as discard. It will be seen from the representation of the defects found to be present that, although the nature of the mould has influenced the type and location of the shrinkage present, this being arranged almost identically in both cases, there was a considerable difference between the amounts of slag and sand inclusions present. In plate 1 there appears to have been a considerable amount of slag introduced into the mould at the beginning of the pouring. This slag collected on the sides of the mould. The steel cast at the end of the pour filled the mould, passed through to the head and swept the slag with it into the head. In the case of plate 2, however, it would appear that the sand of the mould was eroded by the steel. Eventually, some of it broke away towards the end of the pour and was carried into the base of the mould too late to allow it to be removed upwards to the head. This type of defect is, of course, always liable to occur when the steel has to fall through such a distance as was the case in this instance.

In the second investigation, the difficulties of avoiding inclusions were again demonstrated. Fig. 25 gives the results of three methods of casting

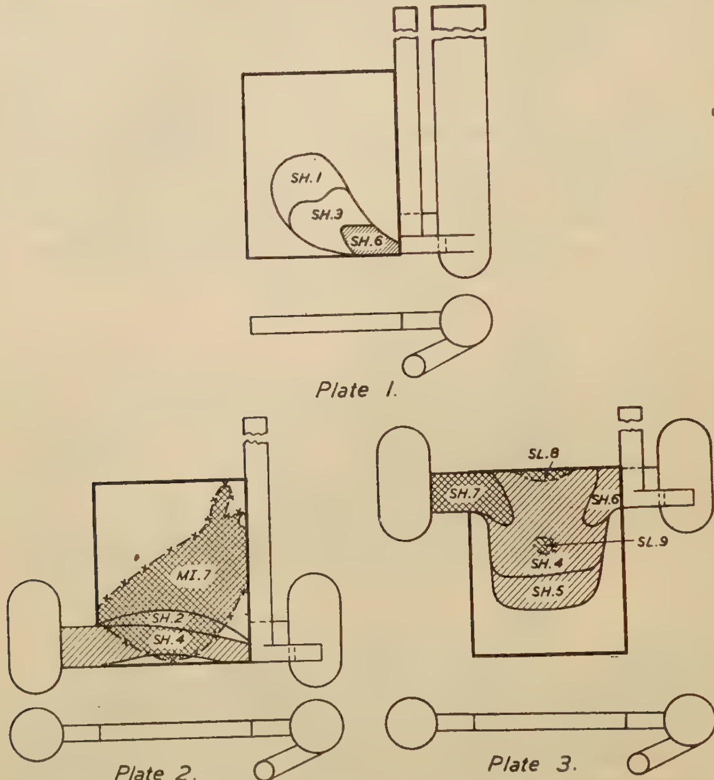

FIG. 25.—Method of Moulding and Diagrammatic Representation of Results of Radiographic Examination of Three Plates 12 in. × 10 in. × 1 in.

plates 12 × 10 × 1 in. in plain carbon steel from a high-frequency furnace. Plate 1 was cast using an open whirl-gate head extending above the top of the plate. Shrinkage was quite small and no other defects were found. Plate 2, however, which was cast with two heads (one of which was run in a similar manner to the head in plate 1, but which were smaller and "blind"), although showing little shrinkage, revealed an extensive area of severe sand inclusions. This was, no doubt, the added effect of more intricate mould design leading to a less robust mould and of a closed top to the head preventing the eroded sand from rising far enough away from the swirl of steel flowing into the mould. It would seem that, once again, the inclusions were put into the mould at a late stage in the casting, as they were clearly introduced from the head which was also the runner, and they had not had time to escape up the plate before freezing occurred. This was borne out by plate 3, which employed two similar heads, but located at the top of the plate; in the latter there were again slag inclusions, but as they were introduced towards the end of pouring they were able to ride on the top of the steel. In this case they

only appeared as a small area at the extreme top of the plate, with the exception of another small area apparently introduced about halfway through the pour. It is clear from the diagrams that, in this case, the heads had little to do with feeding the plate, as the shrinkage was much worse than in the case of either of the other two plates.

(5) *Alien-Metal Inclusions.*

Trouble due to the inclusion of other metals in the casting of steel is infrequent, and will not be discussed here.

(6) *Surface Defects.*

Since radiography is not required for the detection of surface defects, the methods of improving surface finish are outside the scope of this paper.

(7) *Cracks and Pulls.*

As already mentioned, the absence of cracks as shown by the radiograph is not a certain guide to the absence of all cracks; fine cracks may easily be missed.

The use of radiography has been found to be of value in a number of cases in which the surface appearance of the crack is deceptive. It may appear visually as a shallow cold lap or wrinkle in the metal, but may be shown by the radiograph to be of considerable depth. The elimination of the crack type of defect is one of the usual foundry problems and is not restricted by any means to defects only revealed by the radiograph. However, it may be of interest to illustrate the way in which a particular defect of this kind was overcome.

A pull was located at the root of a T-section of the type shown in Fig. 26. As the arms *A* and *B* of the T were part of a tubular section of

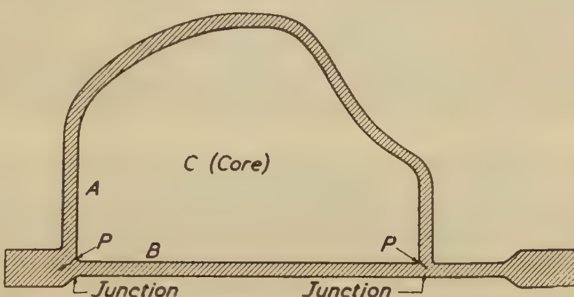

Fig. 26.—Section of Casting which produced Pulls as shown.

the casting, any failure of the core *C* to collapse at the requisite time led to the formation of a pull *P*. The appearance of one of these defects on the radiograph is shown in Fig. 49. The adjustment of the strength of the core to be such that collapse takes place at the right moment is by no means easy, and, while work was progressing to produce the right kind of core, various experiments were tried with chills placed round the T-section. If the hot spot at the centre of the T-section could be removed, the solidification of the steel surrounding the core would proceed more uniformly. The contraction stresses of the casting would be spread more evenly over the steel round the core, with the result that there would be a smaller risk of these stresses exceeding the strength of the steel in its weak condition immediately after solidification.

The first method tried was that of placing a chill below the T-section, as shown in Fig. 27 (*b*). This gave no improvement.

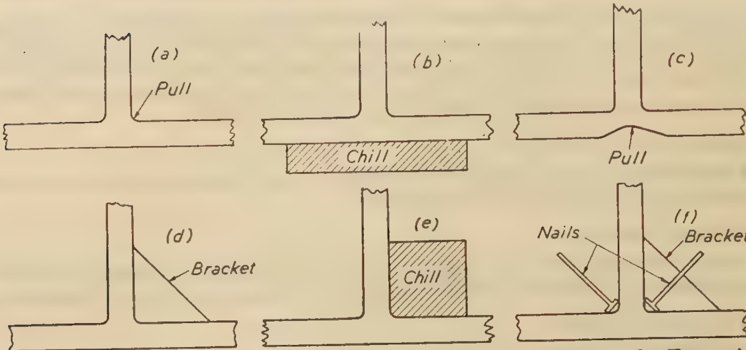

FIG. 27.—Diagrammatic Representation of Methods tried to Prevent the Formation of Pulls at a T-Junction.

The chill method having failed, the attempt was now made to equalise the thicknesses round the T-section by putting a channel in the base of the T, as shown in Fig. 27 (*c*). The temperature distribution was apparently not made uniform, and the weakening of the section caused an even more pronounced tear to be produced at the bottom of the groove. The use of brackets cast between the two arms of the T-section was tried, as in Fig. 27 (*d*), in the hope of strengthening this part to overcome the contraction; this also was unsuccessful. Placing chills inside the casting at the place at which the pull normally occurred (see Fig. 27 (*e*)) proved to strengthen this part too much, the weakness being transferred to a different part of the casting. Clearly a chilling effect less severe than this was needed, and so the solid chill was replaced by a row of nails, as shown in Fig. 27 (*f*), and the junction was also strengthened by means of brackets. The success of this method was at once apparent, and the result can be seen from the radiograph, Fig. 50, taken before fettling in order to show the positions of the nails and brackets.

SECTION IV.—THE EFFECT OF INCREASED SOUNDNESS ON THE STRENGTH OF A CASTING.

The previous Sections of this paper have been confined to an explanation of the way in which defects in a casting can be revealed by radiography, and having been found, how they can be reduced or eliminated. It has been shown that in the case of thin plates the production of a perfectly X-ray-sound casting is not readily effected. The question naturally follows as to the extent to which unsoundness as revealed by the radiograph can be allowed to occur in a casting without impairing its satisfactory service.

Several ways of determining the suitability of a casting for a particular purpose are possible, and in the study which has been made of radiographed castings, some light has been thrown on the advantages or disadvantages of the different methods. The satisfactory casting is the one which performs efficiently the function for which it was designed. Unfortunately, methods based on this definition cannot be used for inspection, as tests in service are usually long drawn out and may be affected by conditions other than the variations in the casting itself. A test to destruction reveals the conditions which would produce failure, and can be carried out much more rapidly than service tests; it does not reproduce service conditions, however, and it also destroys the casting. The latter disability can be overcome by a proof load, but the excess load over that

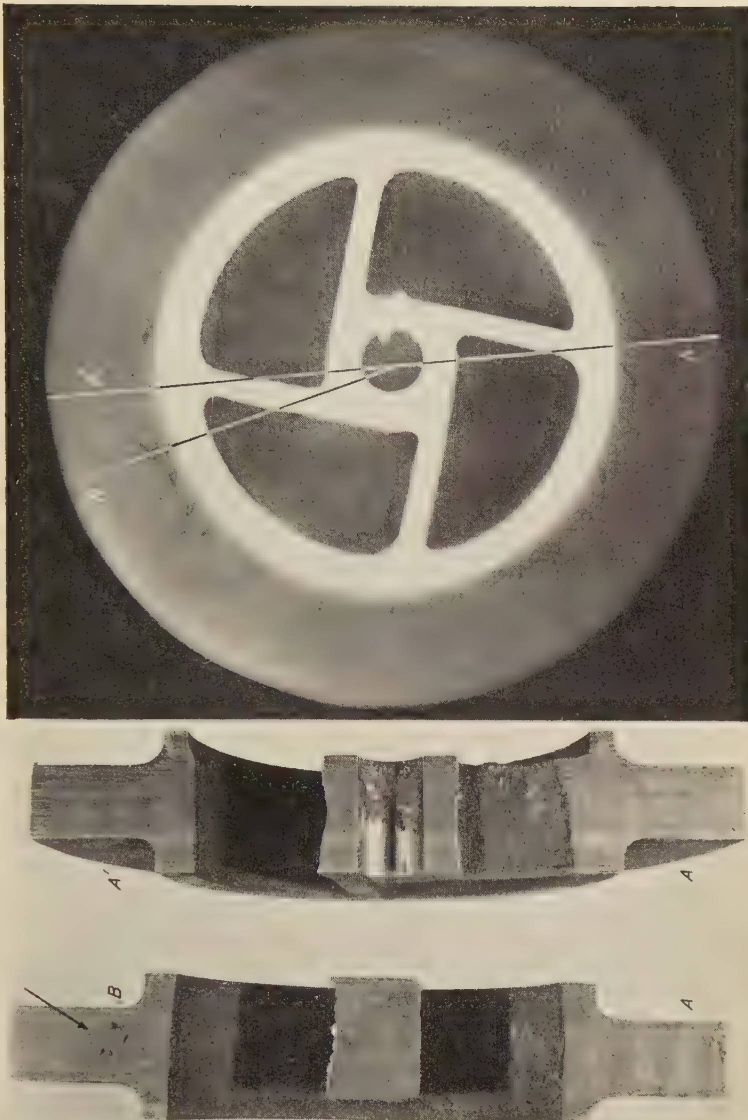

FIG. 30.—Radiograph of Casting, with Photographs of Sections Taken at Positions Marked. First cut, AA', revealed no defects. Second cut, AB, made after reference to radiograph, revealed defect at B. × [illegible].

[*Jackson.*

[*To face p.* 248 P.

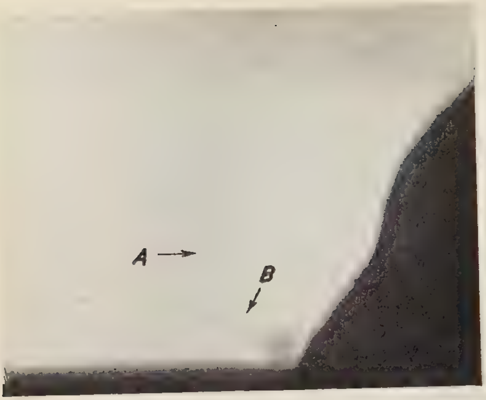

FIG. 31.—Radiograph of Casting Apparently Containing Extensive Internal Defect. *A* was a surface depression; *B* was internal. × 1.

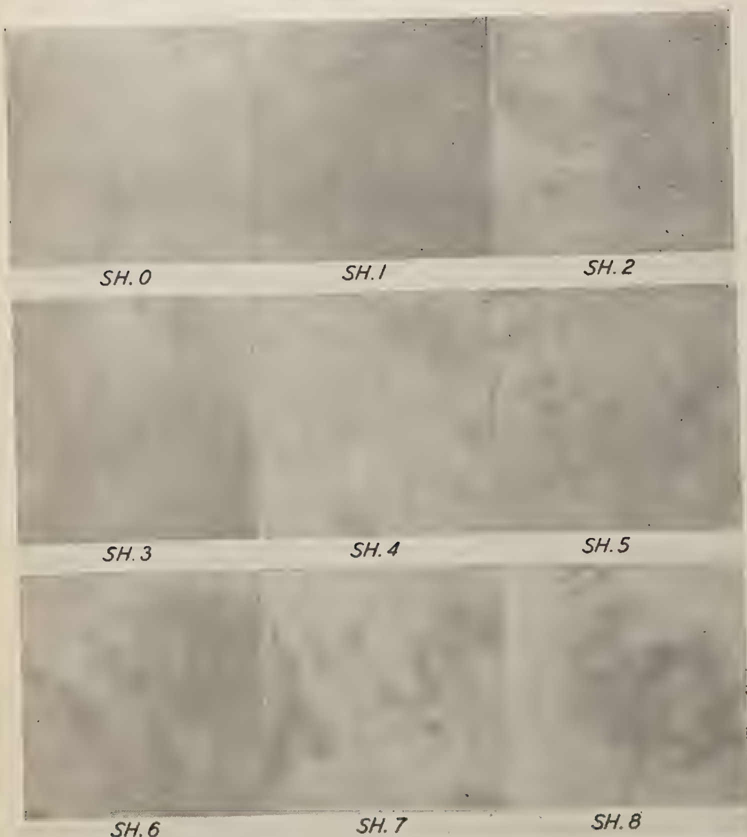

FIG. 32.—Radiograph Standards of Threadlike-Shrinkage Defects for Plates $\frac{5}{16}$ in. in Thickness.—*SH*.0 (sound) to *SH*.8 (unsound). *SH*.9 and *SH*.10 reserved for more severe shrinkage not yet encountered. × $\frac{2}{3}$.

[*Jackson*.

FIG. 33.—Comparison of Radiograph of Plate Showing Severe Threadlike Shrinkage with Photographs of Sections Taken in Places Indicated by Lines on Radiograph. Sections 1 and 3, after polishing and etching; sections 2 and 4, after polishing only. × 1.

FIG. 34.—Radiograph of Plate Cast by First Method (see Fig. [illegible]).

Jackson.

FIG. 35.—Radiograph of Unsound Fan Blade, showing cavity at centre of cruciform section. $\times \frac{2}{3}$.

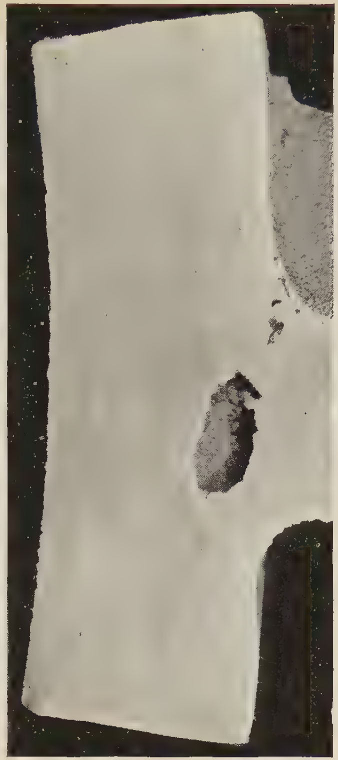

FIG. 36.—Fan Blade, sectioned to reveal defect shown in Fig. 35. $\times \frac{2}{3}$.

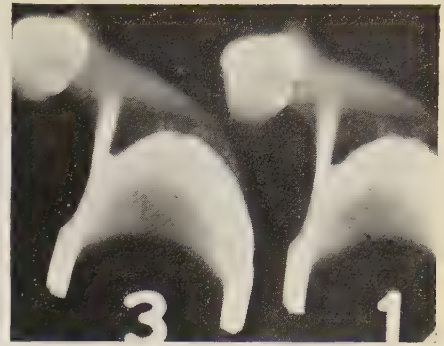

FIG. 37.—Radiograph of First Trials of Small Bracket Castings, showing unsoundness in yoke and boss. $\times \frac{1}{2}$.

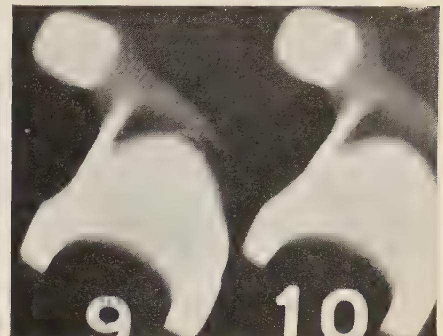

FIG. 38.—Radiograph of Small Bracket Castings, after modification of moulding technique. $\times \frac{1}{2}$.

[*Jackson.*

FIG. 39.—Radiograph through Boss of Pivot Bracket made with Core, showing unsoundness. × ½.

FIG. 40.—Radiograph through Boss of Pivot Bracket. Boss was cast solid, and rough-bored. No unsoundness. × ½.

FIG. 41.—Radiograph of Cruciform-Section connecting Boss and Base-Plate. Unsoundness revealed. × ½.

FIG. 42.—Radiograph of Plate and Boss, showing unsoundness at positions expected from Fig. 23. × ½.

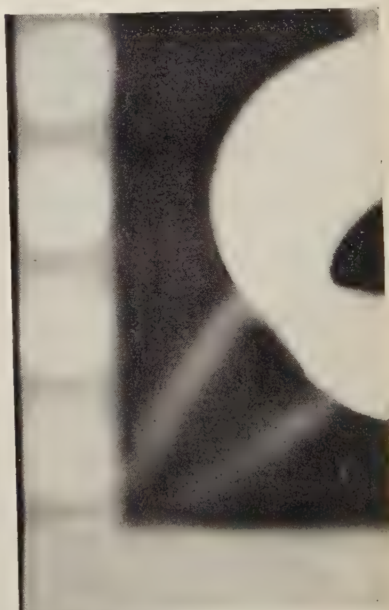

FIG. 43.—Radiograph of Plate (Fig. 42) after introduction of chills along the beading. Unsoundness removed. × ½.

[*Jackson.*

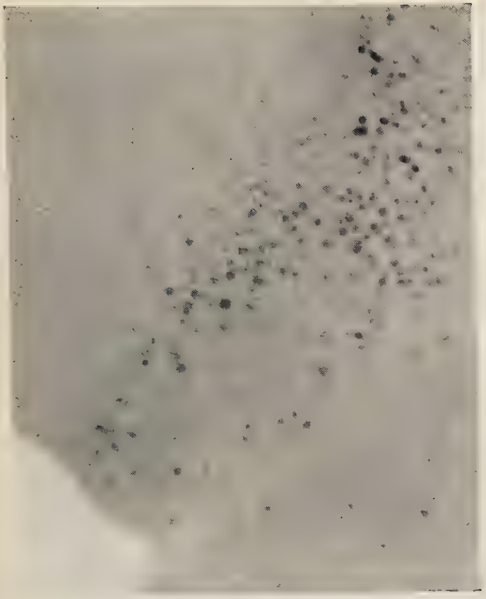

FIG. 44.—Radiograph of Corner of Plate, showing blow-holes produced by use of damp sand. $\times \frac{2}{5}$.

FIG. 45.—Section of the more Unsound Casting shown in Fig. 46. $\times \frac{2}{5}$.

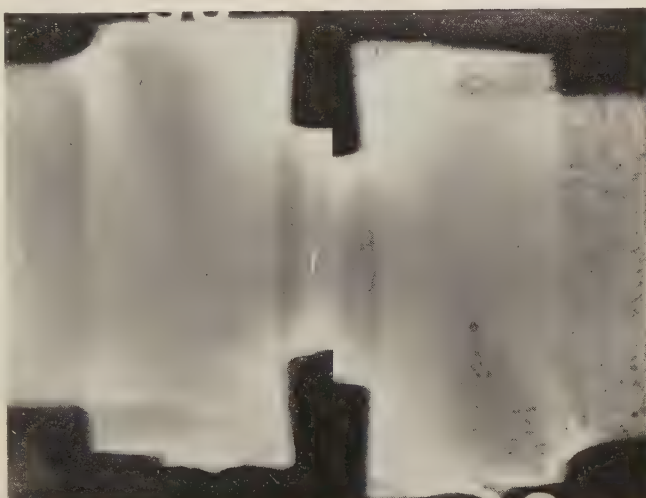

FIG. 46.—Radiographs of First and Last Castings from a Doubtful Heat. $\times \frac{2}{5}$.

[*Jackson.*

FIG. 47.—Radiograph of Defective Cap Casting, showing unsoundness. $\times \frac{3}{4}$.

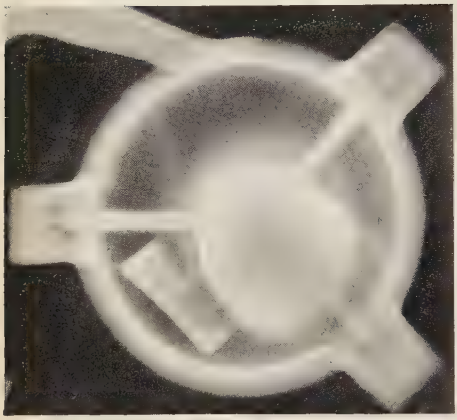

FIG. 48.—Radiograph of Cap Casting, revealing shrinkage cavities previously masked (Fig. 47). $\times \frac{3}{4}$.

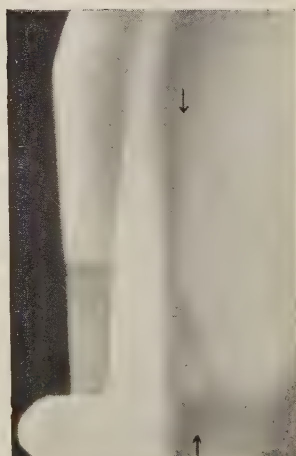

FIG. 49.—Radiograph of Casting showing Pull at T-junction. $\times \frac{3}{5}$.

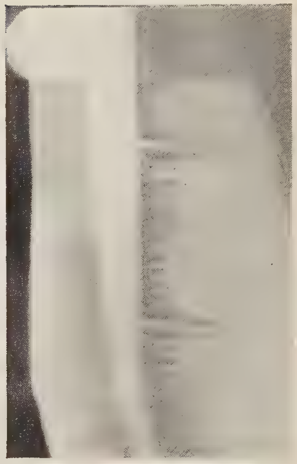

FIG. 50.—Radiograph of Casting, showing presence of nails and brackets. Pull eliminated. $\times \frac{3}{5}$.

[*Jackson.*

Yield point.	Tons per sq. in	27·0	22·7	22·7	22·0	23·8	22·7	18·5
Max. stress.	Tons per sq. in.	35·5	26·0	28·4	25·6	26·0	25·2	19·9
Elongation.	%	6·0	3·0	5·0	3·0	1·0	3·0	2·0
Brinell Hardness		192	187	187	192	192	187	192

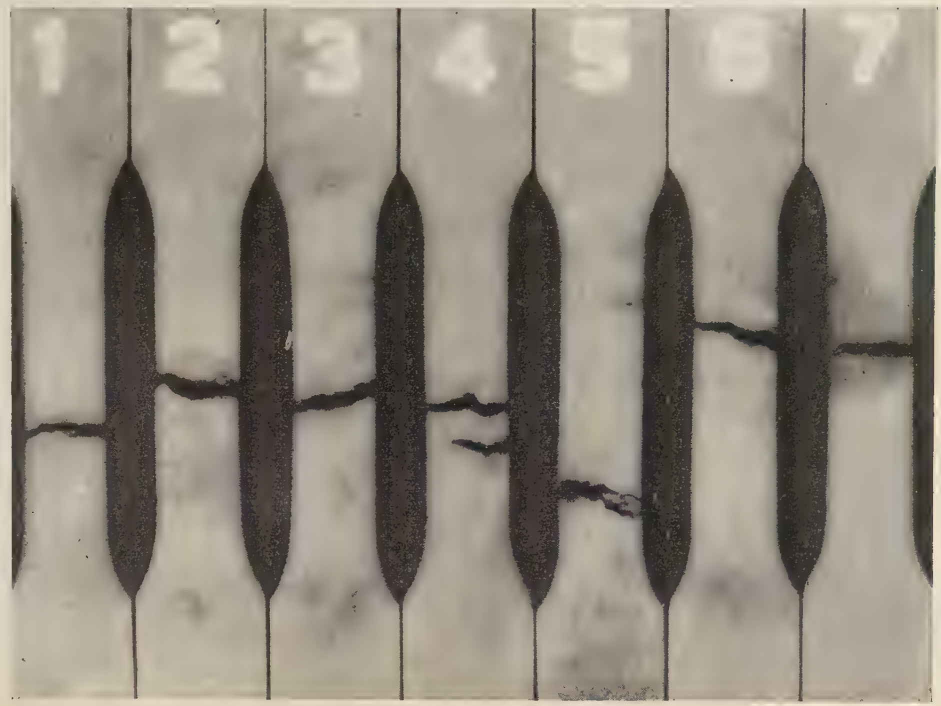

FIG. 51.—Radiographs of Test-Pieces arranged in order of soundness, with corresponding test results. × [illegible].

[*Jackson*.

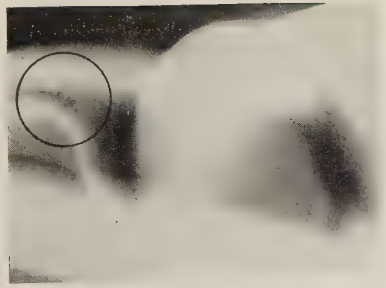
FIG. 52.—Pivot Bracket 1.


FIG. 53.—Pivot Bracket 2.

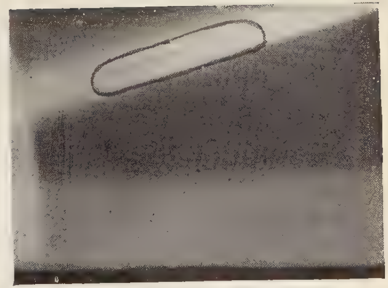
FIG. 54.—Pivot Bracket 3.


FIG. 55.—Pivot Bracket 4.


FIG. 56.—Pivot Bracket 5.

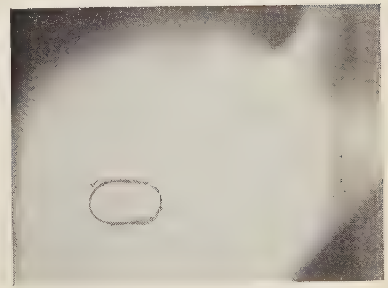
FIG. 57.—Pivot Bracket 6.

FIGS. 52 to 57.—Radiographs of Defects in Pivot Brackets. × ½.

[Jackson.

FIG. 58A.—Section of Pivot Bracket 1, showing unsoundness re vealed by radiographic examination. $\times \frac{1}{3}$.

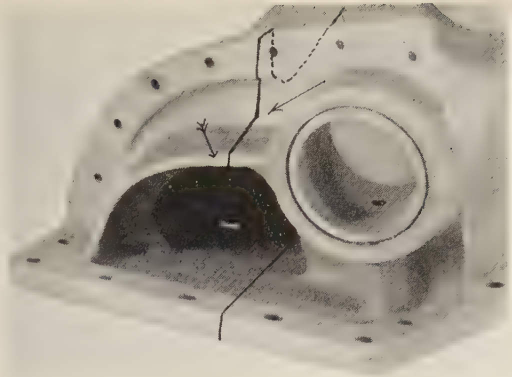

FIG 58B.—Relevant Part of Casting (Fig. 58A) with approximate posi ion of sectioning. *Plain arrow*, position of internal unsoundness; *barbed arrow*, position of surface unsoundness caused by test. $\times \frac{1}{4}$.

FIG. 59A.—Section of Pivot Bracket 2, showing unsoundness revealed by radiographic examination. $\times \frac{1}{4}$.

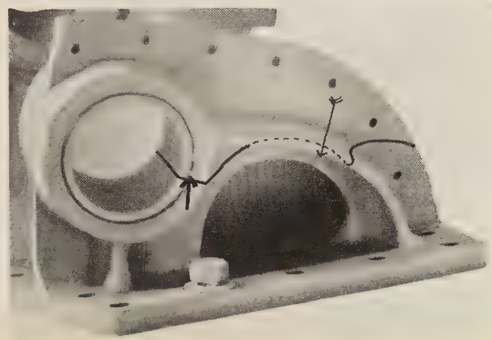

FIG. 59B.—Relevant Part of Casting (Fig. 59A) with approximate position of sectioning. *Plain arrow*, position of internal unsoundness; *barbed arrow*, position of surface unsoundness caused by test. $\times \frac{1}{4}$.

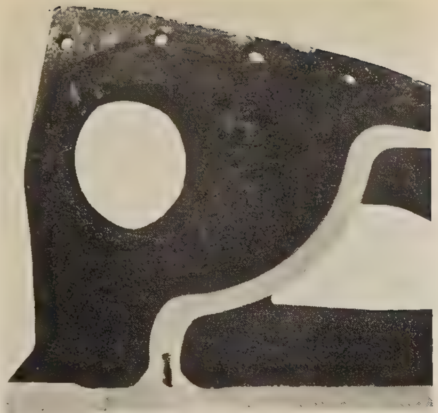

FIG. 60A.—Section of Pivot Bracket 3, showing unsoundness revealed by radiographic examination. × $\frac{2}{3}$.

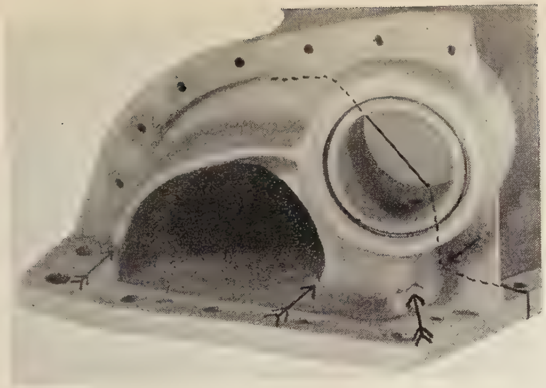

FIG. 60B.—Relevant Part of Casting (Fig. 60A) with approximate position of sectioning. *Plain arrow*, position of internal unsoundness; *barbed arrow*, position of surface unsoundness caused by test. × $\frac{1}{4}$.

FIG. 61A.—Section of Pivot Bracket 4, showing unsoundness revealed by radiographic examination. × $\frac{1}{4}$.

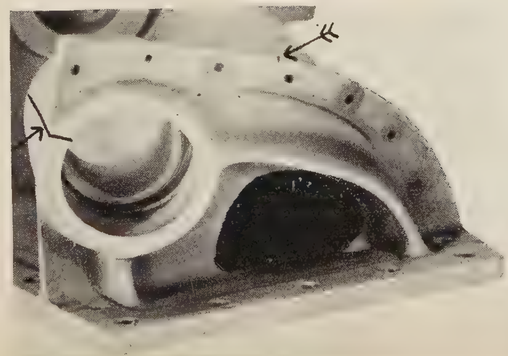

FIG. 61B.—Relevant Part of Casting (Fig. 61A) with approximate position of sectioning. *Plain arrow*, position of internal unsoundness; barbed arrow, position of surface unsoundness caused by test. × $\frac{1}{4}$.

[*Jackson.*

likely to occur in practice has to be fixed in an arbitrary manner. A method which is frequently adopted is to examine the mechanical properties of a test-piece cut from the casting, or a test bar made with it. These are all ways of obtaining an estimate of the strength of the casting, and are satisfactory to the extent to which they supply an answer to the question of the suitability of the casting for the purpose for which it was designed.

(1) *Comparison of Test-Piece Figures with the Results of Radiographic Examination.*

(a) *First Example.*

As a check on the quality of a number of plates which were being cast for experimental purposes, a test-piece was taken from one corner of each plate. The results of two typical tests are shown in Table I. compared with the test bars of clover-leaf section from the same heats of steel. It was deduced from these results that the plates were defective. As a result, a radiographic examination of one of the plates was made, and it was found that a line of shrinkage almost bisected the position from which the test-piece was cut. An examination of the plate as a whole, however, showed that this result was not typical of the remainder of the plate (*see* Fig. 28), and the latter may have been satisfactory for its purpose. The test result agreed with the findings of the radiographic examination for the part of the plate from which the test-piece was cut, but the radiographic examination enabled a better opinion to be formed of the value of the plate as a whole.

TABLE I.—*Comparison of Results of Mechanical Tests on Plates and Test Bars.*

Steel.	Test-Piece.	Test Results.					
		Yield Point. Tons per Sq. In.	Max. Stress. Tons per Sq. In.	Elongation. %.	Reduc- of Area. %.	Izod Value. Ft.lb.	Brinell Hardness.
Chromium-molybdenum	From plate.	...	50·4	3·0	trace	27, 22, 23	277
	Test-bar clover-leaf section.	58·0	64·0	8·0	11·6	32, 31, 30	302
Manganese-molybdenum	From plate.	...	49·5	2·0	...	15, 15, 13	302
	Test-bar clover-leaf section.	62·4	68·0	9·0	18·4	20, 18, 17	...

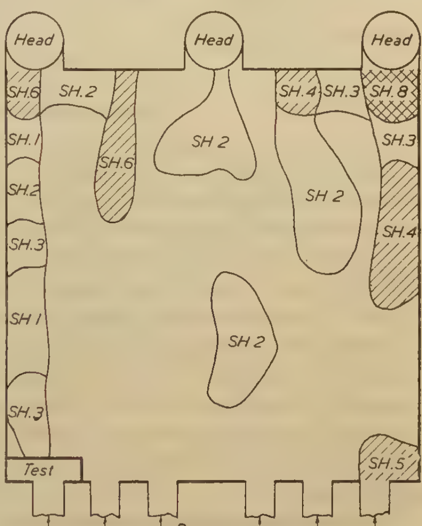

FIG. 28.—Method of Running and Diagrammatic Representation of Defects in Experimental Plate 3 ft. sq. × 20 mm. thick.

(b) *Second Example.*

In a further series of castings over a range of steels, radiographs were taken of parts of castings of a section of 3 in., and the pieces were marked out into areas which were sound or unsound, as defined radiographically. Test-pieces were then cut from the two groups of areas and these were pulled to indicate the strength of the part at that point. The results are given in Table II., in which the composition of the steel, the radiographic soundness of the portion and the mechanical-test results from it are set out; the mechanical-test results from the test bars which were cast separately are also given.

TABLE II.—*Comparison of Results of Mechanical Tests and Radiographic Examination.*

Composition of Casting.	Result of Radiographic Examination.	Test Results.					
		Yield Point. Tons per Sq. In.	Max. Stress. Tons per Sq. In.	Elongation. %.	Reduction of Area. %.	Izod Value. Ft.lb.	Brinell Hardness.
(1) Ni–Cr–Mo	Sound.	52·8	64·0	9·0*	8·4	26, 26, 26	293
	Unsound.	53·2	63·4	10·0	19·9	33, 37, 34	293
	Separate test bar.	55·2	64·8	13·0	21·6	32, 30, 30	293
(2) Ni–Cr–Mo	Sound.	53·6	67·6	4·5	8·4	42, 41, 42	293
	Unsound.	56·0	65·2	8·0	11·6	42, 42, 42	293
	Separate test bar.	52·4	62·8	16·0	42·0	50, 52, 39	310
(3) Cr–Mo	Sound.	50·8	62·4	10·5	24·4	52, 55, 54	302
	Unsound.	50·0	61·6	7·0	18·4	59, 56, 56	302
	Separate test bar.	57·2	67·2	14·0*	27·6	43, 42, 41	310
(4) Cr–Mo	Sound.	49·2	62·4	16·0	33·6	43, 43, 43	302
	Unsound.	50·0	63·2	12·0	21·6	54, 53, 52	302
	Separate test bar.	46·8	59·2	20·0	39·2	41, 38, 39	285

* Broken outside middle half.

There is no clear differentiation in the results between pieces which were sound radiographically and those which were unsound. In some cases the results are as would be expected, whilst in other cases the unsound portion gave results superior to those from the sound portion.

The explanation of the apparent anomaly seems to be that the defects revealed by the radiograph were visible when the test-piece was being cut, and these parts were avoided in shaping the test specimen. The "sound" portions, on the other hand, may have contained some micro-shrinkage which was not visible in the radiograph and which remained unnoticed in preparing the specimen, but which was revealed by the test. It seems reasonable to suppose that if shrinkage in the case of the "unsound" portions had led to the formation of definite cavities inside the steel, then the steel from these cavities may have flowed into the remainder of the casting. The latter, being thus fed, contained no micro-shrinkage and would be perfectly sound as shown by the results of a tensile test.

The relation between soundness as shown by the radiograph and strength as given by a test-piece taken from only a part of the section is indistinct, if not entirely misleading. The only satisfactory method of comparison is to use the full thickness of the plate as the test specimen.

The difficulty of comparing radiographic results with the results of mechanical tests taken from the full thickness of the plate arises from the large size of the specimen and, consequently, of the test machine which would be required.

(c) *Third Example.*

For the reason given above, a series of thin plates were used as the basis for a further test of this type. The plates were 10 mm. in thickness, cast from a steel of the following analysis :

Carbon.	Silicon.	Sulphur.	Phosphorus.	Manganese.
0·33%	0·56%	0·040%	0·034%	0·84%

They were specially made to be defective. The radiographs of the plates were examined and test-pieces cut out so that defects of varying severity were present at approximately the centre of each specimen. The test-pieces were then radiographed and, after normalising to obtain a uniform condition in all cases, they were pulled. Tensile-test results are shown in Table III., the pieces being arranged in an order of merit derived from radiographic examination. Fig. 51 shows the appearance of the pieces after test, and indicates that failure occurred at the position to be expected

TABLE III.—*Comparison of Radiographic and Tensile-Test Results on 10-mm. Plates.*

Treatment of plates . . . . . at 850° C. in air.
Size of test-pieces . . . . . 0·75 × 0·375 in.
Area of test-pieces . . . . . 0·281 sq. in.
Gauge length of test-pieces . . . . 2 in.

Order of X-Ray Soundness.	Yield Point. Tons per Sq. In.	Max. Stress. Tons per Sq. In.	Elongation. %.	Brinell Hardness.
1	27·0	35·5	6·0	192
2	22·7	26·0	3·0	187
3	22·7	28·4	5·0	187
4	22·0	25·6	3·0	192
5	23·8	26·0	4·0	192
6	22·7	25·2	3·0	187
7	18·5	19·9	2·0	192

from an examination of the radiograph. From Table III. it may be seen that there is a good correlation between the results of the tensile test and those of the radiographic examination. There is actually a better correlation between the orders of merit obtained from the radiographic results and the maximum stress than there is between the orders of merit revealed by the mechanical tests themselves. There is some difficulty in deciding from the radiographs the exact position of test-pieces 2 to 6, as some test-pieces have a small area of severe defects, whilst others have a larger area of less severe faults.

The comparison of these results, while not revealing a perfect relation between the radiographic interpretation and the tested strength of the plates, does indicate that the results of radiography can be a good guide to the strength of the casting when the full thickness of the part is considered. All these test-pieces were from the same heat of steel and were sections of only four plates. The undesirability of judging the value of a cast from a single test conducted on only one specimen cut from any one position in one plate would appear to be amply demonstrated. Further, in the previous tests the fallacy of comparing the radiographic soundness of

a section of a plate with a tensile test taken from only a part of that section is clearly shown.

The use of radiography enables the whole of the casting, in most cases, to be studied for soundness, and a better knowledge of the capabilities of the specimen is obtained than by tests taken at random. The taking of test-pieces from the casting can give useful information about the material of the test-piece, but can supply little data about the properties of the casting as a whole.

(2) *The Comparison of Radiographic Examination and a Proof-Load Test.*

A more satisfactory approach to the testing of a casting than the taking of test-pieces from the specimen is the use of a proof-load test. By way of example a comparison of this kind has been applied in the case of complicated castings which are subjected to high stresses in service. The preliminary work in the development of foundry technique was checked by extensive radiographic examinations, but the final inspection for acceptance purposes was undertaken by a proof-load test.

In this case the proof-load test does not give a direct comparison with the soundness of the casting as found by radiography, as the latter is only used during routine production in the event of doubt arising as to the quality of a casting. It does, however, measure the standard of the foundry practice developed by the use of the radiographic method, and some idea of the relation between the radiographic soundness of the casting and the proof-load test may be deduced.

When the development of the casting was such that no obvious defects were noted by the eye, the whole casting was examined as fully as possible with X-rays in order to find the points of weakness in the method of casting. As a result of these findings the standard was improved until an article was produced which, though not perfect as examined radiographically, was considered to be fully satisfactory for the purpose. During these experiments the quality of the steel was tested by taking tensile tests from the body of the casting in comparatively sound parts, to ensure that radiographic soundness included also freedom from small defects such as micro-shrinkage. When these tests were satisfactory, and the unsoundness found by radiography had been reduced to a small amount occurring at random, the casting was considered to be good enough for trial. After machining, it was subjected to a proof-load test.

On one type of such a casting, out of over 1000 castings tested not one failed to carry the required proof load. To obtain even more information, a typical casting was pulled beyond the proof load and eventually failed when the load had been exceeded by a factor of 2·02, the figures being: Proof load, 69 tons per sq. in.; breaking load, 140 tons per sq. in.

These tests show that the foundry practice developed by the use of radiography has enabled a standard of casting to be maintained which fulfils all normal safety-factor requirements, and which, further, has given an additional safety factor of 2·02.

(3) *Comparison of Results of Radiographic Examination with Tests to Destruction.*

In order to determine, on another type of casting, the amount of defect definitely detrimental to a specific casting in the process of manufacture, a series of tests to destruction was carried out, reproducing as far as possible the actual service conditions. Six castings were chosen which, when radiographed, all revealed defects regarded on the standard then established as sufficiently serious to warrant rejection. The portion of the casting tested, the radiographs revealing the defects, and the

position and appearance of the sections of the four castings afterwards cut up to show the defects are illustrated in Figs. 52 to 61. The castings were tested by attaching the body to a base plate by bolts, and applying a load to the eye by means of links attached to a pin passed through it.

TABLE IV.—*Results of Tests to Destruction on Pivot-Bracket Castings.*

Estimated maximum working load = 16·6 tons.

Castings Examined.	Max. Load Attained. Tons.	Results.	Conclusions.
First pair.	37·4	Holding-bolts sheared.	Breaking-load not reached.
Second pair.	46·8	One casting cracked, nuts stripped and holding-bolts sheared.	One casting cracked.
Third pair.	42·3	One casting broke.	One casting failed.

The results of the test are given in Table IV. The maximum working load for these castings in service was estimated at 16·6 tons so that the safety factor in the case of the castings which failed was at least 2·5. These castings were all rejected as defective on the results of radiographic examination, although it is clear from these tests that the castings would have been satisfactory in service. The effect of shock-loads could not be taken into account in the tests, the conditions of service being such that the shock-load could not be evaluated. Although these castings contained defects which were revealed by radiography, it would appear that the general quality of the part was such that it was fully up to its task.

During the examination of these castings the highly stressed eye was tested for the quality of the steel by taking test-pieces from the positions

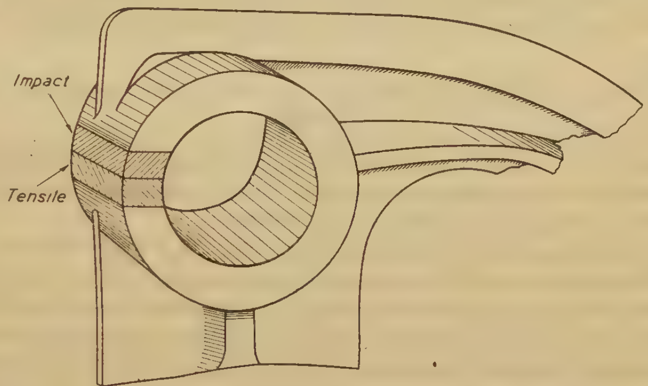

FIG. 29.—Position of Mechanical Test-Pieces taken from Boss of Pivot Bracket.

shown in Fig. 29. The results of these tests are given in Table V., in which the values are compared with the specification to be met by the steel.

The radiographic examination demonstrated that, with the exception of the small defects found, the castings in the highly stressed areas were

TABLE V.—*Results of Tests on Bosses of Pivot-Bracket Castings.*

Bracket Number.	Tensometer Tests.				Hounsfield Impact Test. Ft.lb.	Evquialent Izod Value. Ft.lb.
	Yield Point. Tons per Sq. In.	Max. Stress. Tons per Sq. In.	Elongation. %.	Reduction of Area. %.		
1	27	42·4	30	54	9·7	24·3
2	32	43·0	28	58	3·6	9·0
3	31	44·5	28	50	7·3	18·2
4	32	45·8	23	48	4·5	11·3
Specification.	25	40	26	20	...	16·0

sound, the mechanical tests indicated the quality of the steel and the combined effect of these two factors is shown in the results of the destruction test. The foundry technique was designed to give a maximum of soundness in the eye, which is the most highly stressed part of the casting. It is known that, as a consequence, minor defects may occur in other, lightly stressed, parts of the specimen, but as these parts (even if somewhat defective) are quite strong enough to carry the load required, such faults are of no consequence.

(4) *Comparison of the Results of Radiographic Examination and Service Tests.*

The castings discussed above were tested under conditions designed to simulate service conditions. As a short-period test was required, however, and as the conditions in service are not known with absolute accuracy, such tests are only approximations to the conditions under which the castings will be used. As already stated, a test in service is a long-time undertaking unless premature failure occurs. As far as is known at present, the service life of these castings is quite satisfactory, even when the conditions have been very much more severe than could have been anticipated, and no failures have so far been reported.

SECTION V.—THE INFLUENCE OF RADIOGRAPHIC SOUNDNESS UPON DESIGN.

The consideration of the quality of the castings mentioned above, as measured by normally applied tests and as confirmed by the destruction and service tests, raises the question of the safety factor which it is necessary to apply to castings of this standard.

In the past the quality of normal commercial castings was such that the engineer was obliged to increase his factor of safety in castings to allow for internal defects of considerable magnitude. Such castings were, in general, used for purposes for which the additional weight to cover this increase in safety factor was a matter of little concern to the designer; a factor of safety for castings of approximately double that for forgings was perfectly normal. However, as has been shown by the foregoing discussion, if care is taken in the manufacture of castings, even of considerable complication, and if such castings are examined by radiography to ensure that the standard is as satisfactory as is required, there would appear to be no reason why such castings should not be designed with a factor of safety approaching that of a forging. In this connection it may be said that forgings have been examined which have shown defects of a much more serious nature than any found in a casting.

With castings designed to work at smaller factors of safety than those employed at present, it would be necessary to study the influence of the defects found to be present. Now that the position and size of these defects may be determined, there may be no need to impose an overall factor of safety, but the influence of the separate defects may be individually considered; this opens up a considerable field for future work. A direct approach to the problem would be to study the effect upon the strength of the casting of the different types and severities of defect which occur. In the case of the casting considered in Section IV. (3), this procedure was carried out, and the result, as already stated, revealed that the casting was much stronger than had been expected. Tests such as this are clearly the preliminary step in the adoption of standards for the acceptance or rejection of these high-quality castings on the basis of their radiographic examination. There is little doubt that if this procedure were applied in Great Britain there would be a much greater realisation of the use to which castings could be put than there is at the present time.

Section VI.—Conclusions.

The work set out here has shown that the use of radiography is able to be of great assistance in the development of steel castings. The usual types of defect met with in these castings, such as shrinkage, inclusions and cracks, can be found non-destructively if they make a difference in the effective thickness of the casting of about 2%, and, in some cases, if they are even smaller than this. It is possible by this method to differentiate the various types of defect and obtain some idea of their size and position. In addition, the examination gives a general view of the whole of the casting, and the distribution of the defects is disclosed, a factor of great assistance in finding the cause of the trouble, and in removing it.

The radiographic method of examination has the advantage over that of sectioning, since, in addition to being non-destructive, there is little chance, in a careful survey of the specimen, of serious defects being missed. It is self-evident that in sectioning a casting there must be a very grave danger of defects being overlooked, unless they are so serious that they extend over a considerable volume of the casting.

For the usual purposes of determining the existence and location of any defects, the radiograph is, in general, easily interpreted by anyone with a knowledge of the geometry of the specimen and of the direction of the X-ray beam when the radiograph was taken. The technique of taking the radiograph and its interpretation in detail are, however, matters for the expert, and by a faulty procedure it is readily practicable to obtain from a defective casting a radiograph on which the specimen appears quite sound.

The value of the employment of X-ray equipment in obtaining sound castings is evidenced by the improvement shown in the various examples which have been quoted, as one fault after another has been eliminated from the manufacturing process. It may, in fact, be quite truly stated that, in the case of some of these castings, the required standard of soundness could not have been obtained without the free use of this tool.

The effect upon the use of castings of the increase in soundness should be profound. The use of castings has frequently been based on the idea that a casting is the poorest form of metal, and can only be relied upon if there is a considerable bulk of metal in excess of that which would normally be required, as, for example, in the case of a forging. If the casting can be shown to be sound when required, then the need for many of the additional factors of safety may be removed; the casting may then compete (on a basis of similar weight) with the forging, result-

ing, in general, in a considerable saving in cost. There may arise the need for more accurate interpretation of the effect of faults upon the strength of the article, as small defects which are approaching the limit of visibility by the radiographic method may have a vanishingly small effect upon the strength of the part.

Whilst it is evident that radiographic technique can be applied to the determination of internal defects, there is still a considerable gap in our knowledge of the influence of the defects on the subsequent useful performance of the casting. Even with regard to the interpretation of the radiograph, there may be uncertainty in defining the exact size of a defect. Little has yet been published on the quantitative effect of any specific type of defect on the strength of a casting. This difficulty is overcome, to some extent, in the examination of welds in plates by establishing a series of standard radiographs which are used for assessing the value of the examined weld. So far, the complexity of castings and the lack of knowledge of what constitutes a defective casting have prevented the widespread application of this method to the foundry. There is much scope here for work on the influence of defects upon the strength of castings, provided that the shape, position and size of the defects are known. Standards of assessment are required to be known in terms of these last-named factors. They must also be correlated with the mechanical properties of the casting as a whole and with its ultimate performance in service. The results of some efforts in these directions have been described in this paper.

APPENDIX I.—*Relation Between Density of Radiograph and Size of Defect.*

When a beam of monochromatic X-rays passes through a slab of material the intensity of the rays emerging from the slab is given by :

$$I_E = I_0 e^{-\mu t},$$

where I_E = intensity of emergent radiation,
I_0 = intensity of radiation at the same point if the slab were removed,
μ = coefficient of absorption of the material,
and t = thickness of the slab.

The coefficient μ has different values depending on the wave-length used, and, as all X-ray beams, unless specially filtered, consist of a wide range of wave-lengths, the absorption equation can, in practice, be regarded only as an approximation.

Further, it has been assumed that all the X-rays absorbed from the initial beam were entirely removed. Actually, however, this is not the case. The rays which have been said to be absorbed may possibly have suffered true absorption by the atoms of the material, in which event there will be a corresponding electron emitted from the material at this point; or, as is more probable, the original rays have been deflected away from the initial direction, so becoming lost to the original beam but appearing at some other place. Again, the rays entering the specimen may be altered in direction and also in wave-length. Such alterations to the beam are known as internal scatter, and cause part of the rays which enter the specimen to emerge at some distance from the original path, in a different direction and with a different wave-length, and hence with a different effect upon the film. This type of scatter results in a general blackening of the film, thus producing a background of " fog."

Referring again to Fig. 1, it will be seen that the rays which pass through the hole X may be deflected before they reach AB, and, also, much scattered radiation from the body of the specimen may reach AB,

and so mask the effect of X. In the case of Y, these events are less likely to occur, and so Y will usually give a clearer impression at CD than X gives at AB.

To ascertain the effect in practice of these varying conditions, an experiment was carried out with a number of slabs of steel, in one of which were cut a number of slots—all of the same width but of varying depth.[1] This work confirmed that the slots which are nearer the film are more easily seen, as they produce a greater alteration in density on the film; also a decrease in the kilovoltage applied to the X-ray tube increased the effect of the slots. The relation between density of film and size of slot in any one position was not linear, but varied with kilovoltage. The kilovoltage cannot be indefinitely lowered, thus increasing indefinitely the influence of the defects, as the exposure time is correspondingly increased. In a further series of tests the following exposure times were required to give the same film density through a block of steel:

Kilovolts	400	380	360	340	320
Exposure time. Min. .	25	33	50	75	120

The use of exposure times much in excess of 30 min. is not generally an economical proposition, and so a compromise must be reached between the sensitivity required and the time allowable for the exposure.

Appendix II.—*The Printing of Radiographs.*

As mentioned earlier, the reproduction of a radiograph in the form of a print is not an easy or satisfactory procedure. The main difficulty arises from the considerable difference in the density range of X-ray films and printing paper. The former may have a density range of from 0 to 5, whereas the latter is limited to a range of from 0 to approximately 1. As a result, any attempt to reproduce the density variations of the film on paper must either be done by "flattening" the radiograph, so that the range of density is reduced from 5 to 1, or by reproducing each of the ranges 0–1, 1–2, &c., and producing five complementary prints. The first method results in the complete loss in the print of any small changes in density present in the film, such as result from minor defects on the casting, whereas the second method necessitates large numbers of prints for each casting examined. The latter difficulty may be overcome by taking appropriate regions of each print containing only one of the ranges of density, and so a single composite print may be produced. This procedure is not only time-consuming but also may lead to erroneous conclusions, as a part of the film, say, of density 0·9, being at the high end of the 0–1 range will appear to be more dense than a part, say, of density 1·1, which is at the low end of the range 1–2. The question has been discussed by Thorpe and Davison,[6] who have suggested a method of printing which utilises a mask prepared photographically.

In their method a print of the radiograph is first made on film; this print is termed the mask. The radiograph is now placed on top of the printing paper in the normal way, but the mask, separated from the radiograph by a sheet of plate-glass, is interposed between the radiograph and the printing light. The mask, having densities opposite to those on the radiograph, tends to equalise the light reaching the paper, and so produces a uniform grey print. This equality between radiograph and mask is never obtained, but the contrast to be reproduced in the print is decreased. Although this levelling of density applies to the major areas of the radiograph, the fine details, such as defects in the casting, are apparently not reduced, and show clearly on the print.

This method was tried in the preparation of the illustrations for this paper, but was unsuccessful because of difficulties in registering the radiograph and mask with sufficient accuracy. Although the method of two positioning pins, suggested by Thorpe and Davison, was tried, and apparent accuracy of positioning obtained, the resulting print appeared as a bas relief, because each change of density was outlined in white on one side and black on the other.

To overcome this defect a sheet of flashed opal glass was introduced in place of the plate-glass and the bas-relief effect was eliminated. The light from the mask was apparently spread slightly, but only sufficiently to blur out the sharp-edge effect previously obtained.

The density of the mask, as would be expected, was found to have an appreciable effect on the reduction of contrast obtained, and could be adjusted to suit the individual radiographs. Where radiographs with little contrast were to be reproduced, a straight print was made without a mask, as the latter was clearly redundant.

To avoid confusion between the interpretation of the print and that of a radiograph, all the prints reproduced in this paper have been made so as to resemble the radiograph, *i.e.*, the dark regions in the radiograph are reproduced as dark areas on the print, and not as light areas, as would be obtained by a direct-contact print. This is achieved by printing first upon film, and then from this upon paper.

The masking process for high-contrast radiographs therefore requires three stages. First, a print is made upon film, for use as a mask. Secondly, using both radiograph and mask, separated by flashed opal glass, a second print is made upon film; this is a low-contrast reversal of the original. Finally, this last film is printed upon paper, giving a low-contrast representation of the radiograph.

As examples of the two methods of printing for low- and high-contrast radiographs, Figs. 31, 32 and 33 were made by the normal practice, and Figs. 35, 37, and 42 were made by using a mask.

Acknowledgments.

The author wishes to express his thanks to the Directors of Messrs. Hadfields, Ltd., for permission to publish the paper, and in particular to Mr. W. J. Dawson for suggesting the subject of the work and for his helpful criticism of the paper. Thanks are also due to the Ministry of Aircraft Production for permission to publish the results of those tests referred to in Section IV. of the paper, which were carried out by the Ministry, and to the Foundry Practice Sub-Committee of the Steel Castings Research Committee for permitting the use of results obtained during work carried out on their behalf.

The author is also grateful to Dr. R. J. Sarjant for his inspiration and continual help during the course of the work, and to Mr. J. Rhodes and the other members of the Research Department, who are responsible almost entirely for the illustrations. The assistance received from the members of the Foundry Management, who co-operated freely during the whole course of the investigations, is also gratefully acknowledged.

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DISCUSSION.

Dr. F. C. LEA (Messrs. Edgar Allen & Co., Ltd., Sheffield) : I should like to congratulate the author on this excellent paper. I am going to speak from the point of view of an engineer who is also interested in metallurgy, and who has been interested for many years, as Chairman of a certain Committee, in the attempt to relate the evidence obtained from radiographs to the use of the particular piece of work. It seems to me very important, as Dr. Jackson has indicated, that as far as possible radiography should be used to promote increasing confidence in the use of steel castings in the future.

The author reminded us of an important point, namely, that if there happens to be a thread-like crack in a slab in the direction shown in (*a*), Fig. A (X-ray beam vertical), it is not likely to be revealed by the radiograph, whereas if it is in the direction shown in (*b*) and it is sufficiently wide to show at all there will be a difference of density on the radiograph and the existence of the crack will be evident. I am a little afraid, however, that these thread-like cracks which are so difficult to discover are probably—I should not like to put this in any *ex cathedra* way—from the engineer's point of view more dangerous than most other types of discontinuity within the casting.

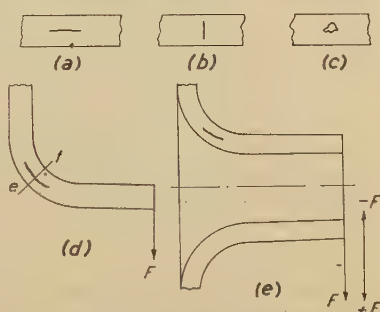


FIG. A.—Cavities in Castings.

For instance, if we have a very considerable hole in the casting, as in (*c*), Fig. A, that will be shown. If, in a bend of the type shown in (*d*), Fig. A, such a cavity occurs, we have to think of its significance in terms of the types of forces that are acting. The author has referred to static tests, and I want to give a word of warning as to the significance of static tests, though I shall not discuss that at any length. If there is in (*d*) a crack or cavity and it is subject to a force as shown by the arrow, then we have a bending moment on the section *ef*; if the discontinuity occurs very nearly at the neutral axis, it may be comparatively unimportant, considered purely from the engineering point of view. If, however, the casting is of the section shown in (*e*), and the static force F is applied as there shown, then obviously the discontinuity is not anywhere near the neutral axis, and the bending moment due to F gives rise to a considerable stress and stress concentration. If a thread-like shrinkage or crack occurs there, we shall obviously have very big stresses at that point. If the load is simply a static one, I doubt whether it will matter much, though there will be a stress due not only to the bending but also to the sharp discontinuity which will increase the local stress considerably; however,

since nearly all castings will have some degree of plasticity or ductility (whichever term we like to use), I do not think that a static stress will of necessity be very important at that point.

Let us suppose, however, that the casting is subjected to repeated stresses due to $+F$ and $-F$. Then the discontinuity is of the greatest importance. The point that I want to make is that in interpreting the discontinuities within castings it is most important to consider how they may be removed by improved methods of casting; whether they are important from the point of view of the use of the casting, however, will depend largely upon whether the part where the discontinuity occurs is subjected only to what may be called a static load or to an impulsive load (*i.e.*, a blow-like kind of load) or, more important still, to repeated stresses. Some of us have seen castings where we know that the stresses, as computed by ordinary methods, are very low, and yet failure has occurred. When the casting has been cut through, we have found that there have been no very large discontinuities, but probably some indication or trace of thread-like discontinuities or thread-like shrinkages—how they occur does not matter for the moment. The only interpretation that we can then give of the failure of the casting is that it was due to repeated stresses, which caused very considerable concentrations of stress at a discontinuity which would have been discovered with the greatest difficulty—and this is the point—by radiography. We may take a good radiograph through the bend, if this is only 2 or 3 in. thick, and yet may not be able to discover that crack or thread-like discontinuity.

I suggest in conclusion, therefore, that we should be cautious in our interpretation of static stresses, and particularly in our interpretation of the meaning of a tensile test-piece cast with the casting—the author has already given us an illustration of a case where that was of very little use; but in considering these discontinuities and the interpretation of the radiograph, we must be careful to consider the significance of any discontinuity which we discover in relation to the particular kinds of stresses, and particularly repeated stresses, which may occur.

Dr. L. MULLINS (Kodak, Ltd., Harrow, Middlesex) : Although X-ray inspection has been widely used during the past decade as an aid in devising sound foundry techniques,* † its potentialities are by no means fully realized, and the author has done a service to foundrymen by so ably demonstrating the possibilities. Too often radiography is limited to the inspection stage, where it can make little if any direct contribution to the improvement of the quality of the incoming castings. Where X-ray inspection is not applied, as is most frequently the case, then even these small possible advantages are completely missed.

That radiography can prove a valuable servant to the foundryman has been amply shown by this paper, and is borne out by the following striking example : 50,000 castings of a certain type were made under what was thought to be ideal foundry practice, and of these 14% were condemned on inspection; following an X-ray study of the castings, the foundry conditions were changed, and there was an almost unbelievable improvement in the resulting quality, only 1·2% of the next 75,000 castings being condemned.

In the past, X-ray technique has suffered from the lack of any sound universal quantitative scheme for grading the extent of any defect. Only recently a grading of defects by numbers relating to verbal descriptions

* F. R. Mansfield, *Light Metals*, 1943, vol. 6, Oct., pp. 469–478.

† E. Cook, " Foundry Applications of Radiography," Symposium on Radiography and X-Ray Diffraction Methods, pp. 25–52. Philadelphia, Pa., 1937 : American Society for Testing Materials.

of their extent was proposed by the appropriate section of the American Society for Testing Materials.* The author has gone further than this by showing actual examples of the various grades of defect, and his work confirms that the proposals of the American Society for Testing Materials could be applied in practice.

There are, however, a few points concerning this grading on which I should welcome further information. In Fig. 32, showing the first eight radiographic standards for shrinkage, the shrinkage corresponding to the standard *SH*.6 appears to be worse than that for *SH*.8. This may, of course, be due to poor printing, but as shown the standards seem to be in the wrong order in the bottom row.

I would also like to know the criterion by which the total extent of the shrinkage is assessed when giving verbal descriptions such as "the amount of shrinkage is reduced." Is this based on the area, in the radiographic image, showing shrinkage, or does it take into account both the area of the image affected and the grade of the shrinkage present?

In an attempt to determine whether either of these methods of assessment gives a clearer indication than the contour drawings of the relative extent of shrinkage in each case, I have prepared Table A, which shows,

TABLE A.—*Grade of Shrinkage.*

Fig. No.	Maximum Grade of Shrinkage.	Sum of Areas of Shrinkage.	Weighted Sum of Areas of Shrinkage.	Author's Remarks.
4	8	58	212	Considerable amount of shrinkage.
6	7	55	209	The effect was slightly to reduce the amount of shrinkage [compared with Fig. 4] but not to alter its shape.
8	6	53	200	The effect . . . was a reduction of the amount of shrinkage [relative to Fig. 6] and an alteration in the lines of flow.
10	4	39	94	The form of shrinkage was again altered and the amount reduced [relative to Fig. 8].
12	8	67	315	Considerable amount of shrinkage.
14	8	61	334	Slight improvement on the previous one.
16	7	82	244	Amount of shrinkage . . . only a slight improvement on the previously vertically cast plates.

for each casting, the maximum grade of shrinkage present, the sum of the areas of shrinkage in the film in arbitrary units, and the author's description of the extent of the shrinkage. In addition, a "weighted" sum of the areas of shrinkage is given, this being evaluated by taking three times the area of regions marked *SH*.3, eight times those marked *SH*.8, and so on. The values for both the sum of areas of shrinkage and the weighted sum support Dr. Jackson's remarks in indicating a slight drop from Fig. 4 to Fig. 6 and from Fig. 6 to Fig. 8, but it is questionable whether these differences are significant in view of the likely experimental error in determining these values. A more significant feature in these examples appears to be the drop in the maximum grade of shrinkage present.

An important point is illustrated by the later values in Table A, in which the weighted sum of shrinkage areas is seen to give a far better

* "Proposed Recommended A.S.T.M. Industrial Radiographic Terminology," Symposium on Radiography, p. 256. Philadelphia, Pa., 1942: American Society for Testing Materials.

indication of the extent of the shrinkage than does the simple sum of affected areas in the image. On the other hand, the value of the total area of shrinkage in the image supports Dr. Jackson's claim that Fig. 14 shows less shrinkage than Fig. 12. However, the weighted sum reverses the relative amounts of shrinkage, whilst the maximum degree of shrinkage is identical. The final vertically cast plate (Fig. 16) is claimed to show an improvement in shrinkage relative to earlier vertically cast plates; this is supported by the weighted sum of affected areas but not by the simple sum.

Table A also clearly illustrates the difficulty of giving a verbal description of the amount of shrinkage, or any other defect, present in a casting. Both Fig. 4 and Fig. 12 are claimed to show a "considerable amount of shrinkage," but both the simple sum and the weighted sum of the shrinkage areas show Fig. 12 to be worse than Fig. 4.

It follows from the theory of radiographic image formation that the degree of blackening of the film depends on both the density and the thickness of the material traversed by the X-ray beam. Moreover, the radiographic image is a two-dimensional representation of a three-dimensional structure. Accordingly, the blackening of the film due to a relatively thick layer of fine shrinkage normal to the X-ray beam might well be confused with that due to a thinner region of coarser shrinkage unless the latter were distinguishable by the form of the shadow, *e.g.*, a branching structure. For this reason it would be helpful to show corresponding macrosections with the graded radiographs reproduced in Fig. 32. At this stage it is pertinent to ask whether Dr. Jackson has sectioned any of his plates to confirm that the shrinkage present is of the order indicated by his contour charts drawn from the radiographs.

The setting up of such standards presupposes standardized exposure and processing conditions. Whilst these may be attained within any one laboratory, there are difficulties, as the author rightly points out, in setting up standards which are interchangeable from one X-ray unit to another. Let us consider the factors involved, apart, of course, from such readily fixed factors as choice of film, processing and viewing conditions. The differences in radiographs of the same subject obtained on two X-ray units, operating at the same kilovoltage, will be due primarily to two causes. The first is differences in radiation quality. These will arise owing to errors in kilovoltage calibration of the equipment, to differences in wave-form of the equipment, and to inherent filtration within the tube window. The effect of these factors on the sensitivity of fault detection is well known.* † To quote one example, the higher the kilovoltage the poorer is the sensitivity of fault detection. The other factor is focal-spot size, that is, the size of the area on the tube target from which the X-rays arise.

The use of calibrating metal step wedges would permit of fairly simple cross calibration for radiation quality, and therefore we are left with the single factor of focal-spot size. This has the effect of creating a penumbra or edge shadow on the image, and it would appear possible to modify the focus–film distance in relation to the focal-spot size so that the penumbra effect on the two images would be identical. A limitation here arises, of course, if there is a wide variation in the sizes of the focal spots, since then the ratio of tube–film to object–film distance would be very different under the two conditions. With these facts in mind, it is suggested that

* J. T. Norton, "The Principles of the Radiographic Process," Symposium on Radiography and X-Ray Diffraction Methods, p. 16. Philadelphia, Pa., 1937: American Society for Testing Materials.

† J. E. de Graaf, *Journal of the Institution of Electrical Engineers*, 1939, vol. 84, No. 509, pp. 545–551.

the possibility of interchangeable standards of the type which the author has described might be more fully investigated with advantage.

A further aspect of the method which requires fuller investigation is, as both the author and Dr. Lea have so ably pointed out, the practical importance of any defect under service conditions. The author's work has confirmed that a fair degree of imperfection can be tolerated with impunity, and many X-ray workers could detail examples in practice where a casting containing radiographically evident defects is nevertheless capable of withstanding service conditions. In the absence of this information, standards of radiographic approval will tend to remain artificially high, as a safeguard against fixing too low a standard of approval, and perhaps also against the possibility of the X-ray method missing certain important though small defects—a feature that Dr. Jackson has not hesitated to mention. The chief difficulty here is that the X-ray method cannot prove the absence of a crack, but can only confirm its presence. The detection of a crack depends not only on its orientation with respect to the X-ray beam, but also on its dimensions. Dr. Jackson has indicated that a crack can generally be revealed by X-rays, provided that its plane is normal, or nearly so, to the plate surface. This is not necessarily so if the crack is very narrow. A striking example of this limitation was found in a ½-in. butt-welded mild-steel plate in which there was no fusion of the noses of the double-U-type weld. The resulting defect was equivalent to a crack with its plane normal to the plate surface, with a width of approximately 0·0005 in. and depth of approximately ⅛ in. This defect was not detected on routine X-ray examination and was only just revealed by a very critical examination under laboratory conditions.

Economic considerations will operate against taking an infinite number of views to track down every crack, so that it must not be assumed that the radiographically proved casting is free from such defects. The possibility of cracks and other fine defects being present can, however, be minimized by utilizing the most critical radiographic technique available, and also by employing an experienced and well-trained radiographer.

Mr. W. H. Salmon (Messrs. Hadfields, Ltd., Sheffield): I should like first to refer to Dr. Lea's sketch of an L-section (Fig. A (*d*)); it is interesting to note that he did not draw it with a square outside corner. Some engineers may think that the square-corner section is stronger, because there is a bigger area of metal across the corner; but the section as drawn by Dr. Lea is stronger as a casting than the other, heavier section. There is less tendency for a crack to occur in the position shown in (*a*), Fig. B, than in the directions shown in (*b*) or (*c*), and the latter types of crack would be more serious, as they emerge at the surface and cause local concentrations of stress.

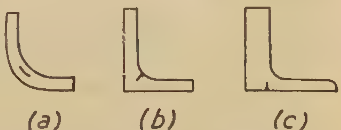


Fig. B.—Cracks in L-Sections.

Turning now to the illustrations in the paper, Fig. 1 is a diagram of the radiographic method of detecting faults in a block of steel. The X-ray beams are shown as a series of parallel lines. In practice, do the divergence of the beam and the scatter tend to make the defects appear larger than the actual size? Furthermore, are there sometimes other markings on the film, such as film faults, thumb marks, and indications of surface depressions caused by slag or sand, which may cause rejection of a casting unless the exposed film is carefully scrutinized and compared with the casting itself?

In Section II. (2), paragraph (*b*) (viii), it is surprising that the author says that it is not possible from the radiograph to distinguish between a hot tear and a clink. A clink type of crack, which occurs in heat-treatment, in local heating operations such as burning-off heads, heavy grinding or heavy machining, is usually narrow and straight. Sometimes it cannot be seen on a radiograph. Would the author agree that the magnetic-powder method is better than X-rays for the detection of such fine cracks in forged, rolled, and cast steel?

In Fig. 2 the author shows a drawing illustrating a method of representing typical faults found in a plate casting. In the first place, does not the author think that he has gone into too much detail in attempting to show ten grades of severity? Would not three grades have been sufficient, namely, *satisfactory* (0, 1, 2, and 3), *fairly good* (4, 5, and 6), and *bad defects* (7, 8, 9, and 10)? The foundryman is not so interested when he has made a sound casting, but he is very concerned to know which part of a casting is so unsound that he must take remedial action. Secondly, does not the author think that his use of full, plain lines to represent so many things on his drawing may lead to confusion? He uses full lines for six different purposes, namely, to represent (1) hot tears, (2) clinks or cracks, (3) the outline of the casting and runners, (4) the outline of the sound areas, and (5) the outline of areas which show thread-like shrinkage, and (6) in cross-hatching to indicate medium defects and double cross-hatching for bad defects. For instance, comparing the schematic drawing Fig. 4 with the print of the corresponding X-ray negative in Fig. 34, the latter is of far greater use to the foundryman; it also shows a hot tear on the edge of the plate which may be more serious in service than the thread-like shrinkage outlined on the drawing. I would suggest that the author should always show the positions of the runners and risers on the actual X-ray film, in order to provide a record of the method of moulding on the film itself.

In the last sentence on p. 239 P the author refers to the possibility of correlating such properties as the fluidity to the castability of the steel as evidenced by the soundness of cast plates. Could the author tell us more about the influence of fluidity on X-ray soundness, and would he consider giving another paper dealing with the X-ray soundness, and service results, of thin plates in different steels cast at different temperatures?

Fig. 17 is an illustration of a small fan-blade casting, the first casting commercially X-rayed by the author's firm. There is a good story about these castings. When the maker's X-ray men were installing and testing the set, they examined a number of production castings, made in the manner shown in Fig. 19, but could not find any defects. They could not make up their minds whether "by accident" some perfectly sound steel castings had been produced, or whether the new X-ray set was not giving sufficient penetration. The foundry made two defective castings for them without feeder heads, as shown in Fig. 17; one corner of the blades was "stopped off" in the mould in case the defective castings became mixed with the sound castings. The maker's X-ray people were greatly relieved to find that the set *was* working properly and did show the defect illustrated in Figs. 35 and 36. After this interlude the production of sound fan-blade castings continued, until one day the foundrymen noticed that some casts of steel started to rise half-way through the heat. When the castings from these heats were examined after cleaning, no signs of surface blow-holes could be seen, but X-ray pictures (Fig. 46) showed a large number of subcutaneous gas holes which would have been revealed on machining the blades. Good, medium, and bad castings were found in the same heat, moulded by the same man in the same sand, the

gas holes becoming worse towards the end of the heat. Another curious feature, illustrated in the sawn section in Fig. 45, was that the gas holes appeared only in the thinner sections, and not in the thicker sections under the heads of the castings. Can the author explain this effect and say how the error was corrected?

Another very interesting example is given in Figs. 47 and 48. Will the author tell us how the gas holes in the cap casting in Fig. 47 were cured, and does he think that the use of a more settled steel may have caused the shrinkage cavities shown in Fig. 48? In other words, does he think that a slightly rising steel may produce one type of defect, namely, internal gas holes, that a deeply piping steel may cause shrinkage defects, and a balanced steel no defects at all?

Fig. 51 represents radiographs of seven test-pieces, arranged in order of soundness, together with the corresponding tensile-test results. Can the author show us the X-ray pictures of the test-pieces before they were pulled? One wonders whether the defects appear worse now that they have opened out in testing. Are the defects associated with segregation? Referring to No. 1 test-piece, 6% elongation seems a very low ductility for a 35-ton tensile steel casting; would not the author expect more than 6% elongation from an X-ray-sound casting in 0·33% carbon steel?

While we are discussing tensile tests, let us turn to Table II. Can the author explain why there is such a difference in the behaviour of these two steels? The unsound sections of nickel-chromium-molybdenum steel actually gave higher values of yield point, elongation, reduction of area, and Izod test than the sound sections. On the other hand, the unsound sections of the chromium-molybdenum steel gave a lower ductility, but again a better Izod value.

On p. 252 P the author reports an effort to design steel castings so that the production of sound castings can be achieved without the use of radiographic inspection, the X-ray method being employed only for development work. Is there not a tendency to rely on proof-load tests, along with magnetic crack detection, rather than on X-ray examinations for the final acceptance of such highly-stressed steel castings? Can the author give the cost of X-ray examinations for different thicknesses of simple steel castings, such as plates? At the bottom of p. 254 P the author says that he has examined steel forgings that have shown defects of a much more serious nature than any he has found in a steel casting. Will he give us illustrations of such defects found in forgings?

The author is to be congratulated on presenting such an interesting paper, and particularly on the excellence and clarity of the illustrations. The X-ray photographs are so clear that one wonders why the author should want to make drawings of them.

Mr. H. N. Pemberton (Lloyd's Register of Shipping, London): In the unfortunate absence of Dr. Dorey, the Chief Engineer Surveyor of Lloyd's Register of Shipping, it is my privilege to offer a few remarks on this paper. In our view the paper is of special significance to the engineering and steel-foundry industries, in that it points the only sure way of putting British steel castings on a level with the highest standards produced anywhere in the world. We at Lloyd's Register know only too well that British castings have not always compared favourably with those obtainable abroad, and, without entering on a discussion of the whys and wherefores, we have in this paper a clear and practical indication of how to improve foundry technique and how to ensure to the engineer that he has a casting which is in every way sound, and in many cases equivalent if not superior to a forging.

As Chairman of a Committee which is engaged in producing, or trying

to produce, radiographic standards for welding, I am specially interested in the author's method of standards of reference. This seems to be an ideal way of representing the results of radiographic examination, particularly for those who are not able to interpret the actual negatives. The use of radiography is at present in an early stage of development so far as steel foundries are concerned, but at Lloyd's Register we have had some twelve years' experience of X-rays for the inspection of welding, especially in pressure vessels, and, as a result of this experience, I should like to give the warning that as the use of X-rays increases in industry there is likely to be a corresponding increase in the number of people, wholly untrained in the subject, who will nevertheless have the final authority to accept or reject a casting as a result of X-ray examination. We have found it necessary to have surveyors specially trained in the interpretation of radiographs, and this procedure has been amply justified by results.

In my view, the value of X-rays in the steel foundry is not so much in the routine inspection of castings as in the working out of foundry technique. Once a satisfactory technique has been established by means of X-rays, only a check X-ray examination on selected parts, and perhaps on selected castings, need be carried out during subsequent production.

The value of this radiographic procedure may be completely destroyed if the radiographs have to be judged by someone with an inadequate knowledge of radiography, both in operating technique and in the processing of negatives. In this connection I think that some emphasis should be given to the author's words on p. 231 P to the effect that interpretation does not of itself enable a decision to be made as to the suitability of a certain casting for a particular function, the answer to such a question being a problem for the engineer and the designer. These are words which should be carefully weighed. No engineer would be justified in surrendering his right to accept or reject a casting to a radiographer, but that is not to say that the engineer should usurp the functions of the radiographer by interpreting his radiographs for him.

With this in mind, I may say that Lloyd's Register of Shipping recognizes the trend to make greater use of X-rays in engineering as a method of non-destructive testing which can be applied to materials, castings, welds, and all kinds of mechanical and electrical assemblies. So far as the Society is concerned, the advances being made in marine, oil, and electrical engineering, involving higher temperatures and pressures and more arduous conditions of service, call for an increasing degree of inspection, and in appropriate circumstances there need be no hesitation to make use of available X-ray facilities. In making use of such facilities it will be essential to ensure that the X-ray technique and equipment, and the radiographer himself, are of a sufficiently good standard to inspire confidence in the results obtained. By exercising some control over these important factors, it is felt that the engineer can be relieved of the job of judging X-ray negatives, and can accept the authoritative report of the recognized and approved radiographer.

For such a scheme to be of real value to industry it will be necessary for the X-ray equipment and radiographer to be approved by an independent authority; and for this purpose Lloyd's Register have decided to establish a list of approved industrial X-ray test houses. Firms possessing X-ray equipment who desire this authoritative recognition of their X-ray facilities will be subjected to periodical expert supervision, and will be authorized to carry out X-ray examinations and issue reports thereon which will be recognized by Lloyd's Register. In this way reliable X-ray facilities will be extended, to the general advantage of industry, and it will not be necessary for firms who wish to make occa-

sional use of X-rays to equip themselves for that purpose. At the same time Lloyd's Register and other authorities will be assured of approved and reliable means for X-ray examinations as and when required.

Mr. G. M. MICHIE (Messrs. David Brown & Sons (Huddersfield), Ltd., Sheffield) : Dr. Jackson has described clearly the manner in which radiographic examination may be employed to improve steel-foundry practice and it is felt that special credit is due to him for the importance that he has attached to diagrammatic reporting. In order that the results of radiographic examination may most effectively be brought to bear upon the production of any particular casting it is most important that they should be presented with a minimum of delay and in as clear a manner as possible to the foundry personnel concerned. The radiographic report therefore must not be unduly cumbersome, and if the radiologist has sufficient knowledge of foundry practice and of the service conditions of the part under examination, he can with advantage omit from his report reference to insignificant flaws, such as superficial defects or isolated sand inclusions, which may not in any way be connected with the casting technique employed. In other words, a concise report indicating clearly the location and nature of prevalent defects in the casting is of greater value for routine quality-control purposes than the more elaborate report which is necessary in the case of a casting under examination for purposes of final inspection and release.

The diagrammatic method of reporting upon unsoundness in cast-steel plates, as employed by Dr. Jackson, has no doubt been developed to provide a clear and concise report in this way, and it is agreed to be especially useful when applied to plate-like castings. As described it does not, however, provide direct information concerning the depth or extent of unsoundness in a direction at right angles to the diagram, an important limitation which cannot be disregarded. For instance, the results of the mechanical tests on sound and unsound plates given in Table II. of the paper, whilst appearing entirely anomalous in the light of a diagrammatic report of the type advocated, would not have done so had the X-ray report indicated that the unsound plates suffered essentially from centre-line weakness (which it is presumed was the case). An indication as to depth would probably have prevented the taking of tests from the sound areas on either side of the central shrinkage and avoided the consequent anomalous results observed.

Dr. Jackson's discussion upon the extent to which radiographically revealed defects may influence the serviceability of a casting are of great interest and correctly emphasize the need which exists for further work to be carried out on this all-important problem. The recommendation made in the paper for further study on the effect of varying types and severity of defects is fully endorsed, and the subject is one which it is felt can only be satisfactorily dealt with by active co-operation between design engineers and the foundries.

CORRESPONDENCE.

Mr. F. H. KEATING (Imperial Chemical Industries, Ltd., Billingham Division) wrote : From the point of view of the user of steel castings, the author has made a solid contribution to the development of the steel foundry. His well-designed and carefully executed experiments will be of inestimable benefit in the important task which faces the industry—creating faith in the day-to-day dependability of its products. It is no doubt realized in the steel industry that lack of faith in the soundness and

reliability of steel castings is prevalent among engineers and militates seriously against their extended use. Only by systematic development of a technique which *proves* soundness will this lack of faith be removed.

For very many engineering applications the casting has both technical and economic advantages over rival forms of construction, and only the soundness question prevents the full realization of these advantages. Ample justification for pressing forward with the application of radiography to achieve this desirable end can be had from the welding industry. In this case the extension of the use of fusion welding to important pressure vessels had been made possible only by the rigid application of radiographic examination to all welded seams on such vessels.

The industry with which the writer is connected absorbs considerable quantities of small and medium steel fittings which are used on high-pressure systems. These fittings are of such form that the economic advantages of the casting over rival forms of fabrication are obvious. From certain technical points of view the casting is also attractive. A shape can be readily and cheaply achieved which reduces pressure drop through the fitting to a minimum. Further, the stresses resulting from internal pressure demand good mechanical properties in all directions. The difficulties of achieving reasonable uniformity in massive wrought products is well known and is a problem which constantly faces the designer and metallurgist. Finally, the mass of raw steel used for the manufacture of a wrought product is generally considerably greater than that entering into the actual body of a casting and, therefore, segregation difficulties are complicated. This is particularly true of those numerous fittings which have to be forged solid and subsequently machined.

These considerations led the writer some time ago to inaugurate experiments on the application of radiography to steel castings for duty on very high-pressure systems. Willing co-operation was forthcoming from a foundry operating the necessary radiographic equipment, and two representative batches of castings were produced and thoroughly examined. In certain cases radiographic examination was co-ordinated with detailed metallographic examination of selected castings.

These initial experiments have given considerable promise and have justified the installation of a number of inspected castings on plants operating at high pressures. The performance of these castings on actual plant duty will indicate the possibilities of further extension of this interesting development.

The detailed work described by Dr. Jackson on the tracking down and elimination of defects is of great interest. Along these lines one can foresee the improvement in technique which will go far to establish confidence. Growing confidence will, in itself, assist in the extremely difficult task of the radiographer. On large and fairly complicated castings it is impossible to obtain the comprehensive "shots" described by the author on some of the simple forms used in his experiments. This necessitates dividing the casting into a number of zones and taking independent shots of each zone. Initially it is, of course, necessary to examine the whole of the casting—a matter of considerable expense. With developments along the lines indicated by the author, however, one can foresee the establishment of an agreed satisfactory technique for each type of casting and the reduction of radiographic examination to selected castings from each batch or to one or two key positions on each casting.

AUTHOR'S REPLY.

The AUTHOR wrote in reply that he was gratified by the interest which had been shown in the paper, and the general agreement which had been expressed with the major arguments contained in it.

It was interesting to have Dr. Lea's opinions as an engineer. When he speaks of the limitations of radiography, however, a number of further points must of necessity be considered. In designing a part, either a casting or a forging, allowance is made for the estimated imperfections of the material by introducing certain factors of safety. If these factors mean anything, they represent an allowance made for lack of knowledge of the finished part; for example, a casting will have a factor of safety imposed because it will be unsound to an extent which can be estimated but cannot be measured. The existence of means of measuring the unsoundness, such as radiography, should therefore reduce the lack of knowledge of the structure of the part and, to this extent, reduce the need for the particular safety factor previously employed. The inability to guarantee crack-free castings by radiography is therefore a limitation not on the use of castings, but on the extent to which the safety factor can be reduced from its present high value.

As Mr. Salmon has pointed out, a crack such as is shown in Fig. A (*d*) and (*e*) is unlikely to occur in a casting; it might, however, occur in a forging, and for this reason alone the safety factor of a forging should be higher than that for a casting, instead of the reverse.

The author agreed with Dr. Lea in regard to the need for accepting the results of static tests with reserve. They should be taken as pointers, and not as the full story of the strength of the casting. What is wanted is a true service test, and the co-operation of the engineer is needed in evolving this. The foundryman himself would be much more satisfied with a test which gave him the strength of a casting in service, than with a figure based on an arbitrary safety factor apparently obtained from castings which he puts in his museum as "horrid examples."

Dr. Mullins has pointed out an apparent anomaly in Fig. 32, *viz.*, that *SH*.6 appears worse than *SH*.8. This is partly due to difficulties of printing. The severity is largely determined by the depth of the defect, as judged by the density of the radiograph, a point which is difficult to bring out consistently in a series of prints such as are shown in Fig. 32; in addition *SH*.8 refers only to the shrinkage apparent, not to the entire area. Areas of severe shrinkage are usually not so widespread as those of less severe flaws, and it was found impossible to obtain a 2-in. square area of *SH*.8. The actual shrinkage in *SH*.8 would, in the original radiograph, be found to be more severe than that in *SH*.6.

The relative severity of the defects shown by two radiographs was decided on the general appearance of the diagrams for the two. It was not considered that the method was good enough to justify applying the refinements, such as total or weighted areas, introduced by Dr. Mullins. The general agreement with the author's results which Dr. Mullins finds after using these refinements is a gratifying confirmation of the soundness of the method for supplying information to the layman.

The particular plates referred to were not sectioned, as the additional printing difficulties did not warrant it, but sufficient similar sections were made to confirm the radiographic findings.

Mr. Salmon referred to the size of defects as revealed by radiography, and to film faults being mistaken for defects. The best radiographs are taken with the source of X-rays a considerable distance from the casting and film, so that the rays are effectively parallel when they reach them. If the distance is great enough, the defects will appear at their true size

on the radiograph. For various reasons, mainly economy in time, the X-ray tube may be placed near the casting; in this case the rays reaching the film will not be parallel, but will clearly be diverging, and will thus lead to a distorted picture. The effect is best illustrated by comparing the shadow of a hand caused by the sun with that produced by a lamp placed only a few inches away. Film faults are usually easy to distinguish from casting defects, though Mr. Salmon is correct in saying that the radiograph and casting should be examined together.

The obvious clink and the obvious pull or hot-tear can be distinguished. Doubt arises at times in visual inspection as to whether a particular crack is a clink or a pull, and the same difficulty arises with radiographs. The radiograph cannot be taken as a sure means of differentiating between the two types of crack.

There is no doubt that, properly used, the magnetic-powder method is more certain than X-rays for detecting fine surface cracks. When the cracks do not reach the surface, however, the magnetic crack detector becomes less valuable and may fail to show quite large defects which are easily visible on the radiograph.

In the schematic representation of defects 10 grades are used, but for easy reference three grades, shown as plain (0–3), hatched (4–6), and cross-hatched (7–10) areas, are actually employed. Any system of drawing, except in colour, has drawbacks when representing many factors by a limited number of signs. As it is easier to draw full lines than many series of code lines, this method has been adopted, and is considered to be in general quite clear. The interpretation of a radiograph is not a simple matter, and the untrained examiner can draw entirely erroneous conclusions. For this reason the schematic diagram produced by the radiologist has a great advantage. In many cases the runners and risers of the casting do not appear on the radiograph and usually cannot easily be recorded on it. The radiograph is better used as confirmation of the findings given in the final report containing all the relevant matter.

The relation between the fluidity and castability of steel and the X-ray soundness of castings is a complicated one and requires much further work to be carried out.

The defects noted in the fan-blade casting made from a steel which began to rise in the course of the heat, and the similar type of defect found in the cap castings, were clearly due to unsatisfactory steel and when recognized as such were corrected by the usual attention given to this type of steel. The making of a semi-killed steel with sufficient gas in it to avoid shrinkage but not sufficient to cause gas holes may be possible, but the author would consider it risky and probably unsatisfactory.

The bars tested of which radiographs are shown in Fig. 51 were radiographed before testing, but the radiographs were unsatisfactory for reproduction; the defects revealed were very similar to those illustrated. All these tests were taken from cast plates specially made to be defective, and sufficient sound material could not be found to permit a test-piece to be cut. Test-piece No. 1 was not sound, but was the best of the series given.

The variations shown in Table II. bear out the contention already made, that a single test from a casting is not typical of the casting as a whole, and can be correlated with the X-ray results only if the full thickness of the casting is used. As Dr. Mullins and Mr. Pemberton have pointed out, the method of employing X-rays which is most advantageous is as a development tool in the foundry. When a satisfactory technique has been evolved, only routine control is required, and X-ray examination of every casting may be impracticable for this purpose. The cost of X-ray examination is difficult to estimate, as it varies so much with the layout

and type of plant. The thickness of the plate determines the exposure required, which increases exponentially with thickness, so that near the limit of the particular set, a small increase in thickness of plate may increase the exposure time to an uneconomical extent.

An idea of the extent of the trouble in the defective forging referred to in the paper can be gathered from the fact that the 2-in. dia. part of the forging at its point of maximum fatigue stress had a transverse internal burst extending to within $\frac{1}{8}$ in. of the outside in all directions. This was discovered by accident, forgings not being normally radiographed, as they are considered to be sound.

The author is pleased to see the value set upon radiography by Lloyd's Register, and he feels that Mr. Pemberton's remarks on the training of surveyors and the use of X-rays in the foundry are well justified. The provision of a series of X-ray test-houses should be a great help in extending the use of X-rays to firms who would otherwise be unable to benefit by their use.

Mr. Michie has raised an interesting point in saying that the depth of a defect should be indicated in the diagrammatic representation of that defect. Such an extension of the present method would doubtless be an advantage and should be borne in mind in planning those extensions to the system which are needed.

Mr. Keating has brought forward several striking points in favour of castings, and the author is interested in his experiments on applying castings to very high-pressure work after suitable radiography. The reduction in the number of radiographs taken on one casting, which was suggested by Mr. Keating in his last paragraph is, of course, carried out in practice with special castings, such as those for aircraft. On the type of castings shown in Figs. 58 to 61 eight shots were needed to examine the casting satisfactorily, and these required about a week. With experience of the position of defects, and with improvement of foundry technique, the number of shots was reduced to two, and these were standardized so that an examination could be carried out in a fraction of an hour.

EXAMINATION OF TWO INGOTS OF FREE-CUTTING STEEL, ONE CONTAINING LEAD AND THE OTHER LEAD-FREE.*

BY C. S. GRAHAM (MESSRS. JOHN LYSAGHT, LTD., SCUNTHORPE, LINCS.).

(Figs. 2 and 3 = Plates XXIX. and XXX.)

Paper No. 31/1945 *of the Committee on the Heterogeneity of Steel Ingots.*

SYNOPSIS.

The segregation in an ingot of leaded steel has been compared with that in a non-leaded ingot from the same cast. Little difference in the chemical compositions at the Heterogeneity of Steel Ingots Committee's standard positions was found, with the exception of a small, but possibly significant, reduction in the oxygen content in the lead-bearing ingot.

The lead itself was evenly distributed, except at the extreme bottom of the ingot.

On the examination of an axial section, the leaded ingot showed less segregation of sulphur and a thinner columnar zone.

THE quality of the free-cutting steel with which this paper is concerned is covered by British Standard Specification No. 32, Grade 4, which lays down the following composition :

Carbon	Not greater than 0·20%
Silicon	0·10% max.
Sulphur	0·2–0·3%
Phosphorus	0·1% max.
Manganese	0·6–1·2%

Large quantities of this type of steel with lead additions (0·15–0·30%) have been used since its introduction in 1938. The effect of the lead in improving the machinability is a well-established fact based on a large number of reports from actual users.

At the instigation of the Ingot Committee, two ingots, one lead-free and the other lead-bearing, were selected from a cast of steel to ascertain the effect of lead addition on the chemical composition and segregation.

STEELMAKING DATA.

The cast of steel from which the two ingots were selected was made in a basic-lined open-hearth furnace of 65 tons capacity.

The manufacturing data at the finishing stages of this particular cast were as follows :

Analysis of the Metal and Slag before Tapping gave the following figures :

Metal : 0·05% carbon. 0·250% sulphur. 0·050% phosphorus. 0·10% manganese.
Slag : 17·0% total iron. 10·0% SiO_2. 9·6% P_2O_5.

The temperature of the metal, taken with an immersion pyrometer, was 1605° C.

* Received January 26, 1945. This paper is published by authority of the Committee on the Heterogeneity of Steel Ingots. The views expressed are the author's, and are not necessarily endorsed by the Committee as a body.

Ladle Additions were made as follows :

Ferro-manganese (low-carbon) . . .	1 ton 2 cwt.
Rock sulphur	60 lb.

Teeming and Sampling.—The steel was top-poured through a $1\frac{1}{2}$-in. nozzle into moulds of the following internal dimensions :

Top of mould	19 in. × 19 in.
Bottom of mould	17 in. × 17 in.
Total height	6 ft. 1 in.

The cast consisted of 28 ingots, and the two ingots chosen were the fourteenth and fifteenth in the order of teeming. Both ingots weighed 47 cwt.

Samples taken during teeming gave the following analyses :

	Carbon.	Sulphur.	Phosphorus.	Manganese.
At the 3rd mould . . .	0·09%	0·29%	0·050%	1·10%
At the 28th mould . . .	0·09%	0·28%	0·054%	1·09%

Lead Addition to Lead-Bearing Ingot.—Fifteen pounds of lead shot were added to the metal stream during teeming. The addition was started when the metal in the mould reached the 3-in. mark.

Examination of the Ingots.

Both ingots were sectioned along the vertical axis. The chemical analyses were made from samples taken from one-half of each ingot at the standard positions as shown in Fig. 1. The results of these analyses are given in Table I.; the oxygen was determined by the aluminium method. The oxygen content at position H in the lead-bearing ingot is omitted, as it was impossible to secure samples sound enough to give results which could be duplicated.

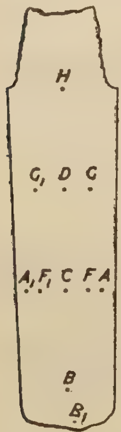


Fig. 1.—Positions from which Samples were Taken.

Ignoring the analysis at point H, which is in the hot-top region, there is practically no difference between the chemical compositions of the two ingots, with the exception of the oxygen contents, which are of a slightly lower order in the lead-bearing steel. Whether this slightly lower oxygen content is of any significance or not will be a subject of further investigation.

Additional lead estimations were made to ascertain its quantitative distribution. These samples were taken in horizontal planes from the edge to the centre of the lead-bearing ingot at four different levels, *viz.*, B_1, which corresponds to 4% from the bottom of the ingot, and C, D and H. The results of these individual lead estimations are given in Table II.; they show the lead to be evenly distributed.

Sections taken from the same four positions (B_1, C, D and H) of the ingot were heated to 700° C. in a reducing atmosphere; in no case was there any exudation of lead visible to the naked eye. The only segregation of lead was at the extreme bottom of the ingot. The section taken 4% from the bottom was clear of segregation.

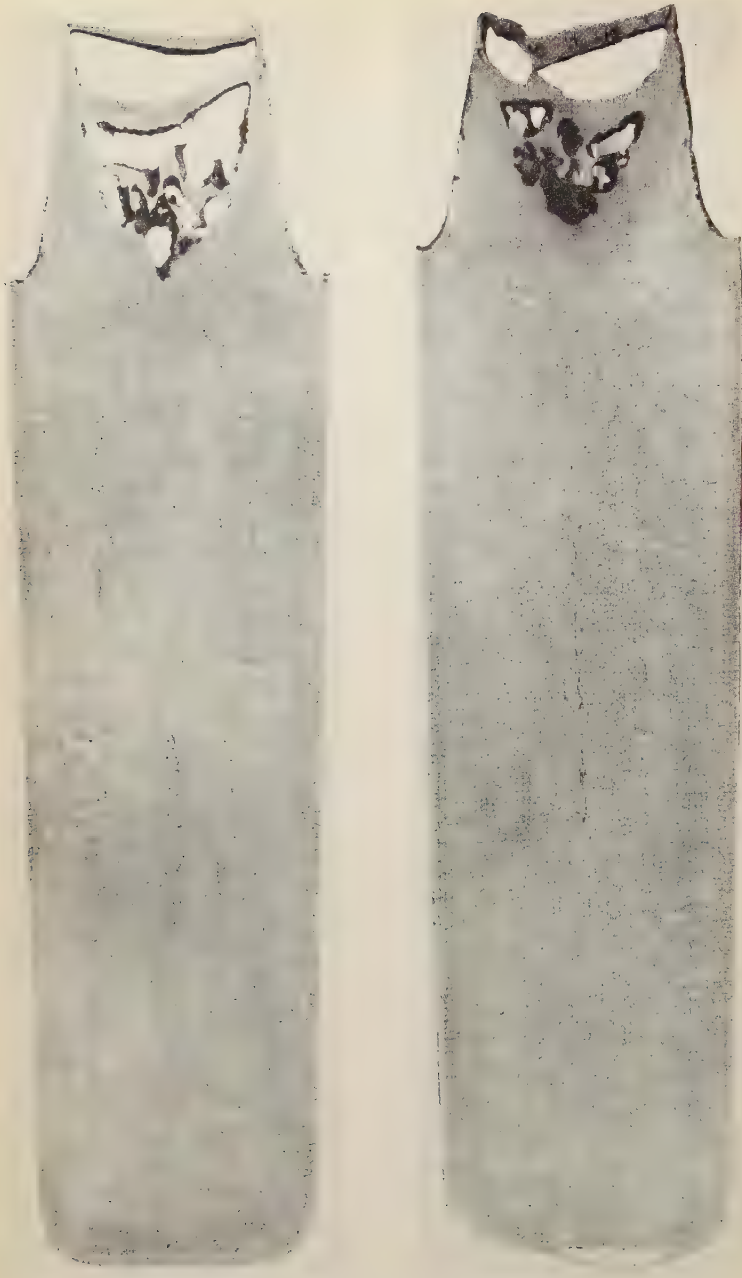
One half-section of each ingot was prepared for sulphur printing and macro-etching. The sulphur prints are reproduced in Fig. 2, and show less segregation in the lead-bearing ingot than in the lead-free ingot. The macrostructure (nitric acid etch) of the two ingots, illustrated in Fig. 3, showed the only difference of importance to be the difference in depth of the columnar zone, which was $\frac{1}{2}$ in. less in the leaded ingot.

(*a*) Lead-Free Ingot. (*b*) Lead-Bearing Ingot.

FIG. 2.—Sulphur Prints of Lead-Free and Lead-Bearing Ingots.

[*Graham.*

[*To face p.* 274 P.



(*a*) Lead-Free Ingot. (*b*) Lead-Bearing Ingot.

Fig. 3.—Macrostructure of Lead-Free and Lead-Bearing Ingots.

[*Graham.*

[*To face p.* 275 P.

TABLE I.—*Chemical Analysis.*

Position.	C. %.	Si. %.	S. %.	P. %.	Mn. %.	O_2. %.	Pb. %.
			Lead-Free Ingot.				
H	0·097	0·019	0·355	0·064	1·10	0·019	...
D	0·080	0·019	0·272	0·052	1·09	0·022	...
G	0·090	0·019	0·290	0·054	1·10	0·015	...
G_1	...	...	...	...	...	0·011	...
C	0·080	0·028	0·241	0·048	1·08	0·027	...
F	0·085	0·019	0·280	0·054	1·11	0·016	...
F_1	...	...	...	...	...	0·009	...
A	0·090	0·023	0·294	0·052	1·09	0·013	...
A_1	...	...	...	...	...	0·010	...
B	0·080	0·028	0·278	0·053	1·10	0·040	...
			Lead-Bearing Ingot.				
H	0·090	0·019	0·315	0·056	1·10	...	0·15
D	0·090	0·019	0·270	0·052	1·09	0·016	0·15
G	0·090	0·019	0·295	0·051	1·09	0·014	0·16
G_1	...	...	...	...	...	0·011	...
C	0·090	0·019	0·251	0·046	1·08	0·023	0·15
F	0·085	0·019	0·267	0·047	1·10	0·010	0·16
F_1	...	...	...	...	...	0·013	...
A	0·090	0·019	0·285	0·052	1·10	0·010	0·17
A_1	...	...	...	...	...	0·009	...
B	0·090	0·019	0·276	0·049	1·09	0·013	0·17

TABLE II.—*Lead Contents at Four Different Horizontal Planes from Edge to Centre of the Ingot.*

Sample No.	Lead Content, %, at Position—			
	B_1.	*C.*	*D.*	*H.*
1. Edge of ingot . .	0·15	0·15	0·14	0·18
2	0·27	0·15	0·16	0·19
3	0·18	0·15	0·15	0·19
4	0·18	0·16	0·16	0·19
5	0·20	0·16	0·16	0·19
6	0·18	0·18	0·15	0·15
7	0·18	0·16	0·16	0·15
8	0·18	0·15	0·15	0·15
9	0·15	0·15	0·16	0·16
10	0·18	0·15	0·15	0·16
11	0·17	0·15	0·17	0·15
12	0·18	0·15	0·15	0·15
13	0·15	0·15	0·16	0·15
14	0·17	0·15	0·15	0·15
15	0·15	0·15	0·15	0·15
16	0·18	0·15	0·15	0·15
17	...	...	0·15	0·15
18. Centre of ingot . .	...	...	...	0·15

The Author wishes to express his thanks to Wing Commander J. B. R. Brooke, Director and General Manager of Messrs. John Lysaght, Ltd., Scunthorpe, for permission to publish the results of this work.

[This paper was discussed jointly with the following two by T. H. Schofield on "The Microscopical Examination of Samples of Lead-Bearing and Lead-Free Steels and Ingot Irons" and by W. E. Bardgett and R. E. Lismer on "Mode of Occurrence of Lead in Lead-Bearing Steels and the Mechanism of the Exudation Test."

THE MICROSCOPICAL EXAMINATION OF SAMPLES OF LEAD-BEARING AND LEAD-FREE STEELS AND INGOT IRONS.*

By T. H. SCHOFIELD, M.Sc. (The National Physical Laboratory, Teddington).

(Figs. 1 to 8 = Plates XXXI. and XXXII.)

Paper No. 32/1945 *of the Committee on the Heterogeneity of Steel Ingots* (*submitted by Dr. N. P. Allen*).

Synopsis.

Microscopical examination of samples of lead-bearing steels and ingot irons has revealed small characteristic inclusions which are not present in similar lead-free materials. Using a modified polishing technique, these inclusions appear white, and an etching test indicates that they consist of or contain lead. The inclusions are removed by mercury at 100° C.

Introduction.

The condition in which lead is present in lead-bearing steels has been the subject of much investigation and speculation. The work of Isaac and Tammann [1] and Tammann and Oelsen [2] has shown that lead is almost insoluble in liquid and solid iron. If it is inferred that lead is almost insoluble in steel, it would appear that most of the lead in lead-bearing steels is present as finely dispersed particles of lead and/or constituents containing lead. Daniels [3] gives some evidence in support of the existence of a compound $FePb_2$, but no worker appears to have confirmed his observations. Microscopical examination of lead-bearing steels has, however, failed to detect inclusions which could be identified with lead,[4, 5, 6] although an etching test suggested by von Vegesack [7] indicates that certain inclusions in these steels contain lead. Radiographs of lead-bearing steel ingots [4] show segregation of lead and microradiographs [8] of wrought lead-bearing steels indicate that fine inclusions of lead or constituents containing appreciable amounts of lead are present. Nead, Sims and Harder [9] have reported that when lead-bearing steels are heated to 900° F. (482° C.) in hydrogen, globules of lead sometimes appear near the inclusions of manganese sulphide, and Whiteley [6] states that when such steels are heated to 400° C. in a neutral atmosphere, lead oozes and forms droplets on the surface. This phenomenon is the basis of the exudation test which a steel is required to pass before being accepted as lead-bearing.[5]

Scope of Investigation.

This investigation was concerned with the metallurgical examination of six materials, four lead-bearing and two lead-free. Four of these materials were steels supplied by the late Dr. Swinden, of The United Steel Companies, Ltd., on behalf of the Heterogeneity of Steel Ingots Committee, for examination by means of the electron microscope, and two were samples from ingot irons prepared by Messrs. Guest Keen Baldwins Iron and Steel Co., Ltd., and supplied through the Armament Research Department. The results of an investigation of the macrostructure and

* Communication from the National Physical Laboratory, received January 27, 1945. This paper is published by authority of the Committee on the Heterogeneity of Steel Ingots. The views expressed are the author's, and are not necessarily endorsed by the Committee as a body.

TABLE I.—*Details of the Materials Examined.*

Source of Material.	Cast No. and Specimen Mark.	Description of Material.	Size and Location of Samples.	Chemical Analysis. Carbon. %.	Silicon. %.	Sulphur. %.	Phosphorus. %.	Manganese. %.	Nickel. %.	Chromium. %.	Lead. %.
United Steel Companies, Ltd.	31/723. Specimens *A* and *B*.	No. 1 ingot of lead-bearing carbon steel.	Two specimens of each steel 1 in. × 1 in. × ½ in. from near rim and centre of mid transverse section of ingot.	0·27	0·145	0·032	0·027	1·02	...	...	0·16
	35/7295. Specimens *C* and *D*.	No. 2 ingot of lead-bearing carbon steel.		0·27	0·17	0·031	0·026	0·80	0·10	0·065	0·16
	*S*4405.	Wrought lead-free nickel-chromium steel.	Section of square billet 5 in. × 5 in. × 2½ in.	0·29	0·20	0·008	0·023	0·47	4·37	1·31	...
	*S*4405.*HM*.	Wrought lead-bearing nickel-chromium steel.									0·26
Messrs. Guest Keen Baldwins Iron & Steel Co., Ltd.	...	Lead-free ingot iron (*TTQ*).	Specimen taken about 8 in. from top and near rim of ingot.	0·02	0·01	0·024	0·009	0·05	...	...	...
	...	Lead-bearing ingot iron (*TTP*).		0·02	<0·01	0·043	0·009	0·04	...	...	0·24

FIG. 1.—Lead-Bearing Carbon Steel Specimen *D* (Ingot No. 2). Unetched. × 500.

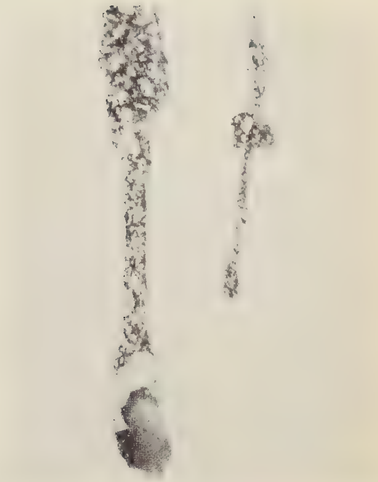


FIG. 2.—Lead-Bearing Nickel-Chromium Steel *S4405.HM.* Unetched. × 1500.

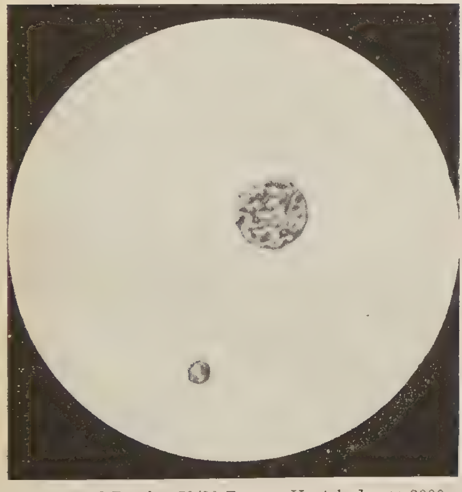


FIG. 3.—Lead-Bearing 70/30 Brass. Unetched. × 2000.

FIG. 4.—Lead-Bearing Nickel-Chromium Steel *S4405.HM.* Unetched. × 1250.

PLATE XXXII.

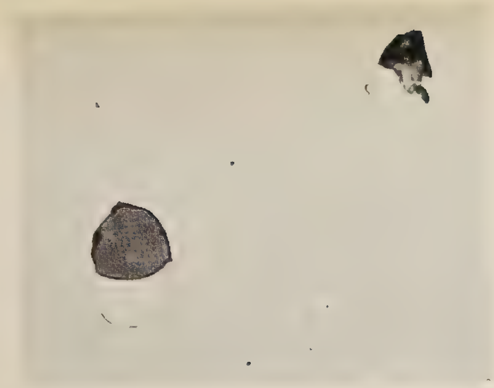


FIG. 5.—Lead-Bearing Carbon Steel Specimen *D* (Ingot No. 2). Unetched. × 1500.

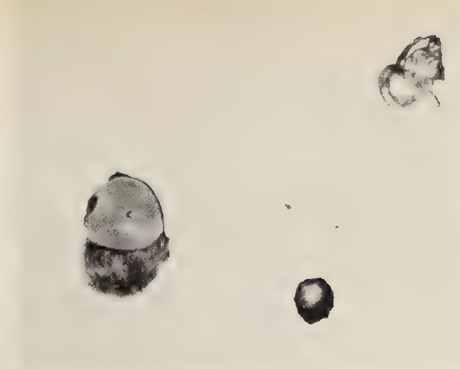


FIG. 6.—Same Field as Fig. 5 after Mercury Treatment. × 1500.

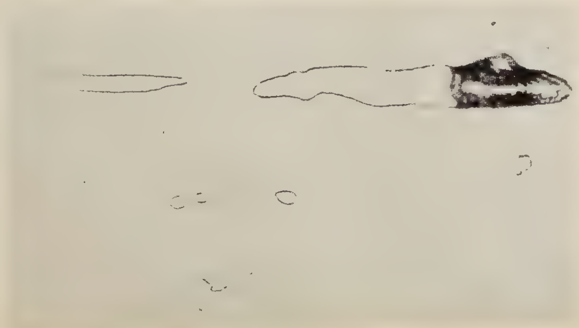


FIG. 7.—Lead-Bearing Nickel-Chromium Steel *S4405.HM*. Unetched. × 1500.

FIG. 8.—Same Field as Fig. 7 after Mercury Treatment. × 1500.

[*Schofield.*

[*To face p.* 279 P.

segregation of these ingots irons have already been published.[4] Further details and the chemical analyses of all the materials are given in Table I.

Microscopical Examination.

Sections of specimens *B* and *D* taken from near the centre of ingots Nos. 1 and 2, longitudinal sections cut about ½ in. from the middle of one of the sides of each of the billets *S*4405 and *S*4405.*HM* and transverse sections taken about 8 in. from the top and near the rim of the ingots *TTQ* and *TTP* were prepared for microscopical examination, using the technique normally employed at the National Physical Laboratory for steels. Briefly, this consists of rubbing the sawn or filed surface on coarse emery cloth, then on progressively finer grades of emery cloth and paper down to the 00 (Hubert) grade. The 00 scratches are removed on a rotating pad of broadcloth charged with a commercial grade of fine alumina and lubricated with water. The surface is finished on a similar pad, using chromic oxide and water. The chromic oxide is prepared by igniting pure ammonium bichromate.

Using this technique all the lead-bearing materials showed inclusions having a rough and mottled appearance, in addition to the inclusions normally present in steels. The inclusions were sometimes free, but were frequently attached to other inclusions. Figs. 1 and 2 show, respectively, typical inclusions in the carbon steel (specimen *D*, ingot No. 2) and the wrought nickel-chromium steel *S*4405.*HM*. None of these inclusions was seen in the lead-free materials, but inclusions of similar appearance were observed in a lead-bearing 70/30 brass (Fig. 3).* A confirmatory examination of one of the specimens by Dr. Whiteley revealed inclusions of similar form to, but lighter in colour than, those obtained by the author. As it seemed that the darker and mottled appearance of the inclusions might be due to their partial removal during polishing, a modified polishing technique was tried.

The specimens were polished on velveteen by hand directly from the 00 stage, using specially prepared alumina and water made slightly alkaline by adding a few drops of ammonia solution. The alumina was prepared by slowly heating aluminium sulphate (A.R.) to Seger cone 04*A* (nominally 1020° C.) in a platinum or silica vessel and cooling in the furnace.

Using this technique the inclusions appeared white and smooth (Figs. 4, 5 and 7). The time necessary to polish a specimen in this way was about 2 hr. or more, and some experience was necessary before satisfactory results were consistently obtained.

Etching Tests.

The polished specimens were etched in a reagent † suggested by von Vegesack[7] for studying the distribution of lead in steel. Under normal illumination the inclusions appeared greenish yellow, and under dark-ground illumination bright yellow. This reagent resulted in an apparent enlargement of the inclusions, probably owing to the precipitation of minute crystals of lead iodide round them. Etched in a 10% aqueous solution of chromic acid for about 10 min., the inclusions had a reddish appearance under normal illumination.

Polished specimens of the lead-bearing materials were treated for 1 hr.

* The author's acknowledgment is due to Mr. D. McLean, of the Armament Research Department, for this photomicrograph. The specimen had been first polished on a revolving pad, and finally by hand on velveteen, using magnesia and copper ammonium chloride solution in both stages.

† Prepared by adding 2–1 volumes of nitric acid (sp. gr. 1·40) to 98–99 volumes of absolute alcohol with an excess of crystalline potassium iodide. The excess of potassium iodide is filtered off and the filtrate applied to the polished surface of the specimen for 10–30 sec. The surface is washed in alcohol and dried in hot air, and must not be touched before examination.

in mercury held at 100° C. This treatment removed the white inclusions. Figs. 5 and 7 show, respectively, inclusions in the carbon steel (specimen *D*, ingot No. 2) and the wrought nickel-chromium steel *S*4405.*HM* as polished, and Figs. 6 and 8, respectively, the same fields after treatment in mercury. Other inclusions were not removed or attacked by this treatment.

Heating Experiments.

Some experiments were carried out to ascertain whether any significant change in the appearance of the inclusions occurred when a polished specimen was heated slightly above the melting point of lead (327° C.). A specimen of steel *S*4405.*HM* was chosen for these experiments, as the inclusions were relatively large. The specimen was examined as polished, then after heating in a vacuum (0·0001–0·005 mm. of mercury) for 15 min., first at 324° and then at 330° C. The specimen was cooled to room temperature between the two heatings. A slight change in the appearance of the inclusions in a marked field was observed, particularly after heating at 330° C., but the change did not provide sufficient evidence to warrant the conclusion that melting had taken place between 324° and 330° C.

Summary and Conclusions.

(1) All the lead-bearing materials show inclusions which have a rough and mottled appearance after using a normal polishing technique for steels; these inclusions appear white and smooth after using a modified polishing technique.

(2) Inclusions of this type are not observed in the lead-free materials.

(3) When polished in the normal way the inclusions resemble certain inclusions in lead-bearing brass.

(4) The rough and mottled appearance of the inclusions after normal polishing indicates that they have been partly removed, *i.e.*, they are relatively soft or brittle.

(5) An etching test shows that inclusions consist of or contain lead.

(6) The inclusions are removed by mercury at 100° C., which suggests that they consist of a metal or metallic compound that alloys with mercury.

(7) Heating experiments do not provide definite evidence that the inclusions had melted at a temperature just above the melting point of lead.

The author desires to acknowledge the skill shown by his colleague Mr. A. E. Bacon in polishing the specimens by this modified technique.

The work described above has been carried out as part of the research programme of the National Physical Laboratory, and this paper is published by permission of the Director of the Laboratory.

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[This paper was discussed jointly with the preceding one by C. S. Graham on "Examination of Two Ingots of Free-Cutting Steel, one Containing Lead and the other Lead-Free" and the following one by W. E. Bardgett and R. E. Lismer on "Mode of Occurrence of Lead in Lead-Bearing Steels and the Mechanism of the Exudation Test."]

MODE OF OCCURRENCE OF LEAD IN LEAD-BEARING STEELS AND THE MECHANISM OF THE EXUDATION TEST.*

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(Work undertaken under the direction of the late Dr. T. Swinden.)

(Figs. 1 to 55 = Plates XXXIII. to XL.)

Paper No. 33/1945 *of the Committee on the Heterogeneity of Steel Ingots.*

Synopsis.

As relatively little is known about the mode of occurrence and distribution of lead in lead-bearing steel, an extensive examination has been carried out on high-sulphur and manganese–molybdenum wrought steels and on 0·25% carbon, 1% manganese steel ingots.

Normal methods of microscopic examination failed to reveal the presence of lead as discrete particles, but an electrolytic method revealed particles believed to be lead, not visible in unetched specimens.

A new electrographic method is described which produces a clear pattern of the lead or lead-bearing particles in the ingot section. Interesting features were brought to light by microscopical examination of samples during heating.

The examination of two leaded-steel ingot sections, described in an Appendix, reveals the normal features of ingot segregation. The distribution of lead is generally uniform, with an absence of massive lead inclusions, except at the extreme base of the ingot. With this exception, the ingot may be regarded as being uniformly lead-bearing, the lead being distributed in microscopic form.

When lead is added to steel, it may occur in a segregated form towards the bottom of the ingot [1] and in this form is readily detected by means of the exudation test which is used as a pass test for lead-bearing steels.[2] Lead may also occur in association with inclusions in unsegregated material.[3, 4] From the lack of evidence of the existence of lead as discrete particles and on the assumption that lead is insoluble or only slightly soluble in steel, it has been inferred that lead is probably also present in a submicroscopic state of division. The presence of discrete lead or lead-bearing particles is, however, clearly revealed by a method of lead-printing recently communicated to us by Mr. W. B. Wragge. This method employs caustic soda as the printing solution and sodium sulphide as the developing solution.

The first part of the present investigation was confined to the microscopical examination of non-leaded and leaded wrought materials of high-sulphur and manganese–molybdenum qualities and examination of the lead exudation test by means of a high-temperature microscope. A short cinema film was obtained illustrating the behaviour of the lead segregates on heating.

The second part mainly consisted of an examination of samples taken from various positions in leaded 0·25% carbon, 1·0% manganese steel

* Received February 9, 1945. This paper is published by authority of the Committee on the Heterogeneity of Steel Ingots. The views expressed in it are the authors', and are not necessarily endorsed by the Committee as a body.

ingots with the object of making a closer investigation into the form of existence of the lead and determining its distribution in the ingots. A film was also obtained, at higher magnification than the previous one, showing the exudation of lead, the samples being taken from unsegregated material.

Part I.—Leaded and Non-Leaded Wrought Steels.

Microscopical Examination.

(a) *High-Sulphur Steels.*

A sample of leaded high-sulphur steel (0·18% lead) has been compared with a sample of standard high-sulphur quality. The samples were in the form of $1\frac{5}{8}$-in. dia. bar, and their analyses were as follows:

	Leaded.	Non-Leaded.
Carbon. %	0·13	0·10
Manganese. % . . .	1·10	0·82
Silicon. %	...	Trace
Sulphur. %	0·235	0·299
Phosphorus. % . . .	0·054	0·063
Nickel. %	0·33	...
Lead. %	0·18	Nil

The inclusions in the standard high-sulphur steel consisted of elongated monophase sulphides, showing a tendency to be streaky (Fig. 1), with very occasional duplex oxide–sulphide streaks (Fig. 2).

The inclusions in the leaded high-sulphur steel were generally identical with those of the standard high-sulphur steel. The monophase sulphides (Fig. 3) showed only a small tendency towards streakiness. The duplex streaks (Fig. 4) were slightly more numerous, and occasional small segregates of oxide particles were evident. Isolated, fairly large segregates of complex non-metallic inclusions were present, similar to those shown in Fig. 5. The segregates consisted of a dark-grey matrix of silicate containing light-grey globules or particles of sulphide and darker-grey particles of cubic form, probably oxide. Etching tests carried out in an acetic acid–sodium chromate solution gave these segregates a bright-yellow, vitreous appearance under polarised light, thereby indicating that the segregates contained lead compounds.

The presence of lead within the structure, either at the grain boundaries or within the grains, could not be detected.

(b) *Manganese–Molybdenum Steel.*

Samples from leaded and non-leaded ingots of acid open-hearth manganese–molybdenum steel of the following analysis were examined:

Carbon . . .	0·36%	Phosphorus . . .	0·045%
Manganese . . .	1·43%	Molybdenum . .	0·43%
Silicon . . .	0·225%	Nickel	0·16%
Sulphur . . .	0·032%	Chromium . . .	0·11%

The ingots were rolled to $4\frac{1}{4} \times 2\frac{7}{8}$-in. slabs, and samples were taken from the slab discarded from the bottom of each ingot, which in the case of the leaded steel was expected to contain lead segregates. A lead-print obtained by the use of ammonium acetate and hydrogen sulphide solutions,* on a section through the bottom slab of the leaded ingot, is reproduced in Fig. 6. Segregation of the lead had occurred to a marked extent in an area $\frac{1}{4}$–$\frac{1}{2}$ in. from the outer surface of the slab. Isolated segregates were evident within the central area of the section.

Microscopical examination of sections through the segregated area

* Private communication from Mr. W. B. Wragge.

revealed the presence of two types of massive streaks. The main type, shown in Fig. 7, was composed of a number of phases (Fig. 8):

(1) A mottled soft matrix of metallic lead.
(2) Mixed sulphide areas, grey in colour and of cubic shape.
(3) Mixed oxide areas, dark grey in colour and more globular in shape.
(4) Mixed sulphide areas, very light grey in colour and irregular in shape.
(5) Very occasional particles of free copper precipitated towards the outer edge of the metallic streak.
(6) Many irregular-shaped particles of steel.

The second type of streak associated with the lead segregates was vitreous and contained no *metallic* lead. Figs. 9 and 10 show the appearance of the segregates, which had smoother edges and were more regular than the metallic segregates. The segregate shown in Fig. 9 consisted of the following:

(7) A dark-grey matrix probably of mixed oxides.
(8) Irregular-shaped globules of mixed sulphide, light grey in colour.
(9) Slightly darker-grey needles having the appearance of sulphides, but with a structure unlike that of either sulphides or oxides.
(10) Dark-grey cubic particles of mixed oxide.

A further segregate, shown in Fig. 10, consisted of cubic particles in a matrix of slightly darker colour. The whole mass polarised brightly and was probably composed entirely of silicates. After etching with hydrofluoric acid, attack took place on the matrix and to a lesser extent on the cubic particles, with the appearance of numerous granules of silica (Fig. 11). Etching tests carried out in acetic acid–sodium chromate solution showed that these segregates contained lead compounds. Numerous small streaks of metallic lead were present in the segregated area.

Examination of the non-leaded slab failed to reveal the presence of segregates or streaks of the type described above. Differences in the microstructures of the non-leaded and leaded slabs are shown in Figs. 12 and 13. The non-leaded slab was markedly banded and had a coarse Widmannstätten appearance. The leaded slab was only slightly banded; it was more uniform, with no evidence of a Widmannstätten pattern, and the sorbitic pearlite was less dense and more spheroidised. No evidence of the presence of lead within the structure could be detected.

The Mechanism of the Lead Exudation Test.

A reliable method of determining the extent of lead segregation in a sample of leaded steel is by means of the "exudation test." This test consists of heating a section of the steel at a temperature of 650–700° C. for a period of 10 min. for each inch of thickness, 2 in. being the recommended minimum thickness. If segregated lead is present, this treatment results in the exudation of the lead on the surface in the form of globules or beads. The appearance of numerous beads of exuded lead on a section through the bottom-discarded billet of the manganese–molybdenum steel is shown in Fig. 14.

Examination of the behaviour of lead segregates during heating to the temperature of the exudation test, has been carried out by means of high-temperature microscopy, and a motion film has been taken illustrating the actual exuding of the lead. The complete film, showing the set-up of the microscope and furnace, and actual exudation of the lead, is available.*

* This film was shown when the paper was presented at the Annual General Meeting of the Iron and Steel Institute on July 11, 1945.

The samples for examination were cut longitudinally from the segregated area towards the outer surface of the bottom-discarded billet of manganese–molybdenum steel, as shown in Fig. 15. The samples were 2 in. in length and machined to ⅜ in. dia. The end of the sample was polished. A thermocouple was attached to the specimen, which was inserted in a small furnace, with the polished surface upwards. The microscope, which had a 2-in. objective but no eye-piece, was fixed above the furnace. A Zeiss Mikrophot attachment and cine-camera were assembled above the microscope, and the surface of the sample was observed whilst the film was being exposed.

A number of tests were made and recorded, and in each test slight exudation of the lead segregates was observed at a temperature as low as 30° C. The lead beads suddenly spurted out on to the surface of the specimen at a temperature of 235° to 240° C. The appearance of the polished surface after heating to 240° C. is shown in Fig. 16. At a temperature between 350° and 400° C., the beads spread out over the surface and coalesced.

Chemical analysis carried out on beads knocked off the surface of a sample heated to 240° C., showed the beads to be of the order of 98–99% lead. On heating to above 240° C., the bright metallic lead beads soon became coated with lead oxide, first bright red in colour, changing with increased temperature to yellow and then to grey.

The character of the lead streaks did not appear to have undergone any change as a result of heating to 400° C. The lead beads, which were up to 0·05 in. in dia., contained particles of sulphide, oxide and steel in a metallic lead matrix (Fig. 17). A number of test samples were examined after heating, and although one would expect to find cavities left by the exuded lead, none was present. Slight distortion of the structure of the steel adjacent to the lead streak was evident, as may be seen by comparison of Figs. 18 and 19, showing the structure before and after heating to 240° C. It was found that only the metallic streaks of lead were exuded, and that the non-metallic, vitreous oxide–silicate streaks remained unchanged. In Fig. 17, it will be seen that the exuding of the lead has caused deformation of the polished surface of the sample below the bead.

Part II.—Leaded 0·25% Carbon, 1·0% Manganese Steel Ingots.

Samples Examined.

The examination was carried out on sections from two leaded steel ingots, particulars of which are as follows :

Cast No.	35/7295	31/723
Type of ingot	75 cwt., 22½ in. cast uphill, narrow end up.	84 cwt., 23 in. cast uphill, wide end up.
Analysis (2nd pit sample) :		
Carbon. %	0·25	0·27
Manganese. %	0·88	1·02
Silicon. %	0·175	0·145
Sulphur. %	0·033	0·032
Phosphorus. %	0·032	0·027
Lead. %	0·165	0·17

Sections were taken from a 1-in. slice cut longitudinally down the middle of each ingot. Sections 1 in. in width, extending from the outer surface to the centre of each ingot, were cut at each of three positions :

(a) Approximately 15 in. from the theoretical top (top of chill).
(b) Middle position.
(c) Approximately 15 in. from the bottom.

Further details of these ingots are given in the Appendix.

Methods of Examination.

The sections were examined after etching in 2% nitric acid and showed typical as-cast structures with no evidence of the presence of lead within the grains or at the grain boundaries. Normal etching reagents failed to reveal any structural variation associated with the lead. The reagents tried comprised picric acid (with and without hydrochloric acid), hydrochloric acid and nitric acid (in alcohol, water and glycerine respectively), copper reagents, alkaline sodium picrate, ammonium persulphate, electrolytic etching in caustic soda, acid ferric chloride, organic acids with oxidising reagents, perchloric acid, iodine reagents, ammonium molybdate (with and without nitric acid), acetic acid with nitric acid and sulphide solutions. Heat tinting on unetched and etched sections was tried without success.

The work subsequently carried out has been centred around two new methods developed in the Central Research Department for detecting the presence and distribution of lead.

Method A.—Electrolytic Etching Method.

(i) *Macro-Examination.*—The examination may be carried out on a sawn section, free from grease and unpolished. The method consists of electrolytically etching the section in a solution of ammonium acetate prepared either from the salt or from acetic acid and ammonium hydroxide in the following proportions :

(*a*)	Ammonium acetate (cryst.) . .	50 g.
	Water to make	1000 c.c.
(*b*)	Acetic acid (glacial)	75 c.c.
	Water	900 c.c.
	Ammonium hydroxide (0·880 sp. gr.) .	Added until solution just alkaline.

Two alternative procedures may be employed :

(1) The solution is placed in a stainless-steel bath, attached to the positive terminal of a 6-v. accumulator. The billet section is placed on the bottom of the bath, surface upwards. A platinum wire loop extending across the surface of the section is attached to the negative terminal of the accumulator. The wire is held about ¼ in. clear of the section and moved backwards and forwards for a half to one minute.

(2) The metal bath is attached to the negative terminal. The billet section is attached to the positive terminal and held within the solution, with the test surface downwards, and about ¼ in. clear of the base plate. The section is immersed for approximately half a minute.

Photographs of the surface of a cross-section of a selected billet from the discarded bottom end of a leaded steel after etching in ammonium acetate solution and after an exudation test, are shown respectively in Figs. 20 and 21. The position of the lead segregates is marked by the formation of a sharp brown stain over the section through the lead streaks. Heavy rubbing of the surface during drying will result in the partial removal of the stain. The surface should be dried freely after rinsing with water and alcohol.

The advantages of the method are :

(1) It is speedy, taking a total of only two minutes to carry out.

(2) The segregates are distinct and sharply defined on a clean surface and show more clearly than exuded lead beads on a scaled surface.

(3) It eliminates the use of a furnace.

(4) The chemicals required for the solution are inexpensive and readily available.

Examination of etched billet sections and of the same billets after exudation tests shows the etching method to be more satisfactory from the quantitative point of view so far as massive segregates are concerned. It is probable that in the exudation tests, lead segregates that are not actually present at the surface of the billet section under examination force their way to the surface under pressure at the temperature of the test. If a billet section contained only a few segregates, this effect became more evident. The solution may be used repeatedly, but it develops a brown coloration owing to the precipitation of ferric hydroxide. This precipitate can, however, be filtered off, leaving a clear solution which may be used again.

(ii) *Micro-Examination.*—An electrolytic micro-etch obtained in 10% ammonium acetate solution in 30 sec. produced brown stains associated with lead-bearing particles. The appearance of the stains in a sample free from lead segregation is shown in Fig. 22. It was noted that certain particles forming the nuclei of stains were not visible before etching (Figs. 23 and 24).

These stains are composed of hydrated iron oxide films similar to those produced by localised corrosion in heterogeneous material. Differences of potential between the lead-bearing particles or inclusions and the surrounding metal caused a corrosive attack, which was accelerated by the electric current provided during etching, with the local formation of corrosion products and interference fringes. The size and density of the interference fringe varied with the time of etching, and some variation may also be found according to whether the lead particle or inclusion is surrounded by primary ferrite or is within or adjacent to a pearlitic area. Since the corrosive film is dissolved by an acid solution, it is necessary to maintain the solution alkaline.

An attempt has been made to produce similar staining on non-leaded steels. Many samples were examined, but the results were negative with one exception, namely, that of rolled high-sulphur rapid-machining quality steel. In this case, rust rings were developed around some of the sulphide inclusions. These may be associated with the presence of a crevice between the brittle sulphide inclusions and the surrounding ferrite. The test is therefore not absolute evidence of the presence of lead.

Method B.—*Electrographic Method.*

A method of lead-printing was devised by the authors in order to reveal the distribution of lead in unsegregated areas. This method was developed before details were received of the excellent method of lead-printing developed by Mr. Wragge, from which, however, it differs. The procedure, which adopts the principle of electrographic printing (well known in the examination of ores), is carried out as follows:

(1) A piece of gelatine paper is soaked in 10% ammonium acetate solution.

(2) After removal of excess solution, the paper is placed, gelatine face upwards, on a double layer of blotting-paper (soaked in the same solution) resting on a flat aluminium plate connected to the negative terminal of a battery.

(3) The micro-polished surface of the specimen is firmly pressed into contact with the gelatine paper. (For good prints but with less detail, a fine emery polish is sufficient.)

(4) A small brass plate connected to the positive terminal of the battery is brought into contact with the back surface of the specimen. A small electric current is applied across the specimen using about 2 V. per sq. in. for about 2 min.

(5) The current supply is removed, and the gelatine paper containing the undeveloped lead pattern is transferred to a solution of 5% ammonium acetate for about half a minute, washed, then soaked in a weak solution of hydrogen sulphide until a satisfactory definition is obtained.

Thus an image of the position of the lead particles or inclusions in the polished surface is obtained.

A small proportion of saturated tartaric acid solution must be added to the developing solution in order to ensure continued solubility of the iron. It is important that the hydrogen sulphide solution be kept at such a distance that it has no influence on the ammonium acetate solutions.

Many different chemicals, including nitric acid, acetic acid and caustic soda solutions, which would attack lead or form soluble lead compounds were tried, but it was found that a solution of 10% ammonium acetate (cryst.) was the most suitable for soaking the gelatine paper. Similarly, many chemicals forming insoluble coloured compounds with lead, including sodium sulphide, potassium iodide and chromate, and dithizone, were tried as developing reagents, but it was found that a weak solution of hydrogen sulphide gas was the most suitable. This gave a well-defined dark-brown image on an almost colourless background. Other chemicals gave well-defined images, but the colours were of low contrast.

The electrographic method was carried out on many non-leaded steels, but the results were always negative.

The method has so far been applied satisfactorily only to specimens up to 2 in. square.

Application of the Electrographic Method to the Examination of Ingot Sections.

Examination of sections from the top, middle and bottom position of each ingot by lead-printing, electrolytic etching and exudation tests on ground surfaces, indicated freedom from discrete lead or lead-bearing particles.

The electrographic method was successful in showing the size and distribution of the lead particles throughout the ingots. Prints were obtained on sections about 1 in. square taken at positions (*a*) up to 1 in., (*b*) 1 to 2 in., (*c*) 4 to 5 in., (*d*) 7 to 8 in., and (*e*) 10 in. from the surface to the centre of the ingot.

Photographs of a series of direct prints taken on specimens from the outside to the centre of ingot Cast 31/723, at the bottom position, are shown in Fig. 25. A further series of photographs, obtained from the prints by means of transmitted light at a magnification of 7 times, are shown in Fig. 26. These prints are generally similar to those obtained on corresponding samples from the top and middle positions of this ingot and from the three positions in ingot Cast 35/7295.

The chief features of the lead distribution were found to be :

(*a*) A thin zone, at the outer surface, apparently containing only a small amount of lead. This zone was about 0·1 in. thick at the top, increasing to 0·2 in. at the bottom (Fig. 26*a*).

(*b*) An apparent abrupt increase from the thin surface zone to a dense random distribution of small particles (Fig. 26*b*).

(*c*) Marked increase in the size of the particles at a position of 1–2 in. from the surface (Fig. 26*c*).

(*d*) Further marked increase in size of the particles, which showed a slight network pattern (Fig. 26*d*); the distribution remained generally similar up to the centre of the ingot (Figs. 26*e* and *f*).

Microscopical Examination of Inclusions.

Microscopical examination of the types of non-metallic inclusions present was carried out on the sections taken at the three positions in each ingot. The sections examined were taken from the same positions with respect to the outer surface of the ingot as in the case of the electrographic prints. The distribution of inclusions was, however, similar at the top, middle and bottom positions of both ingots, and a general description follows, in which any reference to oxides is intended to be interpreted as mixed oxides of iron and lead in varying proportions or lead oxide.

Towards the outer surface of the ingot, the inclusions were mainly small sulphides, randomly distributed, with occasional large silicate globules. The sulphides showed a tendency to be easily pitted when polished, whilst some of the particles were of darker grey than normal manganese sulphide and contained a second phase (Fig. 27). The appearance of a typical silicate globule containing small sulphide particles at the periphery is shown in Fig. 28.

After electrolytic etching in ammonium acetate solution, stains developed round the darker-grey sulphide inclusions (Fig. 29), whilst many stains developed from nuclei which were not visible before etching. It is suggested that the latter are particles of metallic lead. The lighter-grey sulphides were only very slightly darkened. Silicate globules were generally unstained, except at the position of some of the small sulphide particles at the periphery (Fig. 30). Light re-polishing removed the stains and revealed clearly the sulphides which were attacked (Fig. 31) and the particles, probably lead, not visible before etching (as illustrated in Figs. 23 and 24). After light re-polishing, it is not possible definitely to identify the particles believed to be lead, since they are severely attacked by the ammonium acetate.

At a position beyond 1 in. from the outer surface of the ingot, sulphides and silicates increased in size and number, and tended to be more duplex. Fig. 32 shows a typical duplex inclusion of dark-grey sulphide and a dark angular crystal of oxide. After electrolytic etching and light re-polishing, the entire inclusion was attacked (Fig. 33). The attack on the oxide crystal did not appear to have been as severe as would occur in the case of lead particles. Lead-bearing particles were also evident at the periphery of the sulphide and silicate globules (Fig. 34), but after etching and re-polishing, oxide crystals within the globules were more clearly defined (Fig. 35).

The complex inclusions present included silicate globules containing particles of lead and oxide (Fig. 36), and oxide globules containing particles of sulphide, lead and silica (Fig. 37). After electrolytic etching and re-polishing, the inclusion shown in Fig. 36 showed severe attack on the lead constituent and no attack on the silicate constituent, whilst the inclusion shown in Fig. 37 showed severe attack on the lead and sulphide particles, slight attack on the oxide and removal of the silica globules (Fig. 38).

At the centre of the ingot, the inclusions were similar to those beyond 1 in. from the surface, with the presence of large silicate globules containing particles of lead and oxide, and often rendered duplex by sulphide particles (Figs. 39 and 40).

The ingot from Cast 35/7295 showed more numerous sulphide inclusions, enveloped by a darker constituent (Figs. 41 and 42). At the centre

PLATE XXXIII.

FIG. 1.—Standard High-Sulphur Steel, monophase sulphides. × 800.

FIG. 2.—Standard High-Sulphur Steel, duplex oxide-sulphide. × 800.

FIG. 3.—Leaded High-Sulphur Steel, monophase sulphide. × 800.

FIG. 4.—Leaded High-Sulphur Steel, duplex oxide-sulphide. × 800.

FIG. 5.—Leaded High-Sulphur Steel, complex inclusion streaks. × 300.

(Illustrations reduced to four-fifths linear in reproduction.)

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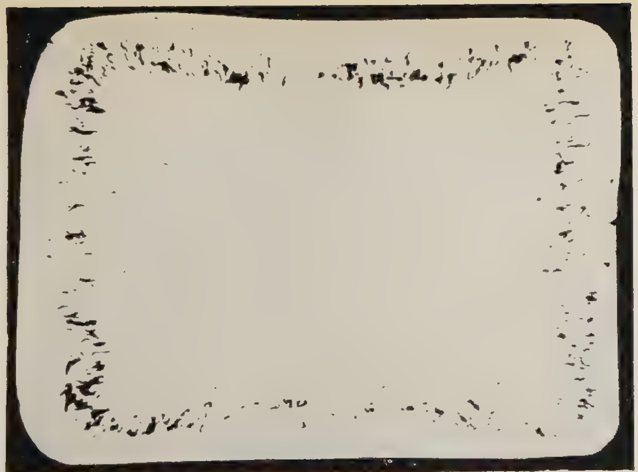

FIG. 6.—Lead-Print of Bottom-Discarded Slab of Manganese-Molybdenum Steel. × 1.

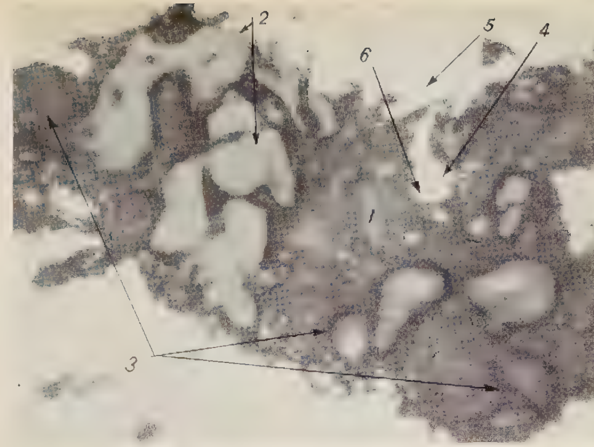

FIG. 8.—Phases Found in Massive Lead Streaks. × 800.

FIG. 7.—Massive Lead Streak. × 300.

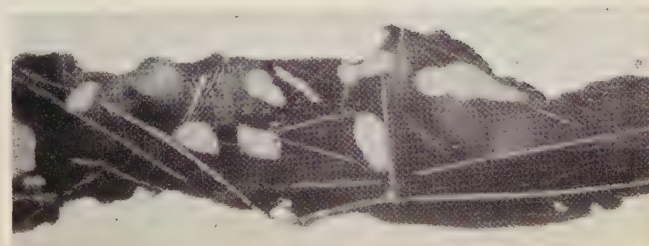

FIG. 9.—Non-Metallic Segregate. × 800.

(Fig. 6 reduced to three-quarters, Figs. 7 to 9 to four-fifths linear in reproduction.)

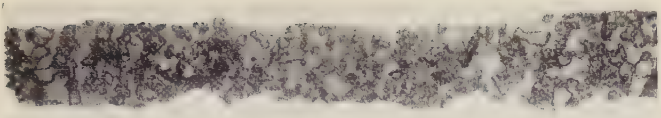

FIG. 10.—Non-Metallic Segregate. × 300.

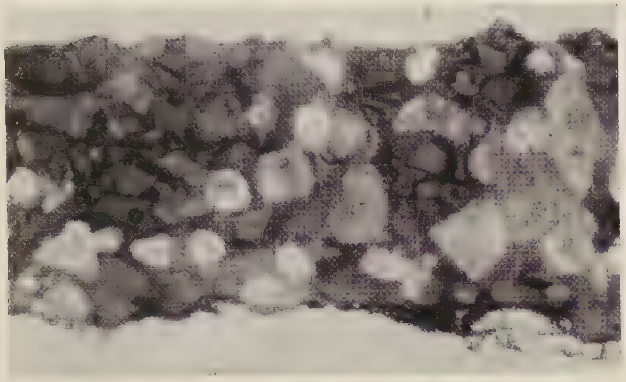

FIG. 11.—Non-Metallic Segregate Etched with 0·5% HF. × 800.

FIG. 12.—Non-Leaded Manganese–Molybdenum Steel. × 300.

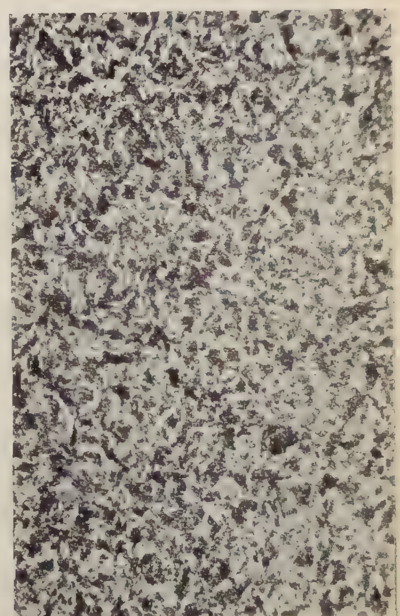

FIG. 13.—Leaded Manganese–Molybdenum Steel. × 300.

(Illustrations reduced to four-fifths linear in reproduction.)

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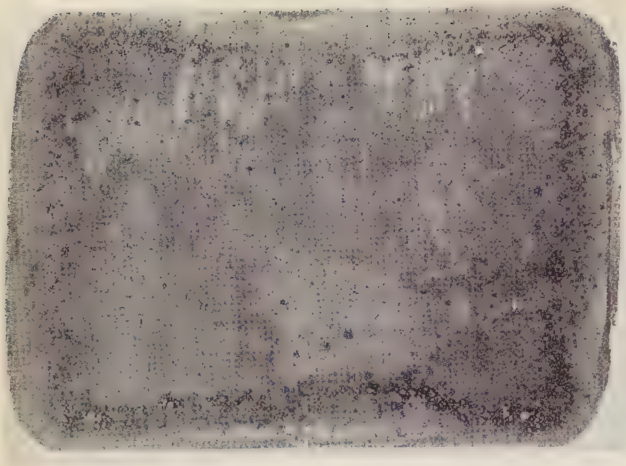

FIG. 14.—Bottom-Discarded Billet after Lead Exudation Test. x 1.

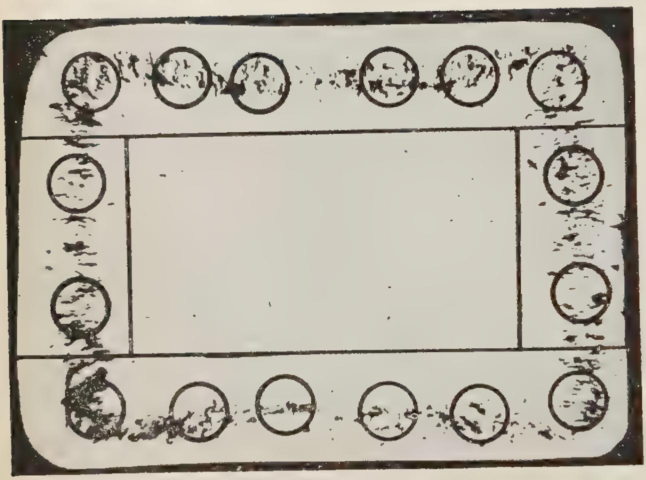

FIG. 15.—Position of Samples for High-Temperature Microscopic Examination. x 1.

FIG. 16.—Appearance of Beads on the Surface of a Sample after Heating. Oblique illumination. × 10.

(Figs. 14 and 15 reduced to three-quarters, Fig. 16 to four-fifths linear in reproduction.)

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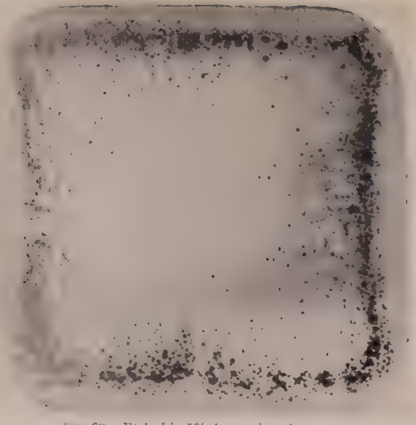


Fig. 20.—Etched in 5% Ammonium Acetate.

Fig. 21.—After Exudation Test.

Figs. 20 and 21.—Lead Segregation Tests of Leaded-Steel Billet. (Bottom discard.)

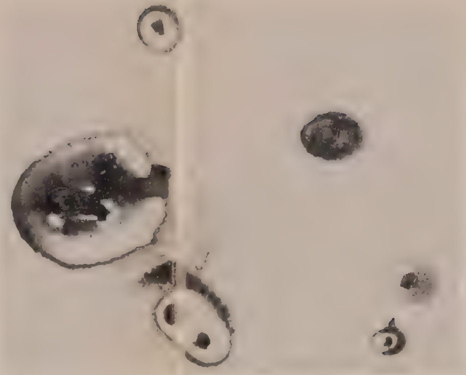


Fig. 22.—Stains and Films produced by 10% ammonium acetate. × 800.

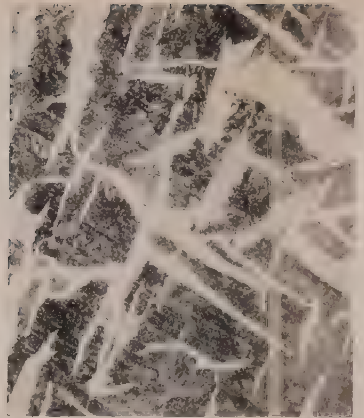
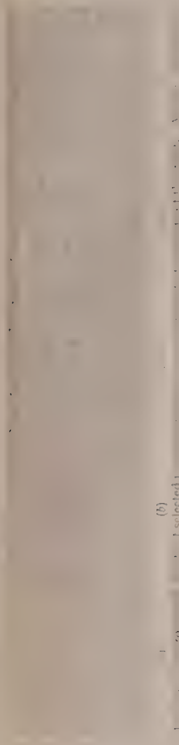


Fig. 23.—Before Electrolytic Etching and etched in 2% nitric acid. × 300.

Fig. 24



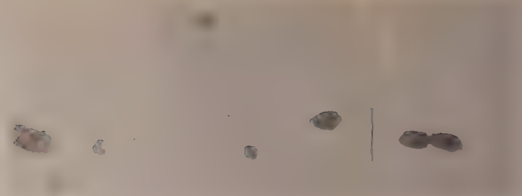
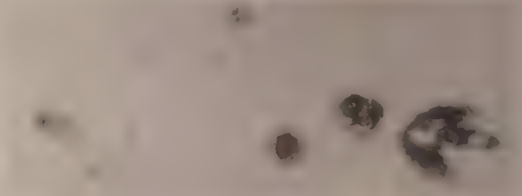


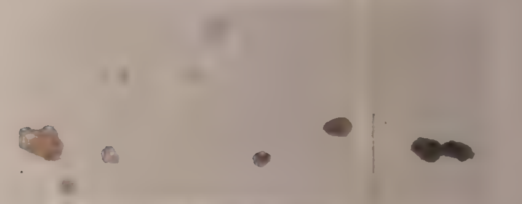
Fig. 27.—Sulphides towards Outer Surface of the Ingots. × 800.

Fig. 28.—Silicate Globules with Small Sulphide Particles. × 1300.



29.—As Fig. 27, after etching with ammonium acetate. × 800.

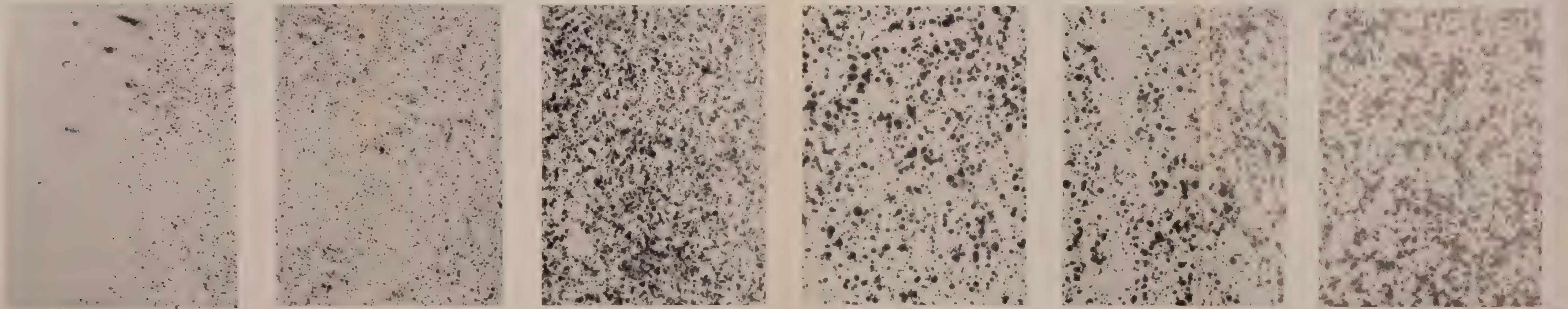
Fig. 30.—As Fig. 28, after etching with ammonium acetate. × 1300.



polishing. × 800.

Fig. 32.—Sulphide Globules with Oxide Crystal. × 1500.

Fig. 33.—As Fig. 32, after etching and polishing. × 1500.



Outside Surface. (a) (b) (c) (d) (e) (f) Centre

Fig. 26.—Distribution of Lead and Lead-Bearing Particles from Outside to Centre of the Ingot, as Revealed by Electrographic Printing. × 7.

(Figs. 17 to 24 reduced to four-fifths, Fig. 26 to three-quarters linear in reproduction.)

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Fig. 34.—Sulphide Globules with Oxide Crystals and Envelopes of Lead-Bearing Particles.

Fig. 35.—As Fig. 34, after etching and repolishing. × 800.

Fig. 36.—Silicate with Lead and Oxide. × 800.

Fig. 37.—Oxide with Lead, sulphide and silica particles. × 800.

Fig. 38.—As Fig. 37 after etching and repolishing. × 800.

Fig. 39.—Silicate with Lead and Oxide. × 800.

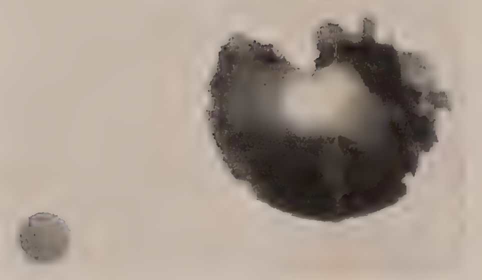

Fig. 40.—Silicate with Lead, Oxide and Sulphide. × 800.

Fig. 41.—Enveloped Sulphide and Lead Particles. × 800.

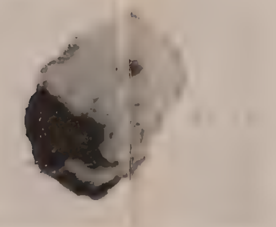

Fig. 42.—Enveloped Sulphide with mass of Lead, etched and repolished. × 800.

Fig. 43.—Lead Segregate at the Centre of Ingot 35/7295. × 800.

Fig. 46.—... Ingot.

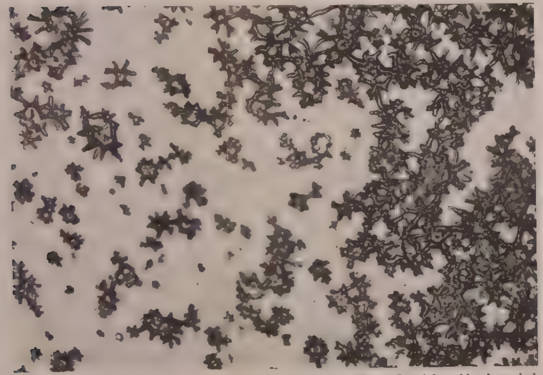

Fig. 44.—Surface of Section after Heating to approx. 400° C. Particles of lead exuded on to the surface. The dense mass at the right was part of an interdendritic formation. × 16.

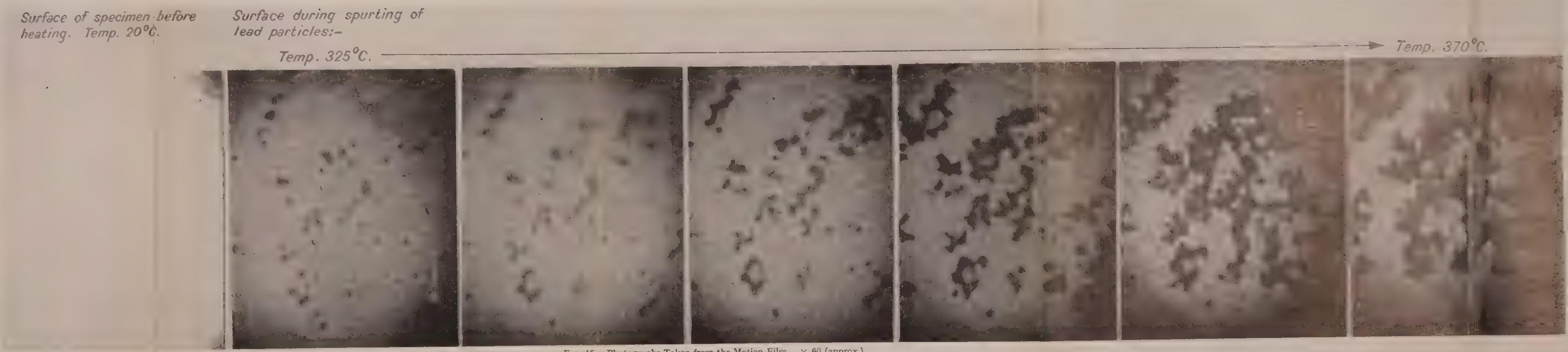

Fig. 45.—Photographs Taken from the Motion Film. × 60 (approx.).

(Figs. 46 and 47 reduced to four-fifths linear in reproduction.)

Fig. 47.—Cast 31/723. Lead Segregation at Base of Ingot. × 1·5.

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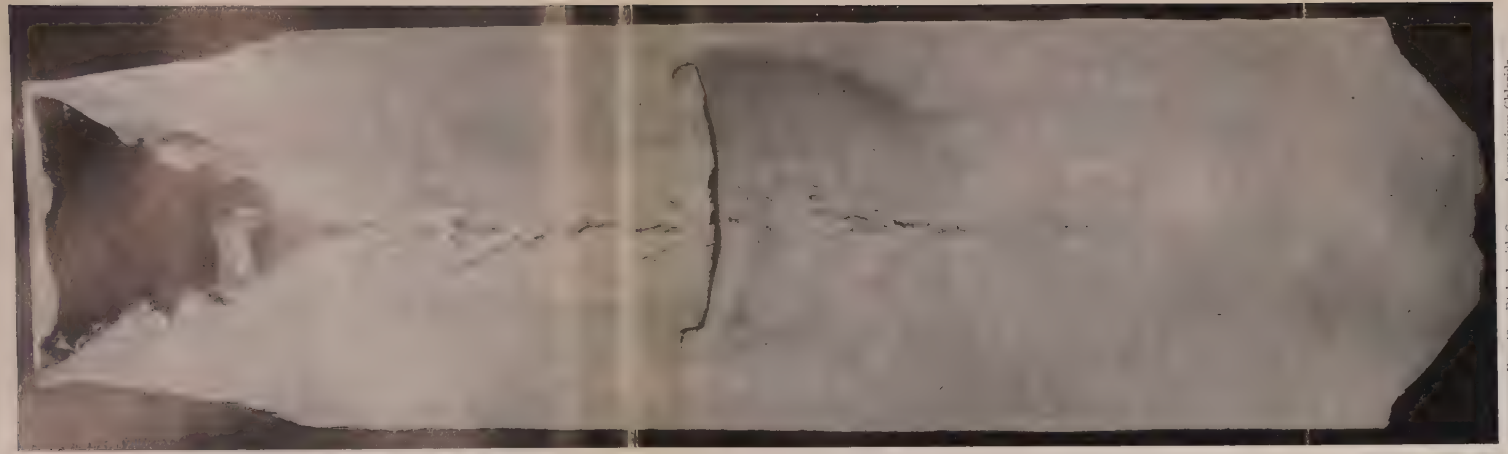


FIG. 49.—Etched with Copper Ammonium Chloride.

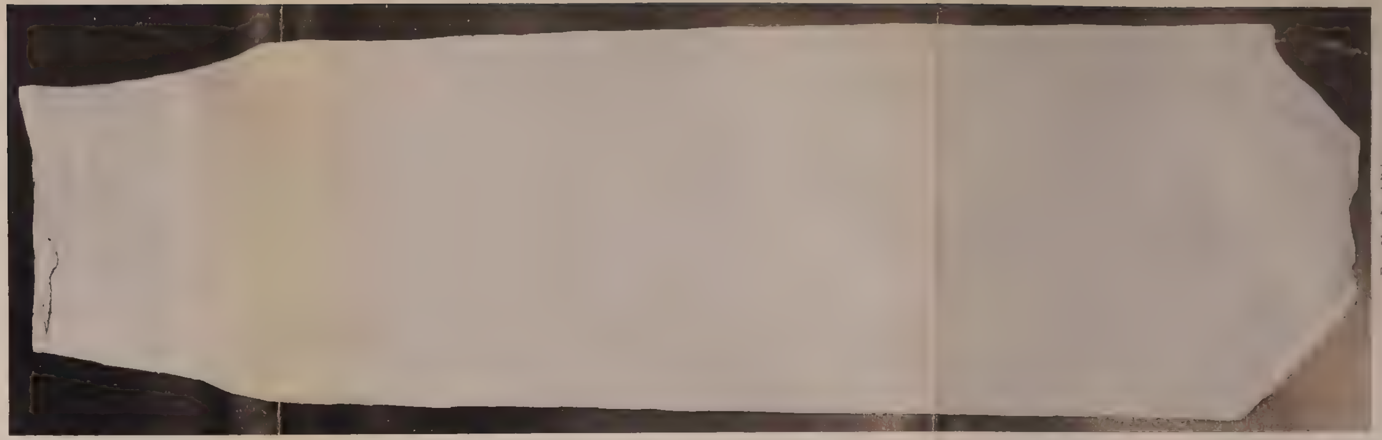
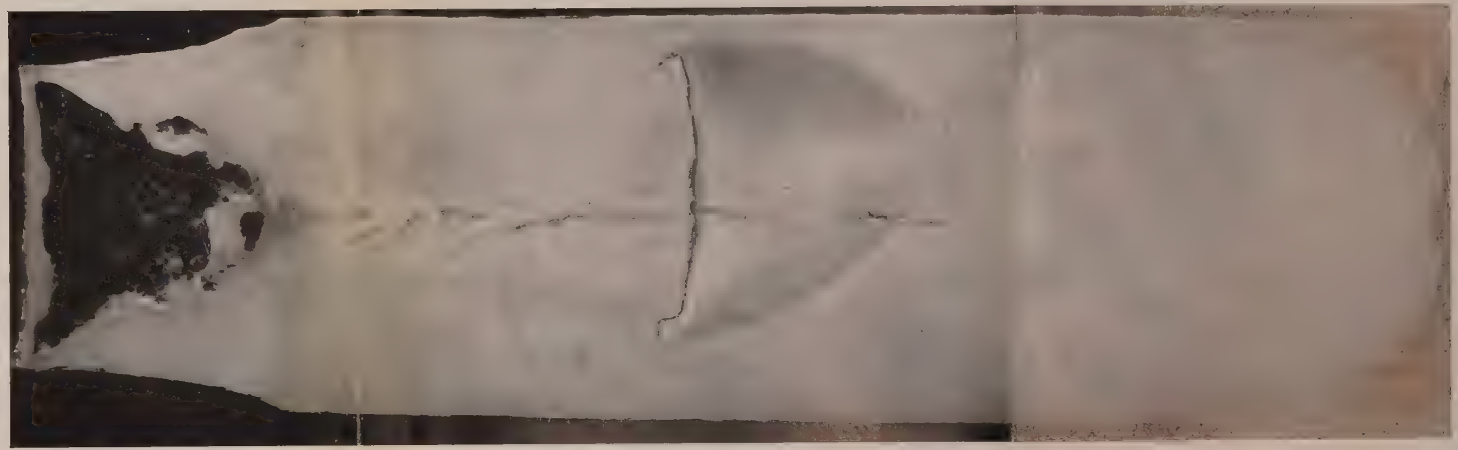


FIG. 51.—Lead Print.

FIG. 48.—Sulphur Print.



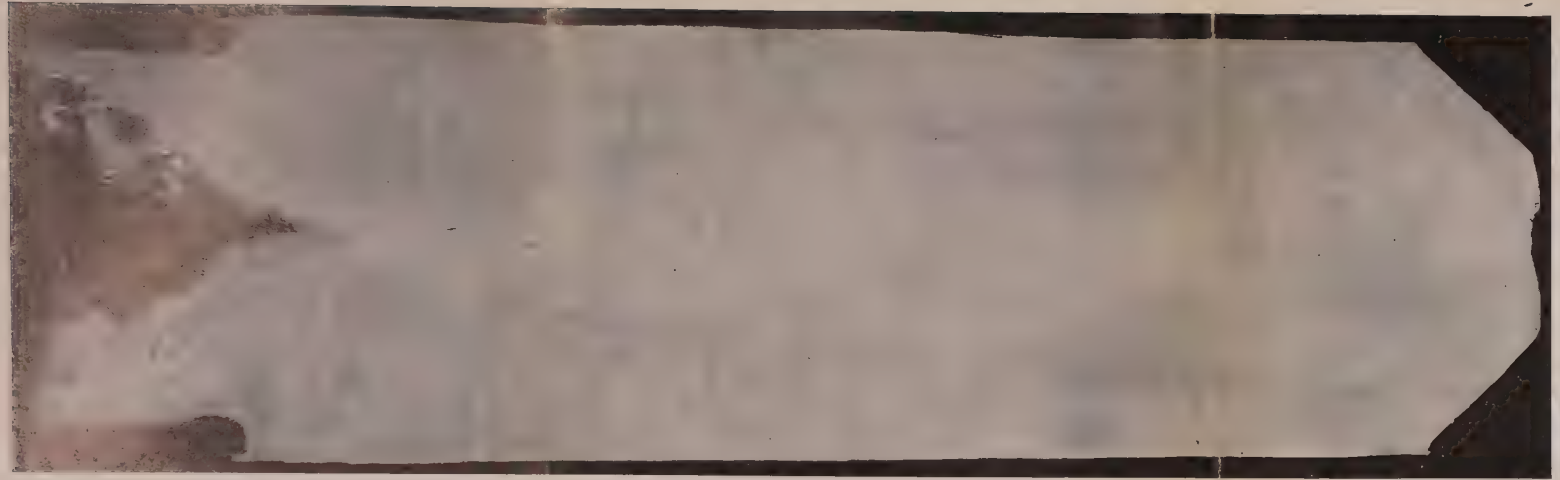


FIG. 53.—Etched with Copper Ammonium Chloride.

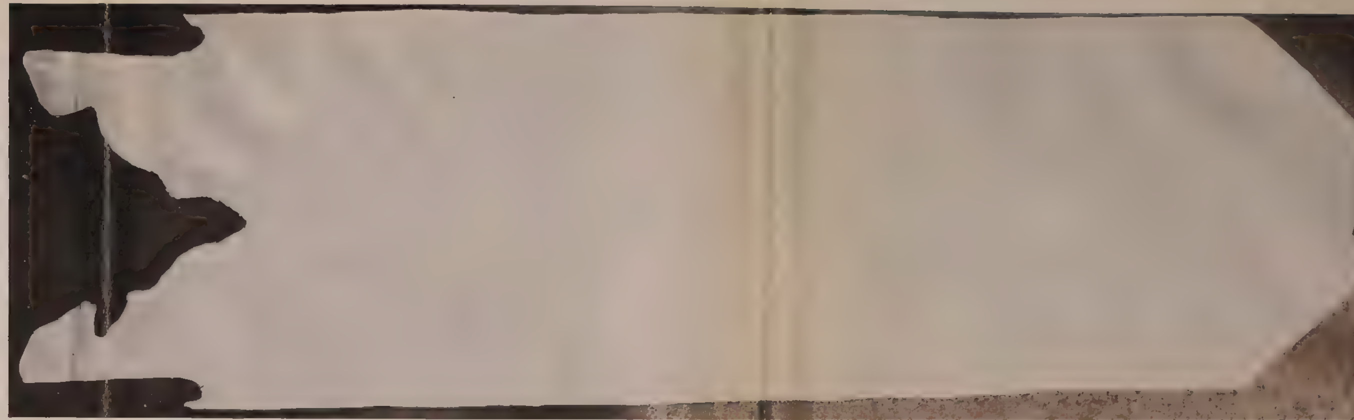


FIG. 55.—Lead Print.

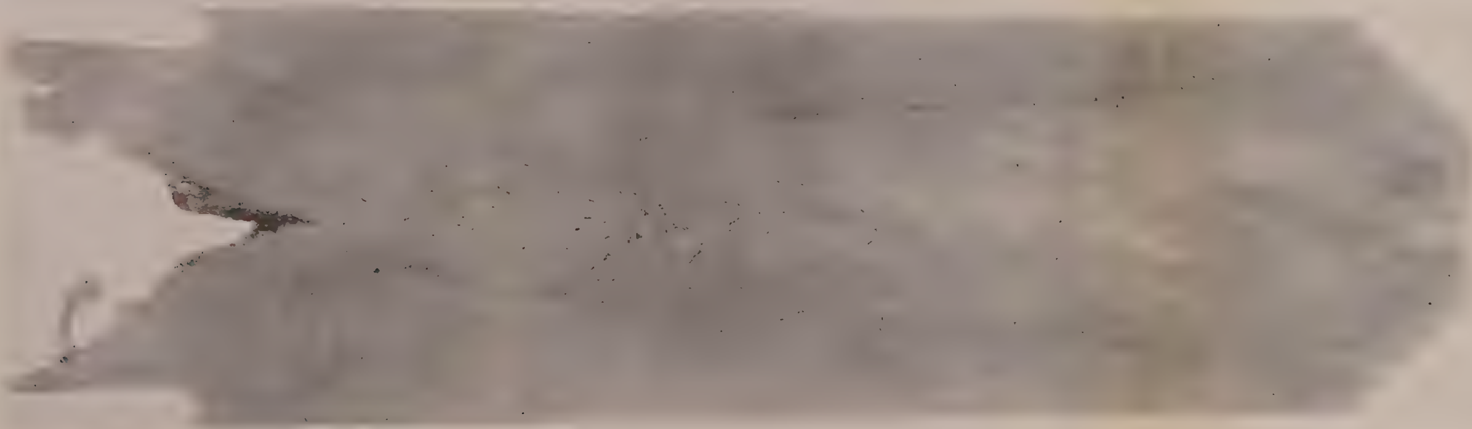


FIG. 52.—Sulphur Print.

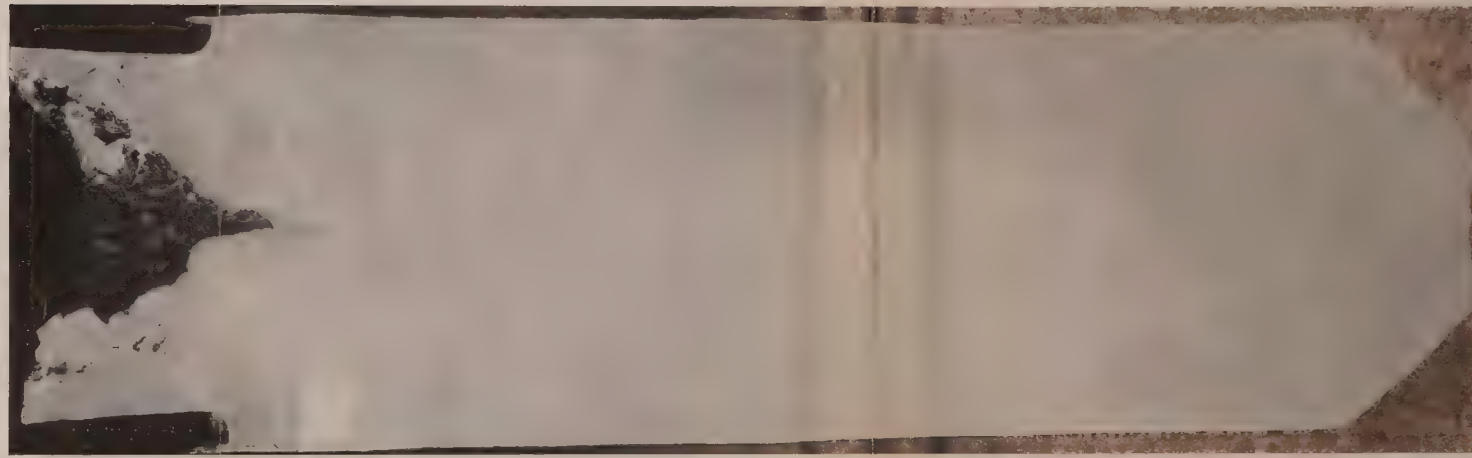


FIG. 54.—Etched with Nitric Acid.

FIGS. 52 to 55.—Cast 31 723.

of this ingot, there was slight segregation, the segregates being mainly composed of lead and oxide (Fig. 43).

Micro-Exudation during Heating.

In order to determine whether the small particles, believed to be lead, in sections from the ingot would exude from the surface when heated, an examination was carried out at a considerably higher magnification than that used in the previous tests, the microscope being fitted with a $\frac{2}{3}$-in. objective. A motion film was taken to show the behaviour of the lead. The apparatus was assembled as in the previous examination.

A number of tests were made, and in each test it was observed that very fine particles of lead began to spurt out on to the polished surface at a temperature of 325° to 330° C. Spurting continued up to a temperature of approximately 370° C. The appearance of the polished surface after heating is shown in Fig. 44. The distribution of the exuded particles had to a slight extent the appearance of an interdendritic network, especially in the early stages of exudation, whilst lead particles originally at the polished surface were unaffected by heating. Photographs from the motion film obtained during the heating of the specimen are shown in Fig. 45.

It will be noted that the temperature of the commencement of exudation coincides with the melting point of lead, *i.e.*, 327° C. The temperature of spurting in billets in which lead had segregated was found to be much lower, *viz.*, 235–240° C. This difference may be associated with differences in internal stresses in the billet and ingot.

Sections taken at right angles to the exuded surface failed to reveal cavities associated with exuded lead.

In view of the observed exudation of lead from unsegregated ingot sections, further tests were carried out in order to determine the behaviour of lead in an unsegregated billet under the same conditions. Lead was found to exude from the billet in a manner generally similar to that of the ingot, but the exudation commenced at a slightly lower temperature of 300–310° C. and the distribution was more dense and uniform.

Summary.

The following observations briefly summarise the results of examinations of wrought leaded and non-leaded manganese–molybdenum and high-sulphur steels and of lead-bearing ingots of 0·25% carbon, 1·0% manganese steels.

(1) The bottom-discarded billet of manganese–molybdenum steel showed marked segregation, which occurred mainly towards the surface. The segregation consisted chiefly of streaks of metallic lead containing particles of sulphide, oxide and steel. Other, less-numerous, segregated lead-bearing streaks consisted of particles of oxide and sulphide within a matrix of oxide or silicate.

(2) Lead was found associated with sulphide and silicate inclusions as lead and in the form of oxide.

(3) Normal methods of microscopical examination failed to reveal the presence of lead unassociated with inclusions.

(4) Electrolytic etching in 10% ammonium acetate solution revealed particles, believed to be lead, within the structure. These particles were not visible before etching.

(5) An electrographic method of lead-printing clearly revealed the distribution of lead in the ingot. There was an increase in the size of the particles from the surface to a position about 4 in. from the surface, with a gradual change from a random to a slightly interdendritic pattern.

From this position to the centre the distribution was generally similar. There was a marked similarity between the lead distribution in the two ingots and at the top, middle and bottom positions.

(6) Examination of lead-bearing specimens under the microscope during heating showed that segregated lead in a billet suddenly spurted out on to the surface at a temperature of 235–240° C., the steel structure being slightly distorted as a result of the exudation. An unsegregated billet showed lead exudation at a higher temperature of 300–310° C., the distribution of the lead being fairly dense and uniform. In the case of the ingot, the lead spurted out at the still higher temperature of 325–330° C., corresponding to the melting point of lead. The distribution of the lead on exuding showed a partial interdendritic formation. The size of the exuded lead particles was markedly less in the unsegregated than in the segregated specimens.

It is to be noted that bottom-end discards were selected for the examination of massive segregates, the remaining samples being free from such segregation.

Appendix.—*An Examination of Two Leaded Ingots of 0·25% Carbon, 1% Manganese Steel.*

Certain work in the preceding paper has been carried out on sections cut from two leaded ingots of 0·25% carbon, 1% manganese steel, one cast wide end up and one cast narrow end up, both from 80-ton basic open-hearth casts.

Lead additions, equal to 0·22% of the liquid steel, were made in the form of lead shot to each trumpet of the uphill cast ingots. The calculated lead addition was fed down the trumpet from a hopper at the same time as the liquid metal from the ladle. The rate of feed was adjusted to give even pouring throughout the entire ingot-casting period.

The general features of these ingots are described below, with particular reference to the distribution of lead.

Narrow-End-Up Ingot.

This was run uphill from Cast No. 35/7295 and weighed 75 cwt. The pit sample analysis is given below :

Carbon	0·25%	Sulphur	0·033%
Manganese	0·88%	Phosphorus	0·032%
Silicon	$0{\cdot}17_5$%		

The ingot was sectioned longitudinally on the central plane, and photographs of the half ingot after (1) sulphur-printing, (2) macro-etching with copper ammonium chloride, (3) macro-etching with dilute nitric acid, and (4) lead-printing by the ammonium acetate–hydrogen sulphide method are given in Figs. 48 to 51. Unfortunately, during parting, the face of the section was damaged owing to the saw cutting off-centre, and the effect of this is seen in the various photographs.

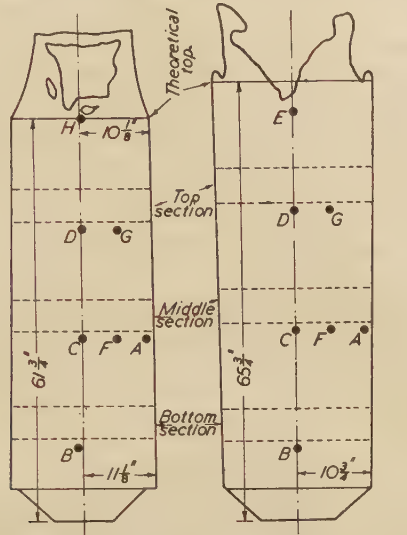
Wide-End-Up Ingot.

This was also run uphill, but from Cast No. 31/723 and weighed 84 cwt. The pit sample analysis is as follows :

Carbon	0·27%	Sulphur	0·032%
Manganese	1·02%	Phosphorus	0·027%
Silicon	$0{\cdot}14_5$%		

The ingot was sectioned in a similar manner to the narrow-end-up ingot and examined by the same four methods (Figs. 52–55).

Both ingots were subjected to chemical analysis, including lead determinations, at the eight standard positions of the Heterogeneity of Steel Ingots Committee. The precise location of the positions in the two ingots is shown in Figs. 56*a* and *b* and the analyses are given in Table I.



(*a*) Cast 35/7295. (*b*) Cast 31/723.

FIG. 56.—Positions of Samples.

TABLE I.—*Composition of Ingot Sections at the Standard Heterogeneity of Steel Ingots Committee Positions.*

Cast Numbers 35/7295 and 31/723.

Position of Sample.	Description.	Carbon, %.	Manganese, %.	Silicon, %.	Sulphur, %.	Phosphorus, %.	Lead, %.
	Cast 35/7295.						
A	Outer chill portion.	0·25	0·82	0·14	0·033	0·025	0·18
B	Bottom of ingot.	0·22	0·81	0·15	0·030	0·023	0·14
C	On centre axis; one-third way between *B* and *H*.	0·23	0·81	0·15	0·028	0·022	0·14$_5$
D	As *C*; two-thirds way between *B* and *H*.	0·26	0·83	0·15	0·033	0·027	0·15
E	As *C*; base of piped zone.	...	...	...	...	...	...
F	Midway between *A* and *C*.	0·25	0·82$_5$	0·14	0·031	0·026	0·15
G	Near *D*, above *F*.	0·27$_5$	0·83$_5$	0·14$_5$	0·033	0·026	0·16
H	Centre of theoretical top.	0·41	0·92	0·17	0·075	0·046	0·15
	Cast 31/723.						
A	Outer chill portion.	0·25	1·01$_5$	0·15	0·033	0·024	0·17$_5$
B	Bottom of ingot.	0·23$_5$	0·99	0·14$_5$	0·030	0·022	0·14$_5$
C	On centre axis; one-third way between *B* and *H*.	0·25	0·99	0·14$_5$	0·029	0·024	0·15$_5$
D	As *C*; two-thirds way between *B* and *H*.	0·27$_5$	1·02	0·14$_5$	0·032$_5$	0·023$_5$	0·16$_5$
E	As *C*; base of piped zone.	0·46	1·15	0·21$_5$	0·101	0·050	0·13
F	Midway between *A* and *C*.	0·27	1·00	0·15	0·034	0·026	0·15$_5$
G	Near *D*, above *F*.	0·28	1·02	0·15	0·035	0·026	0·15
H	Centre of theoretical top.	...	...	...	...	...	...

In addition, 5-in.-wide slices were cut transversely from each ingot to represent the top, middle and bottom positions at approximately 75%, 50% and 25% respectively of the distance from the bottom of the ingot. The location of these slices is also indicated in Figs. 56*a* and *b*. Each slice was further cut up into three transverse sections, each of 1-in. width after discarding the 1-in. outer pieces. The 1-in. sections were used for microscopical examination, lead-printing and exudation tests, as described in the foregoing paper, while chemical analyses for lead were also carried out at half-inch intervals from the outside to the centre. The figures obtained represent the average lead contents of successive half-inches of the ingot at the top, middle and bottom positions and are given in Table II.

TABLE II.—*Variation in Lead Content from Outside to Centre of Ingot.*

Cast Numbers 35/7295 and 31/723.

Sample Number.*	Lead, %.					
	Cast 35/7295.			Cast 31/723.		
	Top.	Middle.	Bottom.	Top.	Middle.	Bottom.
Outside 1	$0{\cdot}18_5$	0·17	$0{\cdot}16_5$	$0{\cdot}18_5$	$0{\cdot}17_5$	0·15
2	$0{\cdot}18_5$	$0{\cdot}17_5$	0·17	$0{\cdot}18_5$	$0{\cdot}19_5$	$0{\cdot}16_5$
3	0·18	0·18	0·17	$0{\cdot}18_5$	0·19	$0{\cdot}17_5$
4	0·18	0·18	$0{\cdot}17_5$	$0{\cdot}17_5$	$0{\cdot}19_5$	$0{\cdot}17_5$
5	$0{\cdot}17_5$	$0{\cdot}17_5$	0·18	0·18	0·18	0·19
6	0·17	0·17	0·17	$0{\cdot}18_5$	$0{\cdot}18_5$	$0{\cdot}18_5$
7	$0{\cdot}17_5$	$0{\cdot}16_5$	$0{\cdot}16_5$	0·18	0·17	$0{\cdot}17_5$
8	0·17	0·17	$0{\cdot}17_5$	$0{\cdot}18_5$	$0{\cdot}17_5$	0·18
9	0·18	0·17	$0{\cdot}16_5$	$0{\cdot}17_5$	$0{\cdot}16_5$	$0{\cdot}17_5$
10	$0{\cdot}17_5$	0·18	$0{\cdot}16_5$	$0{\cdot}16_5$	$0{\cdot}17_5$	0·16
11	0·17	$0{\cdot}17_5$	0·15	0·18	0·18	$0{\cdot}14_5$
12	$0{\cdot}16_5$	$0{\cdot}16_5$	0·15	$0{\cdot}16_5$	$0{\cdot}16_5$	0·15
13	0·17	0·16	0·15	0·17	$0{\cdot}17_5$	0·15
14	$0{\cdot}16_5$	$0{\cdot}16_5$	0·15	$0{\cdot}18_5$	0·17	0·16
15	0·16	$0{\cdot}15_5$	0·16	$0{\cdot}18_5$	0·15	0·16
16	0·15	$0{\cdot}15_5$	0·15	$0{\cdot}16_5$	$0{\cdot}15_5$	$0{\cdot}15_5$
17	$0{\cdot}15_5$	0·16	$0{\cdot}15_5$	$0{\cdot}18_5$	0·15	0·15
18	$0{\cdot}15_5$	0·15	0·14	$0{\cdot}18_5$	$0{\cdot}14_5$	0·15
19	0·16	0·15	$0{\cdot}14_5$	0·18	$0{\cdot}14_5$	$0{\cdot}15_5$
20	0·16	0·15	0·15	$0{\cdot}19_5$	$0{\cdot}14_5$	0·15
21	...	$0{\cdot}15_5$	0·15	$0{\cdot}19_5$	$0{\cdot}14_5$	0·15
Centre 22	...	...	...	0·18	$0{\cdot}14_5$	...
Maximum variation	$0{\cdot}03_5$	0·03	0·04	0·03	0·05	$0{\cdot}04_5$

* Each sample represents an area of 0·5 in. dia. starting from the outside edge.

Comments on the Data Presented.

The sulphur prints and macro-etchings of the two sections reveal the normal features of ingot segregation. Central V segregates and Λ segregates are shown, both types being less marked in section 31/723. The copper ammonium chloride etching reveals lead segregation just within the base of the two ingot sections. The continuous lines parallel to the surface of the ingot base indicate interference in the crystal structure at the position of the lead segregation. This interference is very slight, but is more marked in section 31/723, in which the banded appearance extends up the outer edges of the ingot for about 16 in. Figs. 46 and 47, which have a magnification of 1·5, reveal this segregation more clearly.

The lead-prints show the complete absence of massive lead segregation in the general body of the ingot section, with the exception of areas about 1¼–1½ in. in width towards the outer surfaces of the base of the ingot sections. Ingot section 31/723 is particularly uniform and the segregation at the base much less marked.

The chemical segregation, as revealed by the analyses at the standard positions laid down by the Heterogeneity of Steel Ingots Committee, is of the normal type and order. Positions H and E of sections 35/7295 and 31/723 respectively are in the heavily segregated region, while position B is the least segregated and position G the most segregated of the remaining positions. The distribution of lead as revealed by the standard positions and by the variations in the three slices representing top, middle and bottom of the ingots, is particularly uniform, the maximum variations being of the order of 0·04% of lead. The highest values occur in the chill (position A) and the lowest values at B, one-third of the way up the central axis. The variations along the strips tend to confirm the slightly higher result at the edge and the lower value at the centre. The distribution in the leaded rimming-steel ingot examined by Northcott and McLean,[1] with the exception of the peak at the junction of rim and core, is of a similar type, being high at the edge and low at the centre, although the differences obtained in their ingot are larger than in the present ingots. The segregation of lead does not, therefore, follow completely the normal features of segregation in solid ingots.

Acknowledgments.

The authors wish to acknowledge with thanks the contribution of the Appendix by Mr. W. W. Stevenson and Mr. G. E. Speight of the Central Research Department of The United Steel Companies, Ltd., who were responsible for the work described therein. Thanks are also expressed to the Directors of the United Steel Companies, Ltd., for permission to publish this paper.

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(4) A. von Vegesack: *Jernkontorets Annaler*, 1942, vol. 126, No. 12, p. 559.

[This paper was discussed jointly with the two preceding ones by C. S. Graham on "Examination of Two Ingots of Free-Cutting Steel, one Containing Lead and the other Lead-Free" and by T. H. Schofield on "The Microscopical Examination of Samples of Lead-Bearing and Lead-Free Steels and Ingot Irons."]

JOINT DISCUSSION.

The papers by C. S. Graham on "Examination of Two Ingots of Free-Cutting Steel, one Containing Lead and the other Lead-Free," by T. H. Schofield on "The Microscopical Examination of Samples of Lead-Bearing and Lead-Free Steels and Ingot Irons" and by W. E. Bardgett and R. E. Lismer on "Mode of Occurrence of Lead in Lead-Bearing Steels and the Mechanism of the Exudation Test" were discussed jointly.

Dr. J. H. Whiteley (Consett Iron Co., Ltd., Consett, Co. Durham): I think that the papers by Mr. Schofield and by Messrs. Bardgett and Lismer might both have been entitled "Puzzle, Find the Lead," for in what form that presumably insoluble substance exists in steel as a constituent is a problem very difficult to solve. One would expect that a heavy metal like lead would begin to coalesce and gradually sink as the liquid steel cooled, but only in the ingot described by Mr. Graham is any indication to be seen of such an effect. There (*see* Table II.), the average lead content at position B_1 is 0·18%, whereas at position H it is 0·16%, and even then that difference may be due to the way the lead was added. Apparently, therefore, lead particles persist in a fine suspension to a much greater degree than silicate globules, for the latter readily coalesce into larger particles. In the endeavour to find the lead under the microscope Mr. Schofield has, I think, done a very fine bit of metallographical work, and, further, he has been too modest about it, for he told me that in his first attempts the specimens were hand-polished for 8 hr.! Having some experience of the great difficulty in retaining the white constituent intact during preparation, I would congratulate him both on his achievement and on the excellence of the illustrations. While this constituent undoubtedly contains lead, since it occurs only in lead-bearing steels, my experiments show that the whole of the lead is not present in it. For instance, on heating the specimen which Mr. Schofield sent to me to 500° C. in a non-oxidising atmosphere lead globules appeared on the surface while the constituent remained practically unaltered. The same thing happened when I treated in that way another sample of plain low-carbon lead-bearing steel which also had the white constituent. Hence, it seems reasonable to infer that in whatever form the lead exists in the constituent only a part is there present. It would be interesting to ascertain whether the constituent is still formed when pure lead is added to pure iron.

There is one other point that I should mention in connection with Mr. Schofield's specimen. Lead globules which appear on the surface of a heated specimen can easily be flattened into cakes by pressure, and in several samples I had previously observed that if the specimen was then immersed in a 10% solution of chromic acid in water these cakes were stained black. The sample that Mr. Schofield gave me proved an exception, however, for the lead cakes were unattacked by this solution and remained whiter than the rest of the surface. I surmise that this is due either to the presence in or absence from the lead of some substance which it takes up from the steel.

Turning now to the paper by Messrs. Bardgett and Lismer, I note in the first place that no mention is made of the white constituent that Mr. Schofield has described, and would ask the authors whether they found it. I know they were aware of Mr. Schofield's work, and wonder whether it was out of courtesy to him that they did not include this constituent in their illustrations.

I would next remark that the authors have described the various inclusions which they illustrate as lead, oxides, silicates, sulphide and silica, and would like to know how they have identified them. We are told that

certain dark inclusions are lead, but I would point out that flattened lead globules are as white as the constituent discovered by Mr. Schofield. Until we know definitely what appearance the lead has there is little hope of detecting it with certainty under the microscope. Again, in Fig. 9 the authors show a dark-grey matrix, which they state is probably mixed oxides. As far as I have observed, however, oxides are always opaque, which the matrix in Fig. 9 does not appear to be, since diffraction bands can be seen in it. Fig. 37 provides another example of dubious identification, for we are informed that it shows an oxide globule containing lead, sulphide and silica particles. One would expect that silica in the presence of oxide would form a silicate.

With regard to the exudation of lead, I can confirm the authors' observations on p. 284 P that lead beads spurt out at temperatures as low as 240° C., for I obtained a few globules on a lead-bearing specimen in a 10 min. heating at 250° C. On again heating the specimen at 350° C., however, a much larger number appeared, so that it would seem as though only a part of the lead comes to the surface at temperatures below its melting point. That the exudation of lead should cause even a slight distortion of the structure of the steel adjacent to the lead streak, as the authors affirm on p. 284 P, I simply do not believe. I consider that the distortion shown in Figs. 17 and 19 is due to a burring of the edges in preparation. When a cavity is present at the surface much care and patience are needed to obtain sharp edges free from any overlap; a prolonged rubbing on 00 emery paper is required. That brings me to another point in connection with the preparation of the specimens.

On p. 289 P the authors state that lead particles originally at the polished surface were unaffected by heating, and I would ask them how they know that any lead was there. It seems to me that to retain the lead at the surface the specimen should be prepared under either water or paraffin from the coarse-emery stage onwards, so as to avoid the danger of lead removal by the heat of friction. Supposing that the authors were given two unsegregated samples, one of which contained lead, could they identify that one from a microscopic examination alone after the specimens had been prepared in the ordinary way?

Lastly, I would remark on Fig. 25, where the lead appears to increase in quantity from the outside to the centre of the ingot. The authors say that the electrographic method of examination there illustrated was successful in showing the size and distribution of the lead particles throughout the ingot. In regard to the size that may be correct, but I fail to see how the distribution is revealed, for at and near the outside of the ingot the presence of very little lead is indicated, and yet, according to the analysis given in the Appendix, that part contains quite as much lead as the centre.

Dr. N. P. Allen (Metallurgy Division, The National Physical Laboratory, Teddington): The appearance of the lead in steel is extraordinarily sensitive to the method of polishing the specimen, and Mr. Schofield is to be congratulated on his skill and patience in producing his photographs, Figs. 5 to 8, which leave no doubt that the lead-containing constituent is freely soluble in hot mercury.

Mr. Schofield failed in his attempt to show that this constituent melted at 327° C., but Messrs. Bardgett and Lismer were more successful, showing in one case that exudation occurred very close to 327° C. Had that been the only experiment, one would have been fairly satisfied that the constituent was pure metallic lead. The occurrence of exudations at 240° C. in another sample, however, introduces difficulties. A possible explanation is that the material exuding at 240° C. is another constituent of lower melting point, and that possibly the lead, as Dr. Whiteley suggested, has

dissolved antimony or tin or some other impurity. This, however, is negatived by the observation that the exudations contained 98% of lead. Messrs. Bardgett and Lismer offer the explanation that the lead is exuded at 240° C. in the solid state, because of high pressures exerted inside the steel, but if that is true it destroys the whole of the evidence, because if pure lead can exude at 240° C. the fact that in another specimen the exudation took place at 327° C. means nothing; it does not show that the material exuded was pure lead, and uncertainty still remains.

Mr. T. H. SCHOFIELD (Metallurgy Division, The National Physical Laboratory, Teddington): Messrs. Bardgett and Lismer have carried out a most detailed study of lead-bearing steels and have illustrated their experiments in a novel manner. The exudation test is most interesting but difficult to understand.

On p. 286 P the authors state that "lead segregates that are not actually present at the surface of the billet section under examination force their way to the surface under pressure at the temperature of the test." To what do the authors consider that this pressure is due? It is difficult to understand why lead or other material should be forced to the surface from the interior of the steel unless there is either considerable porosity or the lead is present in some form of network and the exudation is effected by some physical or chemical change such as, for example, the decomposition of a hydride. The greater thermal expansion of lead as compared with that of steel and the increase in volume of lead on melting—about 5–6%—are insufficient alone to account for the exudation even if one could assume that the lead or lead-bearing particles did not contain voids due to liquid shrinkage and volume change on solidification. It seems highly improbable that the phenomenon can be associated with changes in internal stress as suggested on p. 289 P, because exudation can occur so rapidly and at such low temperatures. Clearly there is a field for further research to determine both the properties of the white constituent, if it is not lead, and the precise mechanism which effects the exudation of lead.

JOINT CORRESPONDENCE.

(Figs. A to F = Plate XLA.)

Mr. A. ALLISON (Sheffield) wrote: In connection with the interesting series of papers presented from time to time on leaded ingots the following information may be useful by analogy in connection with the exudation test described by Messrs. Bardgett and Lismer.

Pots for holding molten lead for hardening steel articles were supplied in close-grained cast iron, but it was sometimes found that the molten lead leaked through the pot. One such leaky pot was examined by breaking, since much can frequently be learned from the fracture of cast iron. The fracture showed close-grained iron without cavities or unsoundness. A section examined under the microscope revealed nothing unusual except that possibly the phosphide eutectic was affected.

An exudation test of crude form was applied by placing a piece of this pot on a tripod over two or three Bunsen burners. The clean fracture of metal 2 in. thick literally sweated globules of molten lead. Although molten lead is known to be a liquid of extraordinary penetrative power, it is still considered surprising that it could penetrate through 2 in. of good solid cast iron.

This at once suggests that in hot-rolling leaded ingots into bars there may be considerable loss of lead by exudation, particularly in view of the

authors' experiments, although the physical properties of the steel may not be seriously affected by such loss.

Mr. T. E. Rooney (The National Physical Laboratory, Teddington) wrote: Some attempts have been made to separate lead or lead compounds from lead-bearing steels by the alcoholic iodine method, and in view of the discussion which has taken place on the papers presented to the Institute the results may be of some interest. No attempt was made to obtain a quantitative separation, but by careful control of the temperature of the iodine solution and slow stirring, particularly at the commencement of the reaction, it was possible to get a qualitative separation of lead, and in one case in addition a proportion of lead sulphide was obtained.

In sample 35/7295, No. 2 ingot of lead-bearing carbon steel (Table I. of Mr. Schofield's paper) metallic globules were separated. An X-ray examination of the residue confirmed that the globules were lead, and lead iodide and manganese sulphide were also detected. The lead iodide was probably formed by the action of the iodine on metallic lead. The spacing of the lines for manganese sulphide was slightly displaced, indicating the possibility of another constituent in solid solution.

A similar examination of a sample of lead-bearing ingot iron (Table I., sample *TTP*) produced a residue which contained lead, lead sulphide and an oxide of the type (Fe,Mn)O.

In these experiments no evidence was obtained of an intermetallic compound of lead and iron. The X-ray pattern indicated that the lead obtained was of reasonable purity, but no attempt was made to confirm this by chemical analysis.

With regard to the point made by Dr. Whiteley that the lead may alloy with tin or antimony contained in the steel, it is necessary that the purity of the lead added to the ingot should be taken into account in order to determine, by the chemical analysis of the extracted or exuded lead, whether alloying has actually occurred.

Mr. W. B. Wragge (Ledloy, Ltd., Stockport, Cheshire) wrote: Congratulations are due to Mr. Schofield for what appears to be the development of a very excellent technique for the polishing of micro-specimens of lead-bearing steels. The writer regrets that his own experiments on the lines indicated, even when polishing was carried out under a cooling agent such as water or paraffin, failed to reveal the white areas which, it is suggested, are lead. Whilst Mr. Schofield's method of removing these white areas by treatment with mercury would appear to be almost conclusive, nevertheless it would be interesting to note how such white areas respond to a scratch test.

It is perhaps worth noting that with samples possessing lead segregation and polished by normal means, a sodium–mercury amalgam appears to work better for removing lead from segregated areas. No effect was observed, however, on unsegregated areas.

The writer congratulates Messrs. Bardgett and Lismer on devising some very useful techniques for the electrolytic macro- and micro-etching of lead-bearing steels and on the very precise electrographic method that they employ for lead-printing. Whilst it is not proposed to discuss the results obtained by the authors in their investigation of the constituents comprising inclusions in lead-bearing steels, nevertheless it will be interesting to know more of the technique involved in identifying those inclusions. It is surprising, for example, that in the case of Fig. 11 etching with hydrofluoric acid apparently left granules of silica.

The novel method employed by the authors for following the progress of the exudation of lead under increasing temperature conditions by the micro-

scope has led to some very interesting results. It is worth emphasising that exudation, claimed to have occurred at as low a temperature as 30° C., took place only in a sample containing heavy segregates of lead. This observation is astonishing, as it would appear impossible for the lead to have alloyed with any constituent in the steel to produce an alloy of sufficiently low melting point to correspond with 30° C. The writer can only conclude that the phenomenon must be explained as being the result of some pressure caused by thermal effects.

The statement is made on p. 287 P, middle paragraph, that examination of a ground surface by lead-printing, electrolytic etching and the exudation test indicated freedom from discrete lead particles. In the case of steel specimens with homogeneous lead distribution this statement is true, because in the process of grinding the heat generated is sufficient to cause vaporisation and/or oxidation of the lead at the extreme surface, and to some extent cause the micro-exudation of lead from slightly below the surface. Experiments have shown that ground specimens may be very considerably denuded of lead to a depth of about 0·0005 in. This is sufficient, however, to render lead-printing and electrolytic etching ineffective. It is noted that the authors, when preparing specimens for their electrographic printing process, employ micro-polishing, this operation apparently being sufficient to remove the surface layer affected by previous grinding.

It is stated that the electrographic method indicates the size of lead particles throughout the ingots examined. With this statement the writer entirely disagrees. Scanning the photographic enlargements of electrographic prints given in Fig. 26, it would appear that the diameter of lead inclusions in the mid-portion of the ingot is of the order of 0·010–0·020 in., or in some cases somewhat larger. If this were true then under microscopic examination at normal magnifications the lead particles should appear immense. By the employment of a vacuum or neutral-atmosphere distillation technique a far closer approximation to the true diameter of the lead particles has been obtained. A piece of ingot cross-section corresponding with that employed for the production of Fig. 26 was kindly provided by Messrs. Bardgett and Lismer. Highly polished micro-specimens prepared from this sample were then subjected to a temperature of 1150° C. for 20 min. and 1350° C. for 10 min., both *in vacuo* and in a neutral atmosphere. After cooling down to room temperature, micro-examination revealed the development of grain boundaries plus the formation of cavities (*see* Figs. A to D). It would appear that the average diameter of the cavities in specimens taken towards the outside edge of the ingot is of the order of 0·00032 in., while in specimens taken from the centre portion of the ingot the average value is about 0·00048 in.

The temperatures and times employed are adequate to vaporise all the lead from the surface of the specimens, and also such lead as may be exuded to the surface. This is corroborated by the fact that lead prints carried out by the writer's chemical technique show the absence of lead from a specimen which had received distillation treatment. It is submitted, therefore, that the actual size of lead particle in this particular ingot cannot be greater than the average dimensions given above.

The depth of the cavities was found to be shallow, of the order of only 0·0001 in.; this figure was determined by removing the pitted surface of the specimen until the pits disappeared and noting the loss in weight. The surface of the specimen at this stage is capable of yielding a lead print. It is observed that sulphides leave cavities when subjected to the above vacuum or neutral-atmosphere distillation treatment (*see* Figs. E and F). This has been taken into account when computing the size of the cavities which appear to be the result of lead distillation and exudation. It would seem that the particle dimension will be very much smaller in

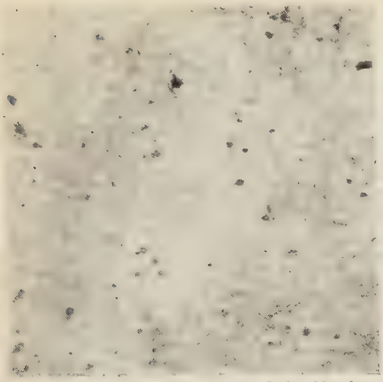

FIG. A.—Edge of Ingot; 1150° C. for 20 min.

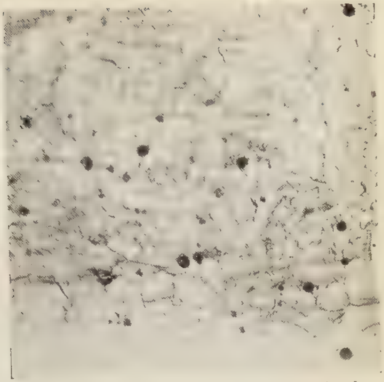

FIG. B.—Centre of Ingot; 1150° C. for 20 min.

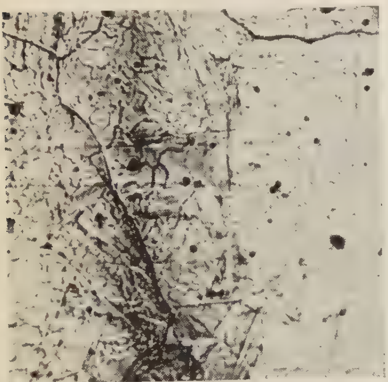

FIG. C.—Edge of Ingot; 1350° C. for 10 min.

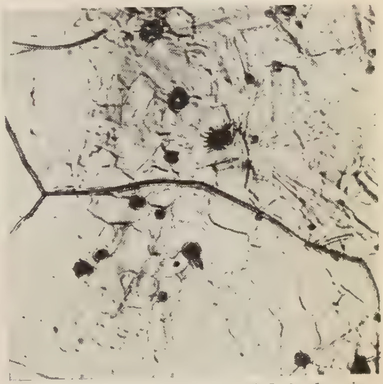

FIG. D.—Centre of Ingot; 1350° C. for 10 min.

FIGS. A to D.—Vacuum Distillation of Lead from Samples of Ingot Section supplied by Messrs. Bardgett and Lismer. × 150.

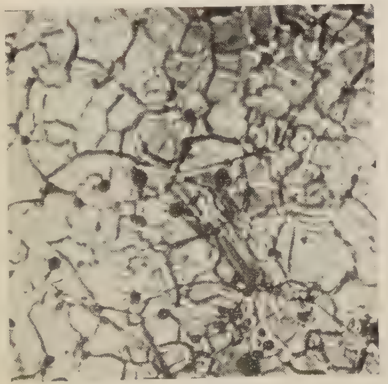

FIG. E.—Lead-free.

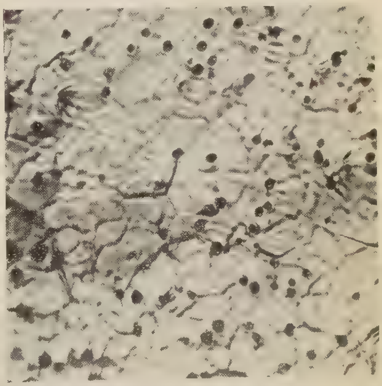

FIG. F.—Lead-bearing.

FIGS. E and F.—Vacuum-Treated Free-Cutting Steel Samples. × 250.
(*See* Wragge's contribution.)

(Micrographs reduced to four-fifths linear in reproduction.)

[*Correspondence on Lead-Bearing Steels.*
[*To face p.* 298 P.

FIG. G.—Unetched Surface of Leaded Steel. × 800.

FIG. H.—Same Surface as Fig. G Electrolytically Etched in 10% Ammonium Acetate Solution and Lightly Repolished. × 800.

(*See* Bardgett and Lismer's reply.)

rolled bar samples, hence the difficulty in microscopically observing lead in such specimens.

The vast difference between the above results and those obtained from the examination of the authors' electrographic prints is the result of very considerable lateral spread of chemicals in the gelatin medium employed on the paper. A very careful investigation confirms this observation.

In conclusion, both papers add very materially to the available knowledge on the subject of lead-bearing steels, though the perfect and infallible etching reagent for the micro-examination of such steel still remains to be found.

AUTHORS' REPLIES.

(Figs. G and H = Plate XLB.)

Mr. Graham wrote in reply : Dr. Whiteley remarks that he would expect coalescence and sinking of the lead to the bottom of the ingot. This in fact does occur, but I would emphasize that a well-made ingot will not show any detectable segregate, except in the extreme base of the ingot. Tests over a number of years have shown that by taking a bottom discard of 4%, all this segregate can be eliminated.

The leaded ingot had a thinner columnar zone and less sulphur segregation, indicating that it had cooled more quickly than the non-leaded ingot, and that there had been, therefore, less opportunity for coalescence and segregation of lead particles than would be expected.

It has been observed that if a leaded and a non-leaded ingot of otherwise similar steel are teemed side by side, the mould containing the leaded ingot shows red heat first, indicating the higher rate of heat transference from that ingot. I believe that this is due to the fact that the small air space between the ingot and the mould wall becomes filled with lead fume, which conducts heat better than the normal air space.

Mr. Schofield wrote in reply : It is noteworthy that Dr. Whiteley's experiments suggest that lead may be present in more than one form in lead-bearing steels and that similar conclusions are indicated by Mr. Rooney's work on the chemical separation of the inclusions.

From my metallographic work, I can only state that the white constituent appeared to be similar in all the four lead-bearing materials investigated.

I regret that Mr. Wragge has not been successful in applying this technique, which, as indicated in the paper, is tedious and calls for the exercise of considerable patience. I have not found it necessary to carry out the polishing under a cooling agent. Unfortunately it has not been possible to make scratch or hardness tests on the white constituent.

It is significant that the diameters of the cavities present on the surface of polished specimens of lead-bearing steels after heating in a vacuum or a reducing atmosphere at 1150° and 1350° C. are within the range of diameters of inclusions of the white constituent, which vary between 0·00005 and 0·0006 in. in the three lead-bearing steel ingots examined. In the wrought material, *S4405.HM* the elongated inclusions of white constituent vary between 0·0003 and 0·002 in. in length and 0·00005 and 0·0002 in. in diameter.

Messrs. Bardgett and Lismer wrote : In reply to Dr. Whiteley, we did not endeavour to reproduce the white constituent revealed so admirably by Mr. Schofield, since we had already succeeded, by our own methods, in revealing particles which were absent from non-leaded steels. The fact that such particles have been revealed by two different methods gives

additional support to the deduction that these are discrete lead or lead-bearing particles.

We would point out that definite identification of lead was made only in the case of segregated streaks present in the bottom-discarded billet. The appearance of the streaks was similar to that obtained on particles of lead shot polished by the normal technique adopted for ferrous materials. The presence of lead or lead-bearing particles in unsegregated material was determined by a special etching technique, which failed to reveal such particles in non-leaded steels. We draw attention to the fact that no claim is made that dark inclusions found in the samples were lead, and it is only to be expected that flattened lead globules would appear white.

Identification of the constituents of inclusions has been based mainly on their optical properties and they have been placed only in broad groups. It will be appreciated that some of the true character of inclusions is always lost in the reproduction of photographs of non-metallic inclusions. We note that Dr. Whiteley has not observed transluscent oxide inclusions, but that such do exist was shown by Portevin and Castro.* Whilst we agree that the presence of a silica particle within an oxide appears strange, we disagree entirely that our identification was dubious, since the silica particle was readily recognized by means of polarized light.

Although Dr. Whiteley obtained exudation at a temperature of 250° C. and further exudation at a temperature of 350° C., we would point out that the partial exudation of lead at temperatures below its melting point cannot be accepted as applying generally to all lead-bearing steels. In our experience, exudation may be practically complete either below or at the melting point, depending on the origin of the sample.

It is unfortunate that Dr. Whiteley is not prepared to believe what is a fact, namely, that distortion has occurred at the point of exudation, since a main feature of the mechanism of exudation is lost if this point is not accepted. Particularly careful polishing was necessary in the case of the specimen shown in Fig. 17, in order that the true form of the lead might be preserved.

The position of the lead particles during heating was revealed at the surface by etching in 10% ammonium acetate solution. If the special polishing under water or paraffin, suggested by Dr. Whiteley, were necessary to retain lead on the surface, normal polishing would result in a pitted surface; but this is not the case.

We consider that we should have little difficulty in identifying which of two samples, prepared in the ordinary way, contained lead. For example, Fig. G shows an unetched surface of leaded steel and Fig. H the same surface etched in 10% ammonium acetate solution and lightly repolished.

As regards the size and distribution of the lead particles throughout the ingot shown in Fig. 25, we had no intention of interpreting the prints in a quantitative sense; by distribution we referred to disposition of the particles, *e.g.*, the random distribution towards the outside and the presence of a network towards the centre.

We have not observed sufficient coalescence of lead to justify an expectation of an increase in lead towards the bottom of an ingot.

Dr. Allen seems to have ignored the fact that the specimen referred to as exuding at 240° C. was from a segregated position in a bottom-discarded billet, whilst the specimen exuding at 327° C. was from an unsegregated portion of an ingot.

In reply to Mr. Schofield, we have suggested the internal stress as a possible factor influencing exudation, since impurities are insufficient to account for the phenomenon and there is sufficient pressure to cause the

* *Journal of The Iron and Steel Institute*, 1935, No. II., p. 248.

lead to penetrate the steel surface. We entirely agree that further investigation of the mechanism of lead exudation is desirable.

Mr. Allison's contribution is most interesting, particularly the fact that, although nothing unusual was revealed by microscopical examination, marked exudation occurred on heating. We have experienced difficulty in revealing lead which has penetrated into cast iron, and this matter is receiving further attention. As regards the loss of lead on hot-rolling an ingot, we have not found any evidence of diffusion of lead through heated steel, and one would not expect, therefore, to find any loss to an appreciable depth.

We are interested in the results of Mr. Rooney's examination of residues obtained from leaded steels. Lead sulphide was not identified by us as a separate phase, and we deduced that this was in solid solution in the manganese sulphide. We notice that he has not suggested, as we did, the possibility of the presence of an oxide of the mixed type containing lead oxide.

Mr. Wragge has contributed some valuable information. With regard to Fig. 11, on etching a silicate streak associated with lead segregation, silica granules were apparently revealed, but this was probably due to the removal, rather than to the presence, of clear silica granules. With Mr. Wragge, we would emphasize that exudation occurred at as low a temperature as 30° C. only at heavily segregated positions in bottom-discarded billets. We also think this slight preliminary exudation is due to pressure caused by thermal effects.

On the question of the examination of ground surfaces, the surfaces of the samples were not ground on a machine, but by hand with emery papers. We are aware of the denudation of lead that takes place in a machine-grinding operation. A micro-polish is not essential for electrographic printing, but gives the best definition. As in all printing methods of examination, the image of the particles is greater than the particle size, and the relative size of particles cannot be precisely obtained from the prints. Tendencies only towards increase or decrease in size are indicated, as we have stated in the paper.

Regarding Mr. Wragge's method of determining the size of lead particles by the vaporization of the lead and measurement of the resultant cavities, it seems doubtful whether the diameter of the cavity represents the diameter of the lead particle. Comparison of the photographs suggests that the diameter is dependent on the temperature of heating. Further, in ingot sections, it may be assumed that the lead particles are more or less globular. In consequence, if the diameter of the cavity is 0·00048 in., then the depth would certainly be expected to be greater than 0·0001 in.

THE INFLUENCE OF CENTRIFUGAL CASTING UPON THE STRUCTURE AND PROPERTIES OF STEEL.*

By L. NORTHCOTT, D.Sc., and D. McLEAN, B.Sc.

(Figs. 1 to 18 = Plates XLI. to XLIV.)

Synopsis.

Thick cylinders of a nickel-chromium-molybdenum steel were cast by the centrifugal process using chill moulds rotating about a horizontal axis without a central core. The casting conditions were varied with respect to mould speed, casting temperature and rate of pouring, and each casting was examined to determine the influence of these factors upon its structure. At low mould speeds the delay in pick-up, *i.e.*, the interval required for the metal to be rotated with the mould, resulted in the liquid tumbling about inside the mould, thus giving rise to the coarse circumferential zoning referred to as "type 1 structures," the explanation of which was confirmed by the preparation, at low mould speeds, of a series of castings in which the mould was rapidly stopped before the completion of solidification. Above these mould speeds there was an optimum range which gave uniform structures, referred to as "type 2 structures," free from segregation. At the highest mould speeds the vibration developed resulted in circumferential bands of segregate referred to as "type 3 structures," and the association of such segregation with vibration was confirmed by tests on vibrated unrotated ingots.

High casting temperatures favour the formation of the type 1 structure, reduce the type 3 banding at high mould speeds and promote the growth of large primary crystals. Rapid pouring was found to favour segregation and, at high speeds, to lead to radial cracking; circumferential lapping was only observed under conditions of very slow pouring and at low casting temperatures.

The mode of solidification of castings showing the different structures is discussed.

Introduction.

The centrifugal casting technique has now been applied commercially to the casting of a wide variety of metals. The distinguishing feature of the process is the use of a rotating mould which, by reason of the centrifugal action involved, leads to a pressure on the liquid, in a direction away from the axis of rotation, of several times the gravitational force normally available for filling the mould. Hence the process can be considered as one in which the fluid pressures involved are dependent upon the mould speed and the density of the metal.

Experience with centrifugally-cast materials has shown that certain peculiarities in structure could occur owing to this special method of casting. A comprehensive investigation was carried out in the Armament Research Department on the metallurgical aspects of the process as applied to a wide range of alloys, both horizontal and vertical machines being employed. Some experimental work on the centrifugal casting of several non-ferrous alloys has been described elsewhere,[1] and reference should be made to this for details of the particular casting machine used and the general scheme adopted in this work. The investigation has been extended to the centrifugal casting of a nickel-chromium-molybdenum steel and the results are described in the present paper.

* Communication from the Armament Research Department, formerly the Research Department, Woolwich, received July 3, 1944.

The purpose of the investigation was to determine the influence of variations in casting temperature, mould speed, mould temperature and rate of pouring upon the structure, segregation and properties of centrifugally-cast thick cylinders with any one alloy composition.

Experimental.

The casting machine used in this work was one in which the mould rotated on a horizontal axis and had a range of speed from 100 to 1700 r.p.m. The standard procedure was to melt in a gas-fired crucible furnace a charge of from 60 to 70 lb. Castings 7·5 in. in length and of 6·4 in. outside diameter were obtained. Steel castings were withdrawn from the mould at a red heat, transferred to the crucible melting furnace and slowly cooled from about 1200° C. A transverse disc 0·5 in. thick was obtained from a position half-way along the length of the cylinder, and the face nearer the pouring end was prepared for examination; the rotation of the

TABLE I.—*Casting Conditions and Structures of Centrifugal Castings of Alloy Steel.*

Stamp Mark.	Pouring Time. Sec.	Casting Temp. ° C.		Splashing.[1] Sec.	Vibration.	Radial Cracks.	Banding.
		Hot.	Cold.				
			Mould speed, 450 r.p.m.				
TJZ.	60	1570	...	75	Barely perceptible.	0	Resembled type 1.
TKT.	8	1570	...	36		0	Type 1.
TKB.	12	1565	...	34		0	
TKE.	24	1565	...	65		0	
TKM.	33	1590	...	N.o.[2]		0	
TKD.	32	...	1510	39		0	
TJY.	5	...	1495	N.o.[2]		0	Type 1 plus another zone.
TJT.	16	...	1515	N.o.[2]		0	
			Mould speed, 650 r.p.m.				
TJW.	37	1550	...	67	Barely perceptible.	0	Type 1.
TJS.	7	1550	...	40		0	
TKC.	42	...	1490	N.o.		0	Trace of type 3.
TJN.	10	...	1510	N.o.	Noticeable.	0	
			Mould speed, 1000 r.p.m.				
TJX.	75	1555	...	N.o.	Noticeable.	0	Type 2.
TKO.	11	1565	...	25		0	
TKA.	65	...	1515	N.o.	Barely perceptible.	0	
TJP.	7	...	1495	N.o.	Noticeable.	3	
			Mould speed, 1450 r.p.m.				
TJM.	35	...	...	N.o.	Considerable.	0	Type 2.
TKP.	10	1565	...	N.o.		1	
TJU.	40	...	1520	N.o.		0	Type 3.
TJR.	5	...	1510	N.o.		4	
			Mould speed, 1700 r.p.m.				
TKS.	98	1570	...	N.o.	Very considerable.	0	Type 2.
TJL.	33	...	1490	N.o.		2	Type 3.
TJQ.	6	...	1490	N.o.		9	

[1] The time from beginning of pouring to end of splashing of the molten metal inside the mould. N.o. = no splashing observed.

[2] No splashing was observed, probably owing to the obscuring effect of flames and smoke, although the casting conditions were such as to favour splashing.

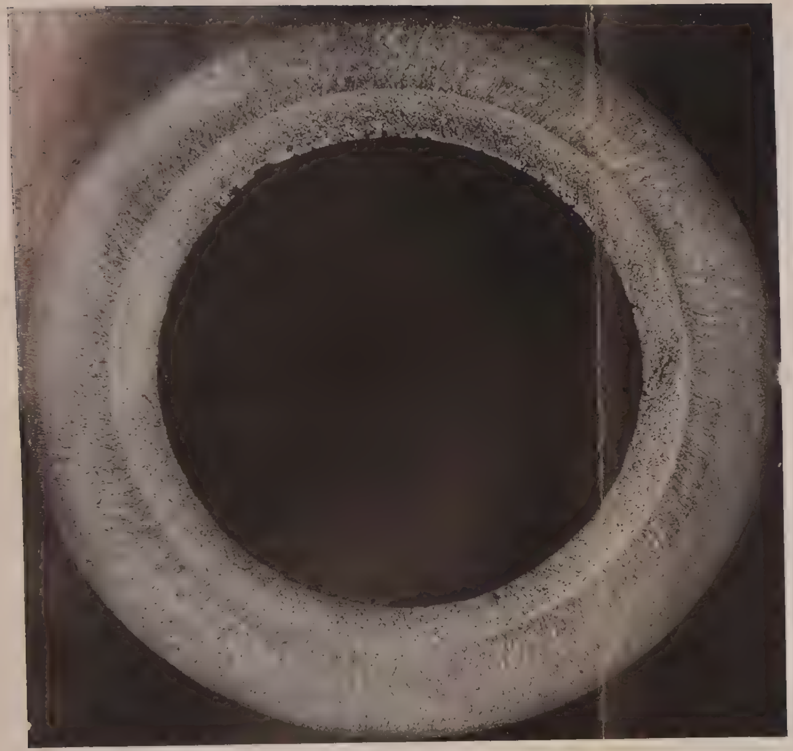

FIG. 1.—Macrostructure.

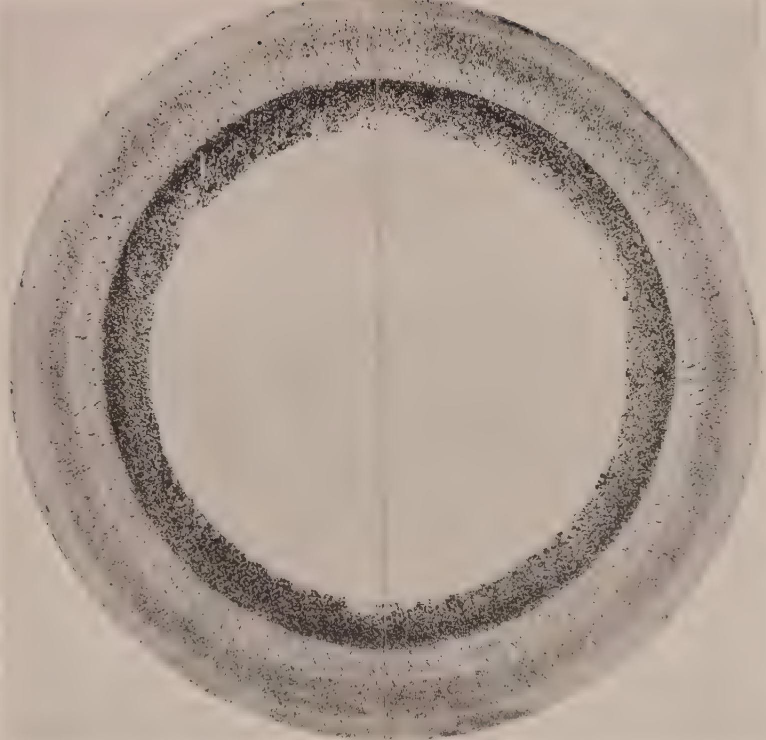

FIG. 2.—Sulphur Print.

FIGS. 1 and 2.—*TJW*, 650 r.p.m., 1550° C., 37 sec. Type 1 structure. × 1.

(Illustrations reduced to four-fifths linear in reproduction.)

FIG. 3.—Macrostructure.

FIGS. 3 and 4.—*TKK*, 450 r.p.m., 1575° C., 21 sec. Casting stopped during formation of ty[illegible]

[illegible] and McLean.

[*To face p.* 3[illegible] P.

Fig. 5.—Macrostructure.

Fig. 6.—Sulphur Print.

[illegible] r.p.m., 1565° C., 10 sec. Type 2 structure. × 1.

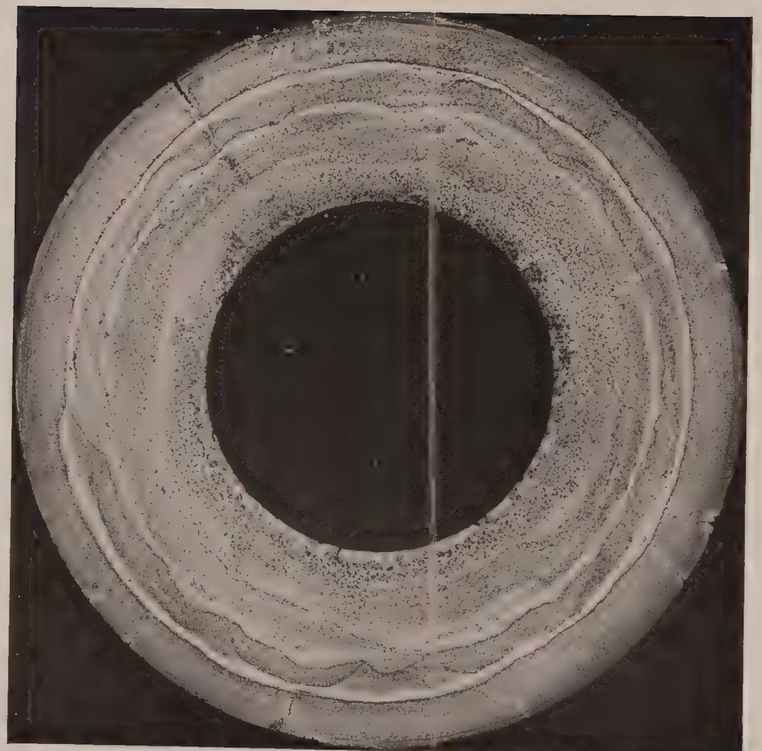

Fig. 7.—Macrostructure.

Fig. 8.—Sulphur Print.

Figs. 7 and 8.—*TJQ*, 1700 r.p.m., 1490° C., 6 sec. Type 3 structure. × 1.

(Illustrations reduced to four-fifths linear in reproduction.)

[Northcott [illegible] ran.

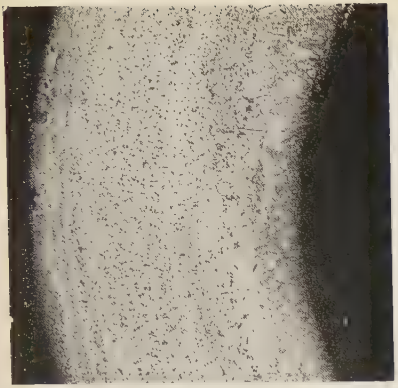

FIG. 9.—*TJU* (1450 r.p.m.). Slight Type 3 Banding. × 1.

FIG. 10.—Sulphur Print showing Type 1 Banding in outer half of *TJT*. × 1.

FIG. 11.—Microstructure near Rim of *TJQ*. × 50.

FIG. 12.—Microstructure of Inner Zone of *TJW*. × 50.

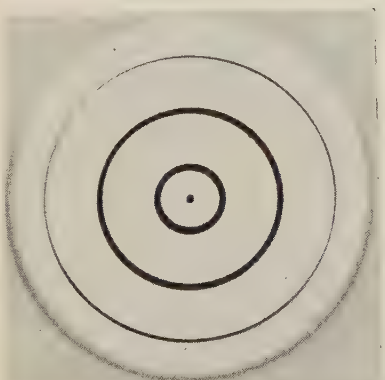

FIG. 13.—Absence of Vibration. Target revolving at 1000 r.p.m. × $\frac{3}{4}$.

FIG. 14.—Vibration. Target revolving at 1700 r.p.m. × $\frac{3}{4}$.

[*Northcott and Mclean.*

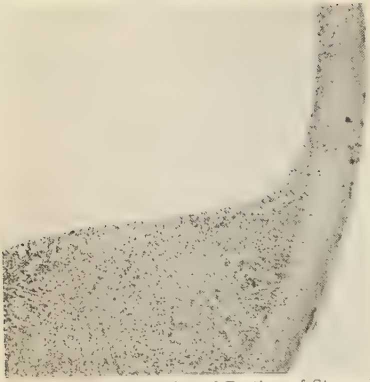

FIG. 15.—Sulphur Print of Portion of Stopped Casting *TKN*. × 1.

FIG. 16.—Sulphur Print of Portion of Stopped Casting *TKL*. × 1.

FIG. 17.—Sulphur Print of Axial Face of Vibrated Unrotated Steel Casting. × $\frac{3}{4}$.

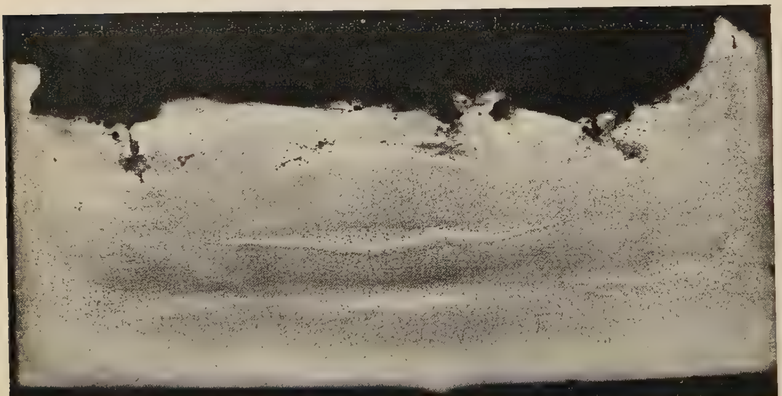

FIG. 18.—Macro-Print of Axial Face of Vibrated Unrotated Steel Casting. × $\frac{3}{4}$.

[*Northcott and McLean.*

[*To face p.* 305 P.

cylinder was such that this face rotated in an anti-clockwise direction. Longitudinal slices 0·5 in. thick were cut from the remaining portions of the cylinder.

The bulk of the work on steel has been carried out with a nickel-chromium-molybdenum steel of the following composition :

Carbon, 0·3%; silicon, 0·28%; manganese, 0·64%; nickel, 2·9%; chromium, 0·6%; molybdenum, 0·5%; sulphur, 0·04%; phosphorus, 0·03%.

The use of salamander crucibles for melting resulted in an increase in the carbon content over that given above, but this is not thought to have had any significant influence on the macrostructures.

The casting variables were controlled as follows :

Mould Speed.—As previous work had shown the mould speed to be the most important individual factor influencing structure and segregation, a wide range of mould speeds was studied, namely, 450, 650, 1000, 1450 and 1700 r.p.m. These speeds were determined with a tachometer attached to the machine and were checked by a stroboscope.

Casting Temperature.—The highest temperature obtainable was 1590° C., and this temperature was therefore selected for the hot metal. Cold casts were poured at about 1510° C., the liquidus of the alloy being in the neighbourhood of 1455° C. The casting temperature recorded for the individual castings refers to the temperature of the steel in the crucible furnace as determined with a quick-immersion couple immediately before casting.

Pouring Speed.—The time taken for rapid pouring was normally less than 12 sec. The purpose of slow pouring was to maintain a layer of melt in the die as thin as possible, so the time was generally greater with high than with low casting temperatures; the actual times for all slow-pouring experiments usually exceeded 30 sec.

The particulars of the various castings are given in Table I.

Description of Structures.

Type 1 *Structures—e.g., TKB, TJT, TJY, TKT, TKE, TKM* and *TJZ* cast at 450 r.p.m. and *TJS* and *TJW* at 650 r.p.m. (*see* Figs. 1 and 2).

The characteristic feature is that the structure is composed of three sharply defined zones differing in purity and crystal structure.

At the outside there is a thin layer of chill crystals merging into fine columnar crystals; this first zone is of medium purity, and periodic structures are frequently observed. The columnar crystals are not truly radial but are inclined in the direction of rotation of the mould.

The second, or middle zone, is purer than the average as judged from the sulphur print; it is composed mainly of columnar crystals which are coarser than those in the outer zone and are slightly more inclined, although in the same direction. The whole of this zone etches more darkly than the outer zone and in the sulphur prints is lighter. It is often demarcated from the outer zone by a thin line of higher sulphur content followed by one of very low sulphur; this second line shows as a dark-etching line in the macrostructure (*TKK*, Figs. 3 and 4).

The third and impure inner zone is composed of equi-axial crystals. The second and third zones have a sharply defined common boundary, the inside one commencing with a band of especially impure material. This zone etches lighter and in the sulphur print is much darker than either of the other two zones.

The occurrence of type 1 structure is associated with splashing in the mould during the casting operation. At some stage during pouring, the

layer of liquid to be rotated by the mould becomes so thick that at low speeds the liquid entering last is not rotating fast enough to overcome gravitation; this liquid then tumbles about inside the mould. It is to this "pick-up" time that the column in Table I. headed "Splashing" refers. A low mould speed, fast pouring and a high casting temperature are the conditions most favourable to prolonged splashing. In the casting of one cylinder (*TKO*) at a higher speed (1000 r.p.m.) some splashing was observed, although the bulk of the liquid rotated with the mould, and in this instance segregation was negligible.

Type 2 (*Unbanded*) *Structures—e.g., TJX, TKO, TKA, TJP, TJM, TKP, TKS* (*see* Figs. 5 and 6).

With the exception of periodicity near the outer rim, segregated bands are absent in this type. The crystal structure is generally not uniform across the section, but may be composed of chill or columnar crystals at the outside, frequently showing periodicity, followed by small equi-axial crystals, large radial columnar crystals and, finally, large equi-axial crystals up to the bore of the cylinder.

Type 2 structures were observed at mould speeds of from 1000 to 1700 r.p.m., but at the higher speeds, where vibration is more pronounced, the casting temperature must be high and the rate of pouring low to avoid type 3 banding.

The conditions favouring a type 2 structure would appear to be freedom from vibration, high casting temperature and slow pouring.

Type 3 *Structures—e.g., TJL, TJQ* (*see* Figs. 7 and 8).

Highly segregated circumferential bands characterise these structures. The essential condition for their formation is marked vibration of the casting during solidification; low casting temperature is favourable to the formation of the structure, as is rapid pouring, but this latter was not a dominant factor over the range of casting conditions investigated.

The association of this type of structure with vibration of the casting has been confirmed by experiments with vibrated unrotated ingots described below.

Longitudinal Sections.

Macrostructures and sulphur prints of longitudinal sections showed the same features; the intensity of banding diminished towards the end of the castings.

Influence of Casting Conditions.

The influence of any particular variable in casting conditions depends to some extent upon the other variables, but, bearing in mind that the results described are intended to refer only to centrifugal casting in a machine of the type employed, the individual effects of mould speeds, casting temperature and rate of pouring may be summarised as follows :

Mould Speed.

The three structures described above occurred at low, medium and high mould speeds respectively.

Casting Temperature.

(a) *Mould Speed* 450 *r.p.m.*—A high casting temperature accentuated the composition difference (as shown by sulphur prints) between the inner impure zone and the other two zones in the type 1 structure generally formed at this mould speed.

(b) *Mould Speed* 650 *r.p.m.*—A high casting temperature favoured the

formation of type 1 structures and of larger primary crystals. When the casting temperature was sufficiently low, segregation was absent, especially with slow pouring.

(c) *Mould Speed* 1000 *r.p.m.*—The casting temperature alone had little influence on the structure or segregation apart from its effect on the size of the rim columnar crystals, which were larger at higher casting temperatures.

(d) *Mould Speed* 1450 *r.p.m.*—At this speed there was considerable mould vibration. Type 3 banding occurred at low casting temperatures (Fig. 9) but not at high temperatures; large primary columnar crystals were present in the body of the hot-cast ingots but not in the cold-cast ones.

(e) *Mould Speed* 1700 *r.p.m.*—At this speed the mould vibration was very considerable, and pronounced type 3 banding occurred in samples cast at low temperatures. High casting temperatures yielded coarse crystals in the middle of the cross-section, and there was much less evidence of segregation.

In general, therefore, the effect of raising the casting temperature is (*a*) to favour the formation of type 1 structure, (*b*) to hinder the formation of type 3 structure, and (*c*) to increase the size of the primary crystals.

Rate of Pouring.

(a) *Mould Speed* 450 *r.p.m.*—Reducing the rate of pouring increased the width of the outer band in the type 1 structure, as follows :

Ingot Mark.	Casting Temp. °C.	Pouring Time. Sec.	Average Thickness of Outer Band. In.
TKT	1570	8	0·05
TKB	1565	12	0·12
TKE	1565	24	0·31
TJZ	1570	60	0·70

Thick castings could not be prepared at high pouring rates owing to excessive accumulation in the lower part of the mould of liquid metal which then ran out of the pouring hole. Slow pouring, combined with low casting temperature, reduced type 1 segregation.

(b) *Mould Speed* 650 *r.p.m.*—The two hot-cast cylinders showed a type 1 structure, the outer band being thicker in the one poured slowly. What little banding was present in the cold-cast cylinders was more noticeable in the one poured quickly.

(c) *Mould Speed* 1000 *r.p.m.*—Pouring slowly reduces the length of the columnar crystals near the rim, but may lead to circumferential lapping near the outside at very low casting temperatures.

(d) *Mould Speed* 1450 *r.p.m.*—Rate of pouring had little effect on structure or segregation, but rapid pouring promoted the formation of radial cracks originating at the rim.

(e) *Mould Speed* 1700 *r.p.m.*—At low casting temperatures rapid pouring favoured the formation of type 3 structures and led to radial cracking.

In general, therefore, the effect of increasing the rate of pouring is to favour the formation of (*a*) banded structures, both type 1 and type 3, and (*b*) radial cracks at high mould speeds.

Some Properties of the Castings.

Microstructures.

Micro-examination of a large number of samples showed three main types of microstructures :

(1) Fine ferrite-pearlite distribution free from segregation (Fig. 11), occurring in the outer portion of most cylinders.

(2) As above, but showing slight evidence of columnar primary structure, as in the outer two zones of type 1 castings, composed of pronounced columnar crystals.

(3) Evidence of segregation of the dendritic coring type (Fig. 12), which was present in all samples near the bore, even in those free from banding; similar segregation characterised the whole of the inner zone of type 1 and of the bands in type 3 structures.

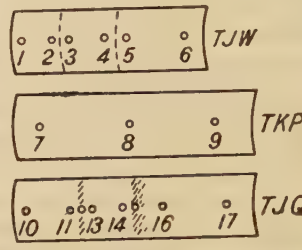

Fig. 19.—Positions of Samples for Chemical Analysis.

Chemical Segregation.

A casting representing each of the three main types of structure was selected for detailed examination. The close correlation found between the chemical analyses and the sulphur prints and macrostructures permitted some idea to be formed (by inference) of the chemical composition of the remaining castings by reference to their respective sulphur prints and macrostructures.

The three castings examined were *TJW* (type 1), *TKP* (type 2) and

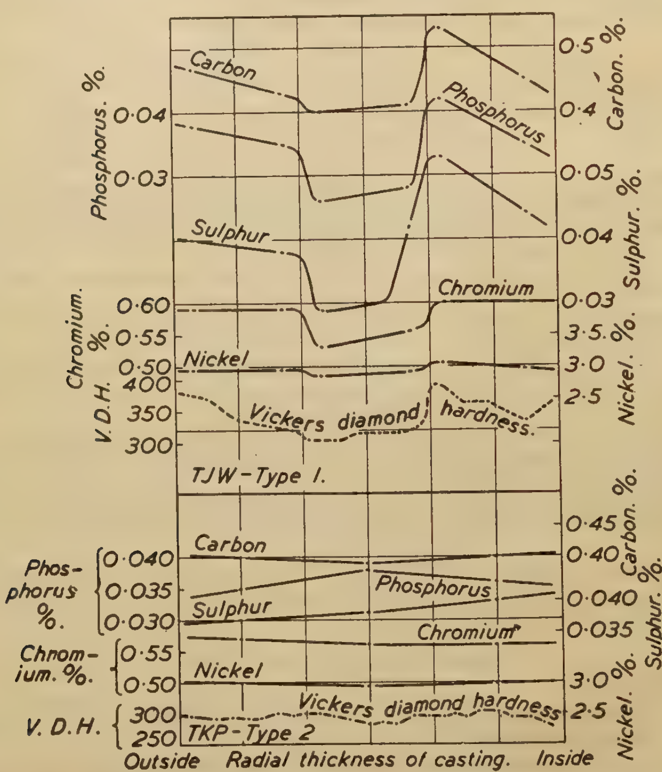

Fig. 20.—Variation in Composition and Hardness across Cylinder Wall.

TJQ (type 3). Samples were obtained by means of a drill of $\frac{1}{16}$ in. dia., but this necessitated the use of micro-chemical methods to deal with the small samples obtained, and the number of elements determined was limited for the same reason. Fig. 19 shows the actual positions of the samples relative to the zones of segregation. The results of the chemical analyses are plotted in Figs. 20 and 21; the sulphur prints were kept in mind when drawing the composition curves.

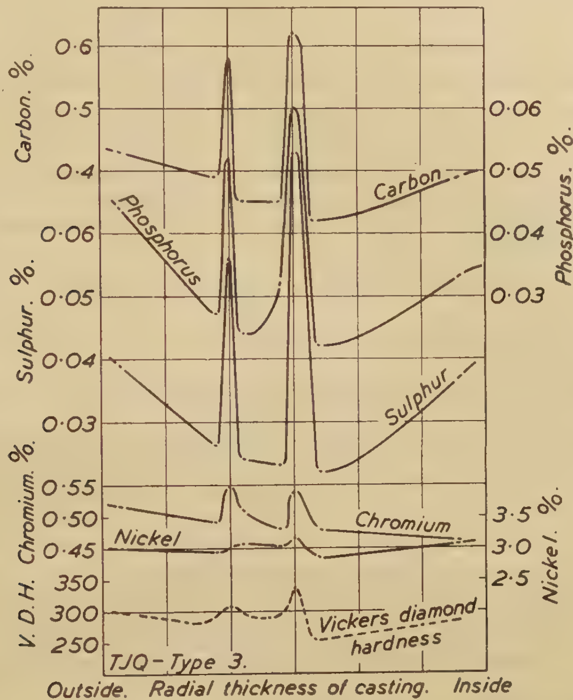

FIG. 21.—Variation in Composition and Hardness across Cylinder Wall.

The pronounced segregating tendencies of sulphur and phosphorus are clearly demonstrated; the variations in carbon content are also quite appreciable, particularly in *TJQ*. The chemical analyses confirm the segregation indicated by the sulphur prints and macrostructures.

Hardness.

Using the Vickers diamond pyramid machine explorations of hardness were made radially across sections of representative castings. The results are included in Figs. 20 and 21.

In type 1 and type 3 structures the hardness variations are considerable, the curves having the same general contour as the composition curves. By comparison the hardness of castings of the type 2 structures is very uniform across the section.

Brinell hardness tests on the three zones in the type 1 casting (using a 5-mm. ball) gave values of 297, 278, and 295 for the outside, middle and inside zones, respectively.

Vibration Characteristics.

A white card was inserted on the inner side of the door plate and, at slow revolutions of the machine, rings were marked to form a target. As the speed of the machine was increased, no change was observed in the target (Fig. 13), but at the higher vibrating speeds the rings became less clear (Fig. 14); the photographs of the target and front of the machine in motion were taken, using a stationary plate, and clearly show the nature of the amplitude of the vibration. The vibration was linear and reached a maximum diagonally, whilst normal to this the vibration was negligible. The maximum amplitude of vibration slightly exceeded $\pm \frac{1}{8}$ in.

In the non-ferrous work,(1) where a different mould setting was used, the speed of maximum vibration occurred at about 1460 r.p.m., and it is noteworthy that castings prepared at speeds much in excess of this (when the machine had passed beyond its period of maximum vibration) showed much less pronounced banding. In that work, too, it was found that castings prepared at the slightly lower speeds at which vibration was in the early stages of development showed banding of an intensity dependent upon the amount of vibration.

Properties of Other Centrifugally-Cast Alloys.

Steel.

A few castings have been prepared of a plain carbon steel of the following composition :

> Carbon, 0·48%; silicon, 0·16%; sulphur, 0·04%; phosphorus, 0·03%; manganese, 0·78%.

The structures were similar to those of the nickel-chromium-molybdenum steels prepared under the same casting conditions. An austenitic steel (Ni 12%, Cr 6%, Mn 6%) when cast under conditions of considerable vibration also showed type 3 structures.

Cast Iron.

An investigation into the centrifugal casting of grey, white and malleable cast irons showed that these materials also respond to the variations in casting conditions, but, as there are certain features peculiar to cast iron, for example, graphite form and size, nucleation number, response to malleablising anneal, &c., this work will be reported separately.

Non-Ferrous Alloys.

Many non-ferrous alloys have now been cast by the centrifugal process and the results obtained with three alloys, namely, 6% copper-aluminium alloy, 6% tin-bronze and 70/30 brass, have been described in some detail in an earlier paper.(1) It was shown that the three main types of structures described here were also obtained in these three alloys. The only difference observed in the behaviour of the different non-ferrous alloys was in the degree of segregation of which they were capable. For example, the banding of both type 1 and type 3 structures was most prominent in the copper-aluminium alloy and least in the 70/30 brass; careful polishing and etching of the latter alloy were required to develop the structure, whereas in the copper-aluminium alloy machining with a fine-pointed tool was sufficient to bring out the main features of both structures.

Taken in conjunction with the results described in the present paper, the important point brought out by the work on the non-ferrous alloys is that the tendency towards the formation of the different types of macro-structure is a characteristic of all alloys and is not confined to steels of any particular composition.

Mode of Solidification of Castings.

Radial Crystallisation.

In the centrifugal casting of cylinders prepared without the use of a central core, solidification proceeds radially from the outside of the cylinder in contact with the mould towards the inside. No evidence has been observed of crystals forming at the inner surface of the melt except immediately before the completion of solidification, but inward crystallisation may occur discontinuously.

Orientation of Columnar Crystals.

The outer columnar crystals are normally inclined at a small angle to the radius of the cylinder, as in Fig. 1. This phenomenon is not a drag effect, since the inclination is in the direction in which the mould rotates. Thus, the crystals slope to the right when the direction of rotation is anti-clockwise. According to the views previously expressed on the orientation of columnar crystals in ordinary steel ingots,[2] a melt flowing along a cooling face gives rise to columnar crystals inclined towards the direction from which the stream comes. In centrifugal casting the liquid does not immediately attain the mould speed, so the motion of the liquid relative to the mould is in the reverse direction, and columnar crystals growing towards the approaching melt should therefore slope in the direction in which the mould is rotating. The present evidence thus confirms the observations previously put forward.

The greater inclination of the columnar crystals in the pure middle zone of the type 1 structure (Figs. 1 and 3) can be explained by the fact that tumbling increases the average relative speed of rotation of solid and liquid by continually bringing the liquid metal back to its starting condition of no rotation.

Type 1 Structures.

Representative illustrations of type 1 structures are given in Figs. 1 and 2.

Several castings were " bled " by stopping the machine before solidification was complete. The liquid metal remaining collected in the lower part of the mould, leaving the upper part of the " casting " composed only of metal that was solid at the moment the machine stopped. The casting conditions were the same in all cases and were chosen to produce the type 1 structure.

The machine was stopped (*a*) at the commencement of splashing (*TKR* and *TKN* *), (*b*) after splashing had continued for a short while (*TKK*) and (*c*) immediately splashing had ceased (*TKL* and *TKQ*).

Particulars of these experiments are given in Table II., and photographs of the structures and sulphur prints are reproduced in Figs. 3, 4, 15 and 16. From these it would appear that the middle zone begins to form at the commencement of splashing; it finishes forming and the inner zone commences to form before splashing ceases. Splashing appears to cease at, or only slightly before, the completion of solidification.

The following tentative explanation of the formation of the type 1 structure is put forward. It has been shown that the outer band develops up to the stage at which splashing commences; owing to rapid chilling by the mould it may be expected that its composition would be similar to, or slightly purer than, the composition of the melt. The crystal structure is similar to that of the outer rim of many castings prepared without splashing.

* In the casting of *TKN* 4 sec. elapsed between the commencement of splashing and the stopping of the machine, but, at a later attempt (on *TKR*), this time was reduced to 1 sec.

TABLE II.—*Particulars of Stopped Centrifugal Castings; Mould Speed, 450 r.p.m.*

Stamp Mark.	Pouring Time. Sec.	Casting Temp. ° C.	Time [1] to Stop after Pouring. Sec.	Remarks.	Zones Present in Ring.
TKR.	7	1585	0 [2]	Machine stopped 1 sec. after commencement of splashing. Still splashing when stopped.	Outer first zone and practically none of middle zone.
TKN.	9	1590	0 [2]	Machine stopped 4 sec. after commencement of splashing. Still splashing when stopped.	Outer first zone and thin middle pure zone.
TKK.	21	1575	17	Still splashing when stopped, therefore there had been splashing for not less than 17 sec.	Outer first zone and middle pure zone.
TKL.	22	1585	30	Stopped within 2 sec. after splashing had ceased.	Outer, middle and inside zones, very little liquid metal left when stopped.
TKQ.	10	1575	20		

[1] Time interval between end of pouring and stopping of the machine.
[2] Pouring as well as machine stopped as soon as splashing commenced.

Rapid pouring, by favouring the early commencement of splashing, reduces the width of the outer zone as described earlier.

The middle zone starts to form at the commencement of splashing. From the combination of the columnar nature and the inclination of the crystals forming this middle zone, continuous growth has evidently occurred in the normal manner; it is unlikely that the purer crystals could have been formed in the liquid and have been centrifuged outwards, as this process is known to give rise to a random orientation of small equi-axial crystals. The greater purity may be explained by the tumbling motion ensuring a uniform composition in the liquid instead of it becoming less pure adjacent to the growing crystals. These, rotating with the mould, become cooler when out of contact with the bulk of the liquid during a part of each revolution, and this condition, leading to a steeper temperature gradient, will favour continuation of columnar growth. With the crystallisation and subsequent removal of purer solid during the formation of the middle pure zone, it would be expected that the liquid remaining would become more and more impure. The sulphur prints reveal this increasing impurity of the liquid as the splashing action proceeds (*TKN*, *TKL*, *TKK*; Figs. 15, 4 and 16, respectively).

It has been shown that the third or inner impure zone starts to form before the end of splashing. A study of a casting stopped during splashing showed the remaining liquid to have solidified in equi-axial form. The macrostructures and sulphur prints of this "stopped" impure liquid are identical with those of the inner zone of type 1 structure, and this suggests that the formation of crystals of random orientation in the liquid was taking place during the later stages of splashing. A stage would then be reached when, in effect, a sufficient number of crystals would be present to raise the viscosity of the liquid, which then rapidly assumes the rotational velocity of the mould and solidifies without further splashing. The small size of the equi-axial crystals is suggestive of their almost complete formation in the splashing liquid before this rotates with the mould.

In casting *TJT* the rate of pouring was high at first (to favour splashing) and then low, the casting temperature being relatively low at 1515° C. This casting showed evidence of type 1 structure in the outer portion but normal structure in the remainder; a section of the sulphur print is shown in Fig. 10.

Type 2 (*Unbanded*) *Structures.*

Representative illustrations of type 2 structures are given in Figs. 5 and 6.

In the absence of casting conditions giving rise to either the type 1 or type 3 structures, solidification proceeds normally, free from pronounced chemical segregation.

Type 3 *Structures.*

Representative illustrations of type 3 structures are given in Figs. 7 and 8.

In order to obtain a better impression of the structures obtainable under conditions of pronounced vibration, reference should be made to the illustrations accompanying the paper dealing with non-ferrous alloys.(1) In view of the close similarity between the structures of, say, the 6% copper-aluminium alloy and of the steel when prepared under similar conditions of vibration, observations made during the examination of the non-ferrous alloys have been taken into account in explaining the banded structures in the steel castings.

The vibration type of banding is considered to be due to the influence of vibration in limiting the undercooling which a liquid metal normally undergoes before it solidifies.

During the normal solidification of an alloy, cooling from the liquid state under conditions giving rise to a steep temperature gradient brings about a composition gradient in the liquid adjacent to the growing crystals, the plane of this gradient being perpendicular to the mould face. The liquid zone in contact with the solid becomes more impure and, therefore, of lower freezing point than the average, so that the liquid next to it, away from the mould, reaches its freezing point and starts to crystallise.(3) A repetition of this mechanism is responsible for the normal periodic structure already described.(4, 2)

The basic diagram representing temperature and composition relative to distance from the mould wall is shown in Fig. 22. The part of curve 1

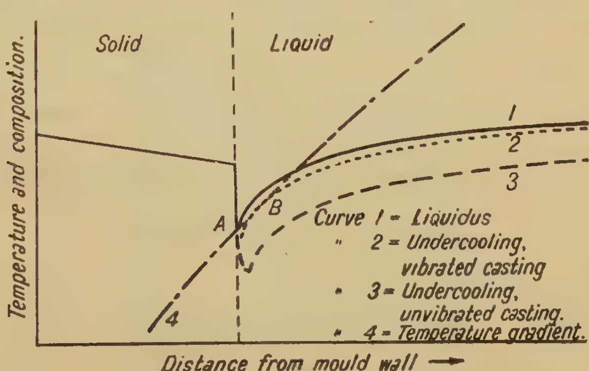

FIG. 22.—Temperature and Composition Relative to Distance from Mould Wall.

in the liquid region shows the liquidus temperature and composition. Under conditions of quiescent but rapid cooling the liquid undercools to curve 3 and solidification occurs at A, where this curve is cut by the temperature gradient, curve 4; crystallisation proceeds continuously at the solid-liquid interface and the formation of a periodic structure is thus

avoided. Under conditions of appreciable vibration, however, undercooling is very considerably reduced and the actual temperature of solidification is shown by curve 2. This results in independent crystallisation at some position B where the temperature-gradient curve cuts curve 2 away from the growing wall of the casting.

According to this view vibration banding is a true form of periodic structure which can be produced under milder conditions of temperature gradient, &c., than are required to yield periodic structures in the absence of vibration.

With centrifugal castings prepared in cylindrical chill moulds, crystallisation at B in the diagram means that a cylinder of solid metal will entrap a cylindrical zone of impure liquid A–B between itself and the growing crystal wall, with the result that the eventual solidification of the impure-liquid zone occurs under conditions of inadequate feeding, so that the segregate zones should be characterised by slight porosity, as they are found to be.

The maximum vibration occurred in the neighbourhood of 1700 r.p.m., but there was also appreciable vibration at 1450 r.p.m. The tendency towards type 3 banding was consequently less at this speed, but examples were observed and a portion of a macroprint of *TJU* cast at this speed is reproduced in Fig. 9.

In order to demonstrate that the type 3 structure is not peculiar to the centrifugal process, tests were made to show the influence of vibration during the casting of a number of small unrotated ingots. For this purpose a machine was constructed to vibrate the mould in a horizontal plane, the amplitude and speed of vibration being adjustable. To ensure conditions favouring uni-directional solidification a squat mould was used consisting of a copper base on which was fixed a warmed refractory cylinder, the height of the casting being similar to the wall thickness of the centrifugal castings. Ingots of the nickel-chromium-molybdenum steel were cast in this mould when oscillating, and control castings were obtained with the mould stationary. The examination of etched axial sections revealed an absence of segregation banding in the castings prepared in the stationary mould, but, under suitable conditions of vibration, banding was found to occur. Photographs of the sulphur print and macrostructure of the axial face of one ingot cast in a mould at an amplitude of $\pm \frac{1}{16}$ in. at a frequency of 1100 cycles per min. are reproduced in Figs. 17 and 18.

To determine the extent to which the steel used for the present castings was liable to undercool, cooling curves were obtained on a 340-g. sample (*a*) with the molten steel undisturbed but superheated to different degrees, (*b*) with the steel continuously stirred. The results are given in Table III.

TABLE III.—*Undercooling of Steel.*

Superheating.* ° C.	Stirred or Unstirred.	Remarks.
39	Unstirred	No recalescence.
47	Unstirred	Recalesced 5° C.
65	Stirred	No recalescence.
77	Unstirred	Recalesced 13° C.
20	Unstirred	No recalescence.

* The cooling curves were obtained in the order given.

Stirring was done with an alumina tube moved round by hand at the approximate rate of one revolution in two seconds, giving a speed of bulk motion of the same order as that of the vibration of the centrifugal casting.

The stirring prevented undercooling, which, in the absence of stirring, would have been about 10° C. About 40° C. of superheating is necessary with this steel under the conditions involved before undercooling occurs, and the undercooling increases as the superheating is increased. Bardenheuer and Bleckmann [5] found that 150-g. melts of Armco iron could be undercooled in the crucible as much as 250° C. after superheating not less than 20° C.; a 0·4% carbon steel could be undercooled 205° C. and a 12-kg. melt of low-carbon steel of normal composition could be undercooled 9° C.

The Formation of Cracks.

(a) *Radial Cracks.*

The number of radial cracks observed in the rings prepared for examination is included in Table I., and the appearance of typical cracks will be seen from Fig. 7. High mould speed and rapid pouring are the main factors favouring the formation of such cracks. No radial cracks were observed in castings prepared at mould speeds of less than 1000 r.p.m., but at higher speeds cracks occurred, provided that the rate of pouring was not too slow. Similarly, at very low pouring rates, no cracks occurred even at the highest mould speeds and casting temperatures adopted.

These results accord with the view that radial cracks may form in the early stages of solidification and that they are due to the centrifugal force of the rotating liquid promoting, in the thin solid shell, a circumferential tensile stress sufficiently high to crack the shell after it has contracted away from the mould. A high mould speed conduces to the formation of cracks by increasing the centrifugal force (the latter increases as the square of the speed). High pouring rates, by increasing the quantity of liquid in the mould in a given time and hence the total effective centrifugal pressure, also result in a higher stress in the shell and hence produce a tendency to cracking. It may be additionally effective by decreasing the rate of solidification so that a thinner shell is formed within any given time.

(b) *Circumferential Cracks.*

Circumferential cracking was only observed under conditions of very slow pouring at a low temperature. This gives rise to lapping, a process which can be responsible for cracks near the outside of the casting.

Summary and Conclusions.

The present investigation describes the effect of variations in the casting conditions upon structure and segregation in thick cylinders of nickel-chromium-molybdenum steel prepared by the centrifugal casting process, using chill moulds rotating about a horizontal axis. The most important individual factor appears to be the rotational speed of the mould. Low speeds result in splashing of the metal inside the mould, owing to the delayed pick-up; the casting is then composed of three sharply defined zones, the middle one being the most pure and the innermost zone the least pure (type 1 structure). Information on the mode of solidification of these castings has been obtained by rapidly stopping the mould at different stages of solidification during the preparation of a series of castings at a low mould speed. Castings prepared at intermediate mould speeds are relatively free from segregation (type 2 structure). At the highest mould speeds, noticeable vibration develops in the casting machine, and the resulting castings then show a circumferential-line type of segregation banding (type 3 structure). Experiments with vibrated unrotated ingots have confirmed the association between vibration and segregation and have shown that the effect is not peculiar to the centrifugal casting process.

The various structures and forms of segregation observed in all the centrifugal castings are discussed and explanations are put forward to account for them.

The influence of mould speed on the structure and segregation is modified by alterations in the casting temperature and rate of pouring. Apart from its normal tendency to increase the size of the primary crystals, a high casting temperature was found to favour the type 1 structure at low mould speeds, since it increases the time interval between the end of pouring and the completion of solidification. Rapid pouring favours the formation of the type 1 structure for the same reason. At high mould speeds rapid pouring also leads to radial cracking; on the other hand, excessively slow pouring and low casting temperatures give rise to circumferential lapping.

Hardness explorations show that the hardness curve closely follows the composition curve across the radial thickness of the casting, so that castings free from segregation are of uniform hardness throughout.

The results of the present investigation have been compared with those obtained in an investigation previously carried out on a series of non-ferrous alloys; close similarity in structure and segregation has been shown, and it has been found possible to duplicate the different types of structure and segregation in many different alloys.

Acknowledgments.

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DISCUSSION.

Dr. J. E. Hurst (Messrs. Bradley and Foster, Ltd., Darlaston) : This paper is one of a number on centrifugal casting which Dr. Northcott has presented, and he and his colleagues are to be congratulated on the work that they have done on this subject. I know the enthusiasm with which he has undertaken this investigation. I do not know whether I can be proud of our part in the work; the machine which he used is one that my firm supplied, and, though the fact that he has modified it since may exonerate us, when he showed a slide indicating the extent of the vibration (Fig. 14) someone whispered to me that we had supplied a machine with very wobbly bearings ! That, however, may have been fortunate, because it led him to investigate this trouble, which all those who have had anything to do with centrifugal castings have encountered. The late Dr. Hatfield used to call this structure the " bacon structure," and Fig. 6 of the present paper shows the appositeness of that description.

This work is of very great importance, for two reasons. One is that there has been a great development of interest in centrifugal-casting processes during the war, so that an unbiased, systematic investigation into some of the problems involved is of value. Secondly, the paper is valuable because of the light that it throws on the problems of heterogeneity in

solidifying alloys—not only steel, but the other alloys which Dr. Northcott has investigated. That is of particular importance at the present time, when we are deeply concerned with these unsolved problems of heterogeneity in solidifying metals and alloys.

The structures recorded by the authors, both in steel and in the other alloys which they have investigated, and which are dealt with in other papers, are familiar and well-recognised, and I can safely say that we have recognised the importance of speed of rotation and pouring conditions in their relation to the formation of these structures.

In connection with the question of speed, in the first paragraph of the paper the authors say that " the process can be considered as one in which the fluid pressures involved are dependent upon the mould speed and the density of the metal." I suggest to them that for " mould speed " they should substitute " metal speed," because it is the speed of the metal that matters; a good many of our problems in centrifugal casting are associated with getting the metal to rotate at the same speed as the mould. It is a question of accelerating this liquid metal. If the authors have that well in mind, it may help them when they go more deeply into some of the problems referred to in the paper.

I am not completely satisfied that the types of structure shown by the authors are completely explained in such a simple manner, by reference to mould speed and vibration. Take the type 1 structure. The authors are able to tell us, from their experimental work, that this type of structure is one which arises with low mould speeds, and is probably associated with the lack of pick-up of the metal during rotation—that is, the failure of the liquid metal to attain the rotational speed of the mould; during the time that it is accelerating it is constantly being turned over.

We obtain exactly similar structures, however, when we are centrifugally casting cast iron; we get a structure resembling the authors' type 1 structure in circumstances in which there is no question whatever that the molten metal has been accelerated to the speed of the mould, and there is no lack of pick-up. I have here a sulphur print, twenty years old, of a longitudinal section of a 4 ft. 6-in. cast cylinder liner; it shows the type of structure which the authors call type 1. With that casting there was no lack of pick-up of the molten metal; it was cast on a machine in which the mould was completely supported on rollers, and there was no question of an overhung mould, so that the vibration which the authors have experienced with the overhung type of machine did not arise. I do not say, of course, that there was no vibration at all; even in a mould carried on rollers we experience types of vibration, though of a different category from that in an overhung mould, and that may have some bearing on the type of structure produced. I think that that example must be characterized as a type 1 structure; it is certainly not a type 3 structure, because I do not think that the vibration had the characteristics to produce that type. I suggest that that is evidence that mould speed is not the sole arbiter of the presence of these layers.

We know that in the centrifugal casting of molten metal the conditions of turbulence of the liquid in the mould and the wave conditions set up in the liquid metal can be expected to have an effect on the structural character of the casting, and perhaps an explanation of some of these types of structure can be sought in that direction. At any rate, those factors cannot be ignored in forming an opinion of why some of these structures exist.

I want to make a plea for further investigation. I hope that the authors are not going to stop their investigations on the centrifugal-casting process, but that they will be inspired by the work which they have done already to endeavour to eliminate many of the complexities which exist in

in the ordinary centrifugal-casting process and practice. They have used a machine in which the controllability of the pouring and of the rotational conditions is limited. I think that an experimental machine could be built which would enable the authors to isolate these potential variants and to investigate them more deeply. I shall be pleased to do anything I can to obtain for them the authority and support necessary to continue their investigations, because the importance of the centrifugal-casting process is sufficient to warrant further experimental work of that kind.

In conclusion, I should like to ask what would happen in a very thin cylinder under the conditions which produced the structure shown in Fig. 8. I have in mind a casting not exceeding 1 cm. in thickness. Would the authors expect the zoning to be of so pronounced a character in such a casting?

Dr. J. H. WHITELEY (Consett Iron Co., Ltd., Consett, Co. Durham): Dr. Hurst has just said that the results given in this paper throw light on the problem of heterogeneity in steel ingots. That remark surprised me, for after reading the paper three times I found myself in complete darkness as to how the process of differential freezing can give rise to the banded structures there shown. I have for a long time past not been satisfied that the major segregation in steel ingots can be explained by differential freezing, and at present hold the view that heterogeneity begins in the liquid state at a point just above the liquidus. It may be that the "bacon" structures which are so liable to occur in centrifugal casting lend support to that contention and it might be instructive in that connection to give paraffin wax containing a coloured emulsion a similar centrifugal treatment in order to ascertain whether any banded separation of the emulsion takes place as the wax begins to solidify. I would remark that in Table I. a considerable overlapping of types 2 and 3 is evident, so that it is doubtful whether the authors have found the real explanation of the structural difference between the two types. Also in Fig. 20 a gradual rise in the sulphur content of the second zone of type 1 towards the third zone is indicated which is not to be seen in Fig 2. That seems to confirm my experience in sulphur-printing, for I find that it cannot always be relied on to give a true picture of the sulphur distribution.

Mr. S. SPRAY (Stanton Ironworks Co., Ltd., near Nottingham): I have been concerned with some experiments in the centrifugal casting of steel tubes, rather thinner but much larger than those of the authors, and must congratulate them on the way in which they have attempted to find explanations for difficulties encountered and structures produced.

I have a question to put in connection with the possible structure of thinner tubes. Dr. Hurst has already asked for information on this point relating to tubes 1 cm. thick, and it is this size with which I have dealt. In all cases, the banding was present in the type 1 form as in Fig. 2, without the dark-coloured inner band. As the thinner casting cools much more quickly than those investigated by the authors, one wonders whether this would result in a smaller number of bands in thin than in thick castings, under given conditions of temperature, vibration, &c.

Can the authors give us the composition of the mould and say whether they have had sufficient castings through it to give an idea of its life? In this connection, the mould thickness and temperature before and after casting would be of interest, and also the effect on the mould surface of repeated use.

There is no mention of physical tests on the spun steel, but perhaps we could be given some idea of the improvement, or otherwise, resulting from this treatment. Can the authors say what is the effect of the variations in

chemical composition or banding on the physical properties of test-pieces cut from the spun tubes and also their effect on further treatment, say, forging.

Mr. M. M. HALLETT (Sheepbridge Stokes Centrifugal Castings Co., Ltd., Chesterfield): Many of the phenomena described by the authors have been known in a general way, but they are the first to describe the facts with precision and to elucidate many of the underlying problems.

Of the three types of structure indicated, type 2 is the aim of the founder, while the conditions needed for the formation of type 1 are so far removed from normal that it is rarely encountered in practice. Interest is thus centred on type 3, which does occur and is often troublesome to eliminate.

There can be little doubt that the segregated bands observed in the last type of structure are due to the entrapment of relatively impure liquid between two disconnected walls of solid metal, and I am satisfied that vibration may be a contributory factor. At first sight, however, our experiences have been the reverse of those of the authors. On one battery of machines producing engine sleeves in a high-carbon nickel-manganese-chromium austenitic steel, there was severe vibration, but type 2 structures were produced regularly over many thousands of castings. On another battery of machines running at 1000 r.p.m. with little or no vibration, type 3 structures were frequently encountered, but on running these machines up to 1400 r.p.m. the structures were all of the desired type 2, although at the higher speed there was excessive apparent vibration. The term apparent vibration is employed, since it was revealed principally by heavy vibration of the floor around the machines. Such vibration is not necessarily a good indication of the actual vibration in the machines. On the other hand, a careful study of possible sources of vibration in the machine enabled steps to be taken which eliminated most of the defective type 3 castings in the machines running smoothly at 1000 r.p.m. In other words, it appears that only certain sorts of vibration induce type 3 structures and that satisfactory castings can be produced in the presence of quite heavy vibrations of other types. The authors' measurements of vibration by the observation of marked rings are very suggestive, but, to be completely satisfying, need to be carried out on machines loaded with molten and solidifying metal.

Attention might be drawn to the work of Herzog.* Using large steel ingots with a lower frequency but greater amplitude of vibration than those employed by the authors on their statically cast ingots, he was able to induce segregated bands of the authors' type 3 at whatever point in the solidification of the ingot the vibration was applied. It is a little surprising that there appears to have been no similar experience in the work carried out in America by Egler † on the jarring of solidifying steel ingots. It was claimed, in fact, that vibration eliminated segregation, though it should be noted that a much lower frequency of jarring was used.

Clearly one of the most fruitful lines of research following from this valuable paper will be the determination of the relative influence of the various modes of vibration on the formation of segregates, in order to ensure concentration in practice on the least dangerous modes.

Dr. E. W. FELL (Northern Aluminium Co., Ltd., Birmingham): In connection with the valuable work on centrifugal casting by Dr. Northcott and Mr. McLean, I should like to offer some views concerning the freezing of steel and non-ferrous alloys when in the presence of vibration.

* *Stahl und Eisen*, 1934, vol. 54, p. 462.
† *See* Egler and Tatman, *Journal of The Iron and Steel Institute*, 1934, No. II., p. 238.

It is clear that the freezing process resulting in type 2 structures is different from that causing type 3 structures. The state of the solid at any distance *s* from the mould immediately after it has frozen there may be regarded as depending on the sequence of states of the solid in the interval from the mould to the point *s*. This may be expressed diagrammatically as in Fig. A, where the state is defined as a function of both the chemical and physical structure of the solid.

The two curves shown in Fig. A for type 2 and type 3 structures are largely imaginary and are intended to indicate that the states for the two types differ at any given value of *s*. There is, obviously, one path or process of freezing whereby either sequence of states from the mould to a state *a* or to a state *b* may be reproduced, which is to retrace the respective path from *o*. These paths, *oa* and *ob*, are clearly complex and depend, for example, on temperature, on the components of the metallic system, on the motion of the system, and in this case apparently on the absence or not of vibration. Thus, the two processes of freezing have no simple correspondence with each other. I think it would therefore be advisable not to combine an explanation of the two types in one diagram as is done by the authors in their Fig. 22.

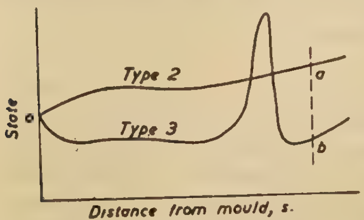

FIG. A.—Imaginary Solid States.

As to the origin of type 3 structures and the periodicities that they exhibit, I believe that a decrease of the temperature gradient for the liquid region, which occurs on an increase of the distance of the solid/liquid interface from the mould, permits increasing time for readjustment to take place between the solid and liquid during freezing, and hence a longer travel of the interface until there is sufficient accumulation of secondary component in the liquid to stop its motion. Thus, there seems to be a simple correspondence between the size of the gradient in the liquid region and the distance of travel from one " impure " ring to the next, though the vibration at the interface will, presumably, be also increasingly damped as the thickness of the solidified layer of metal grows, owing to the plastic nature of the layer at high temperatures, and this also has to be taken into account. That such readjustment may occur is suggested by the tendency for the impure rings to become farther and farther apart as the distance from the mould grows (*see* their Fig. 8; *also* Northcott and Dickin,* Fig. 49 for aluminium-copper alloy, presumably, and Fig. 35 for aluminium-magnesium alloy) and for the rings to become, likewise, more diffuse (their Fig. 8). Support of this view is given by the fact that a high casting temperature results in no impure rings being formed which would otherwise be present (*see* p. 307 P), because in heating up the mould with hot liquid metal, and so reducing the cooling rate, more time is provided for readjustment to occur, whereas a cold mould tends to promote them (*loc. cit.*, p. 308), because less time is available.

With regard to type 2 structures, since there are no impure rings and since there is little change in composition from the outside to the inside of the casting, it appears that " equilibrium " between the solid and liquid during freezing alters little as crystallisation proceeds (*i.e.*, as *s* increases in value). This may be expressed, in terms of unit volume, thus:

$$\frac{\text{Amount of secondary}}{\text{component in solid}} = \frac{\text{Amount of secondary}}{\text{component in liquid}} \quad . \quad . \quad . \quad . \quad (1)$$

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* *Journal of the Institute of Metals*, 1944, vol. 70, p. 301.

Equation (1) is thus satisfied by crystallisation of the solid in the presence of a composition gradient in the liquid which remains relatively constant while the solid/liquid interface moves.

With type 3 structures, " equilibrium " is now assumed to be different by being displaced a certain degree in favour of the crystallisation of material richer in primary component owing to vibration, so the amount of secondary component on the left side of equation (1) is then decreased and on the right side correspondingly increased. (Assuming the vibration and other conditions to be the same, the degree of displacement would, I expect, alter if the secondary component were changed for a different secondary component.) This shift from the left-hand side takes into account the supposition that the amount of secondary component in the liquid, in the neighbourhood of the interface, continually increases as the interface moves. Hence it would appear that (*a*) the greater the shift the shorter is the distance travelled by the interface with reference to the primary-rich section, and therefore the nearer one impure ring is to the next. It also appears that (*b*) the stage when the primary-rich state ceases to form may occur sooner, and the travel of the interface therefore be shorter, as the percentage of primary component in the original melt decreases. The number of impure rings per unit distance may thus increase in proportion to the percentage of secondary component, and I should like to ask the authors if this would be observed, say, in thick-walled centrifugal castings of aluminium-copper or aluminium-magnesium alloys.

The mould thickness and its composition should, I feel, be recorded in the paper (in view of the dependence of the cooling rate). It would appear desirable to complete the analysis of the vibration, for example by plotting, with respect to time, the displacement from its mean position of a point on the inner surface of the mould.

Dr. W. F. Chubb (London) : It would seem from the evidence provided that the authors have now arrived in their studies at a stage when the results to be obtained may be predicted with some accuracy, but others, on the other hand, have stated that they on their part have secured very different results under conditions assumed to be essentially similar. It cannot, of course, be asserted that the conditions in each case were the same, and there may accordingly be other facts not reported or examined which would be capable of explaining the varying results secured by different investigators. In this connection it seems to me that perhaps the segregation phenomena might the more readily be explained if further information concerning the condition of the metal before casting were available, and it would also be helpful in this respect to know more about the methods used for melting the metals. These considerations then lead to further questions which, I believe, the authors would do well to investigate, for it might be possible to overcome certain of the difficulties reported by modification of the composition, such as by alloying, and by methods of a like character. It would accordingly be interesting to know what is the effect of alloy additions, such as nickel, chromium, manganese, and other elements. This in itself, I think, would provide a very fruitful line of research, as would also attention to some of the details that I have enumerated, for it would appear quite possible that in this way some at least of the difficulties and differences discussed this afternoon might be explained and perhaps eliminated.

CORRESPONDENCE.

Mr. H. O. Howson (Sheffield) wrote: Centrifugal casting deserves more widespread attention in view of the increasing industrial importance of the process and also because of the extremely interesting structures developed on macro-etching. The elucidation of these banded structures is essential before the process is universally accepted, and, additionally, the study of centrifugal casting is a valuable aid to research into the solidification of metals and the consideration of structures encountered in statically prepared ingots and castings.

The principal value of the paper is the presentation in a very satisfactory manner of a full account of the macrostructures encountered in a series of centrifugal castings, prepared under carefully controlled conditions. The writer is in full agreement with the relationship between macrostructural appearance and casting conditions indicated by the authors. Many of the conclusions drawn in the paper are supported, but it is considered unfortunate that the size of casting examined has been confined to a diameter of 6 in. Observations on the appearance of castings of larger diameter (up to 40 in.) indicate an entirely different macrostructural pattern. The etched appearance of the casting bears a relationship to the bulk and diameter of the casting, and entirely fails to suggest any connection between vibration and periodicity. The theory of the limitation of supercooling by vibration advanced in the paper necessitates the conclusion that the resulting bands are of approximately annular form. In the larger-diameter castings the bands consist of small segments of circles interlaced in a complex pattern, which it is considered impossible to associate with vibration.

Practical observations in industrial centrifugal casting tend also to refute the vibration theory. The type of motion developed in the liquid must correspond to the condition described in hydraulics as "forced vortex." In this form of motion in liquids flow is extremely turbulent, and consists of a series of eddy currents, resulting in a very thorough mixing and agitation of the fluid metal. This eliminates any possibility of the existence of layers of liquid in different states of supercooling.

Castings are prepared under routine conditions in which a dial gauge, applied to the outside of the mould during teeming, indicates that the amplitude of vibration is less than 0·005 in. Under the appropriate casting conditions the resulting microstructure is found to be highly banded. Alternatively, a condition of extreme vibration has been induced on certain occasions, owing to lack of dynamic balance in the machine. In no instance has the slightest change in macrostructural appearance been observed capable of being correlated with vibration. If the casting conditions are suitable, banding may be almost entirely suppressed, even in the presence of intense vibration.

The writer suggests that the primary factor in determining the degree of banding in type 3 structures depends on the relative extent of the columnar zone. It will be observed from the examples exhibited in the paper that the columnar zone in centrifugal castings is entirely free from periodic structures, and additionally that the bands are confined to the area corresponding to the equiaxed zone. It therefore follows that if the conditions are such as to produce columnar formation—that is, high casting temperature, slow rate of pour and a high chill ratio—the metal is deposited in the form of columnar crystals free from banding. Alternatively, if the castings are poured at a rapid rate, the extent of the columnar zone will be restricted, and the banded non-columnar region will occupy a considerable area. Examination of the examples shown in the paper will confirm this viewpoint.

It is agreed that a columnar structure is preferable in centrifugal castings, as this zone generally presents mechanical test results of a very high order. The paper suggests that the banded structure forms under conditions of inadequate feeding in which liquid metal may be entrapped between the advancing faces of two separate series of crystals. In the opinion of the writer, any crystals of metal developing at a position remote from the liquid/solid interface would be immediately centrifuged outwards, owing to their greater density as compared with the liquid metal. The experience of over 2000 test results taken in the banded zones of centrifugal castings has shown a reasonable, and in some cases a very high, standard of mechanical properties. Radial tensometer tests have confirmed the absence of looseness across the bands, and no inherent weakness in this type of structure has been discovered. It is considered that the macrostructural appearance in centrifugal castings designated as banding arises through the presence of bands of material slightly lower in sulphur and phosphorus content than the matrix, but structurally completely continuous. A considerable series of analyses and hardness determinations across the face of steel centrifugal castings has shown complete uniformity, except for a zone approximately $\frac{3}{8}$ in. wide immediately adjacent to the bore. It is to be hoped that the authors will give further reflection to the mechanical properties of banded zones, if possible substantiated by tensile test results, or some retardation of this most interesting process may be caused.

Mr. J. F. B. Jackson (Messrs. David Brown & Sons (Huddersfield), Ltd., Penistone) wrote : The authors' investigations upon the influence of centrifugal (horizontal axis) casting on the mechanism of freezing are of particular interest to the writer, who has been associated with the industrial application of this production method for some years.

In discussing the various major factors influencing crystal formation the authors have stressed rotational speed as being of first importance. For any given alloy it is suggested that the ultimate and basic factor determining crystal structure in centrifugal castings, as in the case of static castings, is the rate of freezing alone. Such variables as rate of pouring, speed of rotation, casting temperature and metal fluidity, together with two others to which the authors have not referred, *i.e.*, mould temperature and mould length, all have predictable and specific influences upon the rate of freezing. Speed of rotation, in other words, is regarded by the writer as merely one of several secondary factors of similar importance.

Turbulence has been indicated as causing the fine-grained structure generally associated with the centrifugally cast product, and by way of support comparison has been made, fallaciously it is suggested, between the crystal formation of metal cast in stationary and in rotating moulds of similar dimensions and material. It is suggested by the writer that the fine-grained structure of the centrifugal casting is not the result of turbulence *per se* but purely of an ultra-rapid freezing rate. In other words, rotation of the mould causes the molten metal to be spread in a relatively thin film over the complete surface of the mould receiving it, causing an extremely rapid removal of heat and consequently a thick outer zone of chill crystals, on some castings extending even right through the section. In a static mould the area of mould/metal interface is very much reduced, particularly during the initial pouring stage, and even in the case of a metal mould the freezing rate is so reduced comparatively that only a thin layer of chill crystals is generally formed, followed by a marked columnar zone (*cf.* ferrous or non-ferrous ingot structures cast statically in metal moulds).

Turbulence as a major factor in the formation of fine grain-size in centrifugal castings is, in the writer's opinion, largely to be discounted.

Due credit must be given the authors for their demonstration of the phenomenon of banded segregation which can occur under certain conditions of casting, although the writer feels incompletely convinced by the explanation put forward as to its cause.

It would seem reasonable to visualise a possible mechanism whereby a thin band of segregate might occur between the intermediate columnar-crystal zone and the internal equiaxed zone in a centrifugal casting of heavy section, the residue of mother-liquor from which the columnar crystals have formed being trapped before diffusion could take place by the deposition of equiaxed crystals under centrifugal action. No such simple explanation would appear to cover either the form of irregular segregate that the authors have produced or even the existence of the line segregate between the outer chill-crystal zone and the intermediate columnar zone.

It should be stressed that the occurrence of the authors' banded segregation is not normal to industrial practice, where castings of heavy section, short length and small diameter would be regarded as unusual. In other words, the dimensions of the test-casting employed are such as to give rise to freezing conditions not normal to industrial practice, ferrous or non-ferrous, which fact may partially explain the lack of evidence from industrial sources of the existence of this form of heterogeneity.

Dr. D. Binnie (The Lancashire Steel Corporation, Ltd., Irlam, near Manchester) wrote: The authors have shown clearly the conditions required to form segregated zones in their centrifugal castings. Low speed of rotation, from a scrutiny of Table I., is by far the most important factor for developing the type 1 structure. Pouring time and casting temperature are of less importance.

The problem raised by the authors seems to resolve itself into asking the reason why, in the type 1 structure, the analyses of the castings for low rotational speeds should fall to a minimum at a position in the middle third of the depth, *see* Fig. 20. On p. 312 P the authors suggest that this may be due to the tumbling motion of splash ensuring a uniform composition in the remaining liquid instead of the liquid becoming less pure adjacent to the growing crystals. This view is reminiscent of an explanation of the pure-rim zone in rimming steels, and would suggest that splash in the freezing of centrifugal castings induces similar mechanics of freezing as are said to happen during the freezing of a rimming steel.

This would imply that the authors' type 1 zoning is due to a rimming action, which view is supported by a detailed analysis of a centrifugal casting (*TKE*) kindly forwarded for examination. Analysed for sulphur progressively at depths of $\frac{1}{8}$ in. from the outside to the inside, this casting yielded the following figures:

Position No.	Sulphur, %.	Sulphur Print Shade.
Outside 1	0·041	
2	0·033	Light
3	0·035	Light
4	0·028	Light
5	0·033	
6	0·055	Dark
7	0·048	Dark
Inside 8	0·053	Dark

Casting *TKE* was not sufficiently deoxidised, but this has not affected the zoning structure.

Rimming action can occur in normally killed steels such as have been used by the authors, *i.e.*, in tail-end ingots where some oxidising slag is allowed to work with the steel.

The authors' view that splash has caused type 1 zoning must be supplemented by the effect of a slight oxidation of the metal which could easily occur when splashed metal is churned through air. While splash has undoubtedly caused zoning, the zoning has formed by oxidation of the metal rather than by a purely mechanical homogenising of the liquid metal inside the freezing wall.

I have looked to see if there was any fundamental similarity between the zones as shown by the authors and those which are at times seen near the surface and in the lower portion of ingots.* Superficially they are similar. In Fig. 2, commencing at the outer surface, there is a light zone, a darkish zone, a light zone and then a deep dark zone. This sequence is similar to the zoning, when present, in ingots, and even the thin white line bisecting the first dark zone in ingots is present in the outer dark zone of Fig. 10. At this point the comparison stops. In the ingot the zoning is not brought up so readily by sulphur-printing and there are no analytical differences to explain the light and dark shades.

In a micro-examination of casting *TKB* two sharply defined structures were seen which corresponded exactly with the inner dark and light zones of the sulphur print. The inner dark zone could be seen microscopically to be higher in sulphur and it was also more porous. Heating to 925° C. for 3 hr. and again heating to 1150° C. for 4 hr. produced no change in the sulphur-print zoning, although during the latter treatment a large amount of diffusion had taken place.

The foregoing observations include some of the differences between ingot zoning and type 1 zoning in centrifugal castings. The two zonings, although superficially similar, are dissimilar on close examination.

AUTHORS' REPLY.

Dr. NORTHCOTT replied verbally : I should like to thank all those who have taken part in the discussion. There were a few questions about the effect of reduction in thickness of the castings; no great amount of work has been done on these particular steels, but similar work on the non-ferrous alloys shows that if the thickness of the casting is reduced sufficiently the type 3 structure disappears altogether, but with smaller reductions one coarse band only is observed.

Dr. Whiteley commented on the difference between Fig. 20 and the sulphur print. I think a simpler explanation of that would be that the sampling for analysis did not miss the inner zone.

I am convinced that the composition of the mould, provided that it is a chilled mould, does not matter at all. The only effective factor is the rate of cooling, and for both the type 1 and the type 3 structures a fair rate of cooling is required. In a sand mould the type 3 structure would be missed, but type 1 would not be. With regard to the repeated use of the mould, the effect with ordinary cast-iron moulds is to get craze-cracking, which is one reason why not infrequently steel moulds are used, though these suffer, or at least those that I have used suffer, from the defect of warping.

Mr. Hallett was justified in raising the criticism that under his particular conditions pronounced vibration did not necessarily lead to type 3 banding, whereas a lesser vibration did. One of the difficulties is to determine exactly what are the conditions of vibration in the metal as it solidifies. Our work on the influence of vibration on crystallisation tends to show that there are certain types of vibration which have peculiar effects. The original work of Herzog and others, carried out to reduce the crystal size, in some instances

* D. Binnie, *Journal of The Iron and Steel Institute,* 1944, No. I., p. 353 P.

was effective. In tests on the vibration of static castings we have found that vibrating at $\pm\frac{1}{32}$ in. at 1000 cycles per min. produced segregation in a 3-in. ingot, but no segregation at all, or segregation of a different type, with a 7-in. ingot.

The AUTHORS (in further reply) wrote: An interesting point emerging from the discussion is that some contributors consider type 1 banding to be more prevalent than type 3, and others the reverse. The remarks by Dr. Hurst are much appreciated, especially his encouraging comments and generous offer of further help. In reply to his final query, the general effect of reducing the thickness of a cylinder cast under conditions favouring the type 3 structure is to reduce the number of concentric segregated zones. When the cylinder thickness is reduced to about 1 cm. or slightly less, only one central zone of segregation is observed; with further reduction in wall thickness, segregation is reduced or eliminated completely.

The answer to the first question by Mr. Spray will be found in the reply to Dr. Hurst; the authors would, however, prefer to consider the banding observed by Mr. Spray as a particular example of type 3 rather than a modified form of type 1, since all type 1 structures show the dark-coloured inner band on a sulphur print. Similarly, without further data, the authors feel some uncertainty about Dr. Hurst's example of a type 1 structure in the 4 ft. 6-in. cast-iron cylinder liner. The composition of the cast-iron mould was typical for a mould iron, and the mould had been used for over 200 castings. The working surface of the mould crazed fairly readily, and was, therefore, machined from time to time, but, with the relatively short length of casting involved, considerable crazing could be tolerated before trouble arose in ejecting the casting. The mould was about $1\frac{1}{4}$ in. thick and was normally at 120° C. prior to casting. A few mechanical tests on small samples cut longitudinally from the different zones of thick centrifugally cast cylinders indicated that slight unsoundness affected the mechanical properties far more than variations in composition, the outer zones having higher strength than the inner zones near the bore.

In reply to Dr. Whiteley, it is doubtful whether the "bacon-streak" structure lends support to the liquid-separation theory, and the occurrence of type 1 structures in non-ferrous alloys is an argument against it, since evidence generally is opposed to liquid immiscibility in these alloys. The type 1 structure occurs only under a condition of violent splashing, which should mitigate pre-existing liquid separation and thus, on the separation theory, be less likely to produce the bacon-streak banding than should the comparatively quiescent condition which produces type 2 structure. The overlap of mould speeds for types 2 and 3 is due partly, as mentioned on pp. 306 P and 307 P, to the influence of casting temperature and rate of pouring, and partly to the existence of a subsidiary vibration which occurred in the casting *TJN*.

The authors consider that Mr. Hallett raises an important point. The vibration of surrounding floors, buildings, &c., is a poor indication of the vibration in the mould, as that of the former may be determined more by the proximity of the machine's speed to their natural frequency of vibration than by the amplitude of vibration of the mould. Furthermore, the kind of vibration of the mould is doubtless an important factor. It would seem that the precise condition which prevents undercooling is relative motion between adjacent layers of the liquid metal. This is not produced by all kinds of vibration. For instance, it is not produced by a "whirling" motion whereby the centre spot of the target in Fig. 13 describes a circle, although this motion could produce considerable vibration of surroundings. Vibration of the machine used in the present work was such that this spot moved to and fro along a straight line, and can be shown to produce the

required liquid motion. A dial gauge to measure the amplitude would obviously have to be placed in the correct position indicated by Fig. 14; two positions could be chosen to give zero reading. It will be seen that the authors are in full agreement with Mr. Hallett's suggestion that only certain sorts of vibration induce type 3 structures. As mentioned during the presentation of the paper, work on the effect of vibration on static castings indicates that there appears to be a critical range of frequency and amplitude of vibration for each size of casting, capable of inducing zones of segregation similar to those observed in the centrifugal castings. This explains the difference in the results found by Herzog and by Egler and Tatman to which Mr. Hallett draws attention.

The authors agree with Dr. Fells' explanation of the increasing diffusivity and separation of the bands in the type 3 structure with increasing radial distance from the outside circumference. Although it is not given in the paper, they had already arrived at this explanation and, since it fits in very well with existing ideas, they regard the observed behaviour of the bands as confirmation of the explanation of type 3 banding. In this explanation, " equilibrium " at the crystal/liquid interface is not assumed to be affected by vibration, as Dr. Fell appears to suppose. Vibration operates by inducing separate solidification a millimetre or two ahead of this interface.

The authors agree with Dr. Chubb that melting conditions, &c., may play a part, for instance, where grain-size is greatly affected by the degree of superheat, and that alloy additions may have considerable influence. On the other hand, experience with the centrifugal casting of a wide variety of alloys indicates that, although some alloys are more prone than others to show segregation effects, their inherent behaviour, if one may use the term, is only slightly affected by alloying additions.

The authors' views are completely opposed to those of Mr. Howson, who disagrees with the vibration theory. It is probable that the basis for the difference in view is, as mentioned above in the reply to Mr. Hallett, that each size and type of casting has a critical form of vibration required for the formation of banded structures. Mr. Howson's view that the liquid motion is extremely turbulent cannot be accepted as applying generally. If no other factor such as vibration or splashing intervenes, then after pouring is completed the liquid motion must soon become steady. However, should the mould speed wobble about a mean, peculiarities will doubtless be introduced into the liquid motion. The authors would like to draw Mr. Howson's attention to the remarks in the reply to Mr. Hallett on the use of the dial gauge. It is clear from Mr. Howson's remarks that he has been fortunate in experiencing in his thin cylinders only the less obtrusive forms of segregation where porosity would be either negligible or absent, and where the structures are in fact identical in appearance with the normal periodic structures.

Mr. Jackson's comments have been read with interest, coming as they do from one associated for many years with the industrial application of the centrifugal casting process. In the authors' view it is no simplification of the problem to consider the rate of freezing alone as the basic factor determining the crystal structure in centrifugal castings, and they would not concur with the view that the effect of turbulence can be discounted. Although the initial rate of solidification is high in centrifugal chill castings, turbulence is also high, and the two must be considered together. When a mould is filled with liquid metal, the temperature gradient under conditions of freedom from turbulence leads to the formation of long columnar crystals; turbulence, on the other hand, leads to a fine-grained structure whether there is a steep temperature gradient or not.* The authors cannot agree

* Northcott, *Journal of The Iron and Steel Institute,* 1941, No. I., p. 49 P.

with the statements comprising Mr. Jackson's last paragraph. It was owing to the occurrence of non-uniform structures in some examples of commercial castings that the investigation was initiated. Furthermore, the casting machine used was one commonly found in the industry, and the dimensions of the castings obtained are similar to those of castings made in large quantities in practice. In the present investigation the conditions of casting and the methods of preparing the specimens for examination were selected to accentuate the difference in the various structures obtainable, and there has been no lack of evidence from industrial sources of the various forms of structures observed.

The authors thank Dr. Binnie for his work on specimens from the castings. His view that additional oxidation occurs during splashing must undoubtedly be correct, but the authors are cautious in attributing too much importance to this factor. By itself it could not account for the impure inner zone of type 1 structures nor for the similar structures observed in non-ferrous alloys.

FIRST REPORT OF THE FOUNDRY PRACTICE SUB-COMMITTEE [1]

OF THE STEEL CASTINGS RESEARCH COMMITTEE.[2]

(Fig. 2 = Plate XLV.)

Paper No. 12/1945 *of the Steel Castings Research Committee (submitted by the Foundry Practice Sub-Committee).*

Summary.

The Foundry Practice Sub-Committee was formed to carry out practical investigations into problems encountered in the steel foundry.

Part 1 of this Report records their work on the running and risering of simple shapes used as 4-in., 6-in. and 9-in. cube castings, and castings 6 in. square by 12 in., 18 in. and 24 in. long. The effectiveness of various types of feeder heads normally used in steel foundries is illustrated by a large number of sawn sections of the experimental castings. The disadvantages entailed in using narrow heads which freeze prematurely, and the advantages of wider heads, insulated heads and the use of mould material which promotes differential solidification are illustrated. The establishment of definite principles to be applied to steel castings will help to remove some of the uncertainties which constantly confront the technical staffs of steel foundries.

The pioneer work in developing the so-called "whirl-gate" head method of feeding steel castings and recent examples of the application of this type of head and of the "atmospheric" head are described and illustrated in Part 2.

Foundry trials comparing the different types of feeding compound used on the fluid steel as it rises into the heads of steel castings are summarised in Part 3.

In Part 4 experiments designed to provide a numerical index of the tendency of different steels to hot tears in the mould are reported and attention is drawn to the fact that the degree of hot-tearing in a particular design can be minimised or often eliminated by using steel with a lower sulphur content.

In Part 5 some of the war-time activities of the Sub-Committee are set out, and some fundamental casting problems are mentioned in Part 6. With a return to more normal conditions attention will be devoted to further research under practical foundry conditions. The future progress of the steel castings industry depends on the co-operation of the foundry executives, who have already given so much of their time and experience, and on the generosity of individual manufacturers, who have carried out these practical experiments at their own expense.

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[1] Received January 20, 1944.

[2] A Joint Committee of The Iron and Steel Institute and the British Iron and Steel Federation, reporting to the Iron and Steel Industrial Research Council.

Introduction.

The Foundry Practice Sub-Committee was called into being by the Steel Castings Research Committee in 1938. The personnel has changed from time to time and the attendance of the Members has been seriously curtailed at times by pressure of work during the War, by the visits of some of the Members to America and by their service in Government departments and on other steel casting committees of more urgent national importance during war-time. The following is a list of members of the foundry industry who have served on the Sub-Committee :

Mr. M. Brown (Messrs. Hadfields, Ltd.).—Member from 25th May, 1938, to 18th April, 1940.

Mr. F. Cousans (Messrs. Thos. Firth and John Brown, Ltd.).—Member from 25th May, 1938, Chairman from 30th October, 1941, to date.

Dr. C. J. Dadswell (English Steel Corporation, Ltd.).—Member from 25th May, 1938, Chairman from 29th March, 1940, to 9th April, 1941, when he went to the Ministry of Supply (Iron and Steel Control). Resigned in July, 1943, when he left the steel casting industry.

Dr. C. H. Desch, F.R.S. (The Iron and Steel Industrial Research Council).—Member from 9th April, 1941, to date.

Mr. Basil Gray (English Steel Corporation, Ltd.).—Member from 1st July, 1943, to date.

Mr. S. J. Hewitt (Messrs. Edgar Allen & Co., Ltd.).—Member from 20th November, 1941, till his death on 1st October, 1943.

Mr. C. H. Kain (Messrs. Lake and Elliot, Ltd.).—Member from 20th November, 1941, to date.

Mr. S. Leetch (Davy and United Roll Foundry, Ltd.).—Member from 8th September, 1943, to date.

Mr. A. B. Lloyd (Messrs. F. H. Lloyd & Co., Ltd.).—Member from 11th December, 1941, to date.

Mr. F. N. Lloyd (Messrs. F. H. Lloyd & Co., Ltd.)—Chairman from 25th May, 1938, to 16th June, 1939, then on war service until he returned to serve as a Member from 18th April, 1940, to 20th October, 1940, when he resigned.

Mr. J. E. Mercer (Messrs. Head, Wrightson & Co., Ltd.).—Member from 25th May, 1938, to date. Also represents the General Steel Castings Association.

Mr. G. S. Palmer (Messrs. Wm. Jessop & Sons, Ltd.).—Member from 10th November, 1942, to date.

Mr. A. Riley (Messrs. Wm. Jessop & Sons, Ltd.).—Member from 18th April, 1940, to August, 1942, when he joined the Royal Air Force.

Mr. W. H. Salmon (Messrs. Hadfields, Ltd.).—Secretary. Member from 10th November, 1938, to date.

Mr. T. R. Walker (English Steel Corporation, Ltd.).—Member from 25th May, 1938, to date. Also represents the Institute of British Foundrymen.

Mr. John T. Wells (Messrs. Edgar Allen & Co., Ltd.).—Member from 26th November, 1943, to date.

In addition, the following have attended occasional meetings of the Sub-Committee : Mr. D. A. Oliver (Messrs. Wm. Jessop & Sons, Ltd.), Mr. T. W. Ruffle (Messrs. Lake and Elliot, Ltd.) and Mr. H. T. Protheroe (Sheffield University).

The Sub-Committee desire to express their warm appreciation of the services rendered by Members, and wish to thank the steel

foundry firms who have made trial castings and submitted reports to them. The Sub-Committee arrange a rota of visits to the foundry of each Member in turn and wish to thank these firms for their hospitality on these occasions.

This Report is divided into the following Parts :

PART 1.—The Feeding of *Simple Shapes* of Steel Castings.
(*a*) 4-in., 6-in. and 9-in. Cubes.
(*b*) Castings 6 in. Square by 12 in., 18 in. and 24 in. Long.
PART 2.—The "Whirl-Gate" Head and the "Atmospheric" Head.
PART 3.—Feeding Compounds.
PART 4.—Hot Tears.
PART 5.—War-Time Activities (Design and Production Committees, Problems such as Thin Plates and Shells).
PART 6.—Fundamental Casting Problems to be Considered.

PART 1.—THE FEEDING OF *SIMPLE SHAPES* OF STEEL CASTINGS.

(a) 4-*IN.*, 6-*IN. AND* 9-*IN. CUBES.*

The Sub-Committee was set up to carry out practical investigations into problems encountered in the steel foundry, the following programme of initial work being approved :

(1) The investigation of the effects of heading and running methods on the solidification cavities formed when casting simple shapes in steel. The object of the investigation is to enunciate rules to be followed in order to obtain :

(*a*) The absence of any cavities or segregated areas in a casting.

(*b*) The best ratio of the size and weight of the feeder head to those of the casting, and the best position of the feeder head.

(2) The application of the results from (1) in the study of selected castings.

(3) The investigation of the influence of brackets and chills, both external and internal, in avoiding defects or overcoming manufacturing difficulties.

Current practice in the methods of gating, running and feeding steel castings is based on the training and subsequent experience of individual foundry workers. Reliable records of past practice are very scanty, and little quantitative investigation has been carried out in a systematic way.

It was considered that by making in different foundries steel castings in the same simple shapes, with heads which were alike

in a single series and altered from one series to another, some quantitative rules could ultimately be enunciated which, when applied in specific cases, would indicate the arrangement likely to give the best results.

So far as steel castings themselves are concerned, the most important factors governing the feeding of the castings and the formation of cavities are :

(1) The ferrostatic pressure in operation.
(2) The dimensions of the runners and gates.
(3) The location of the runners in relation to the castings.
(4) The shape and size of the castings.

To obtain results of value it was therefore considered necessary to maintain these factors constant for each series investigated. The properties of the steel as poured into the moulds influence the results, and to minimise any variations from this cause it was decided to standardise as far as possible the process used in the method of steel manufacture, the chemical composition of the steel, the temperature of the molten metal, and its fluidity or running quality. Foundry conditions, such as casting in green or dried sand, the size of moulding box, and the rate of filling the mould, also affect the results, and an attempt was therefore made to ensure that these conditions also were the same in any series of tests.

The examination of the castings to determine what feeding had taken place, and what cavities had been formed, was carried out by sectioning the casting and head medially.

When the Sub-Committee, as a preliminary step, surveyed the practice current in different foundries, it was found that widely divergent views were held concerning the procedure regarded as best, even in the simplest cases of running and feeding, and equally good results were secured in foundries using substantially different methods. It was also evident that considerations of the most efficient feeding did not always govern the methods of running and feeding. Other conditions sometimes become of greater practical importance in any given foundry, and these are referred to in more detail at the end of the present Report.

Investigation of the Feeding Characteristics of Steel in Dry-Sand Moulds.

The Sub-Committee decided, in the first stage of the investigation, to study the feeding characteristics of steel in dry-sand moulds, in order to determine the most economical methods of feeding a casting such as a ring of which the section was 6 in. square. The problem was to be attacked in steps, beginning with the effect of different common shapes of heads on 6-in. cube castings, using a 0·20% plain carbon steel made in a basic electric-arc furnace, with a standard ladle addition of 8 oz. of aluminium per ton of liquid steel. The castings made were divided into seven groups, more

details of each group being given below. The results obtained are illustrated by a photograph and tracings of the sawn sections and by a diagram in which the height of sound steel in the head, or the extent of the pipe cavity in the cube, is related to the weight of head used. This is expressed :

(*a*) As a percentage of the weight of a solid cube, and
(*b*) as a percentage of the total weight of the cube and head.

Group 1.—6-*in. Cube with Six Different Heads.*

Six types of head were selected to represent the practice adopted in different foundries. They are illustrated in Fig. 1, which also

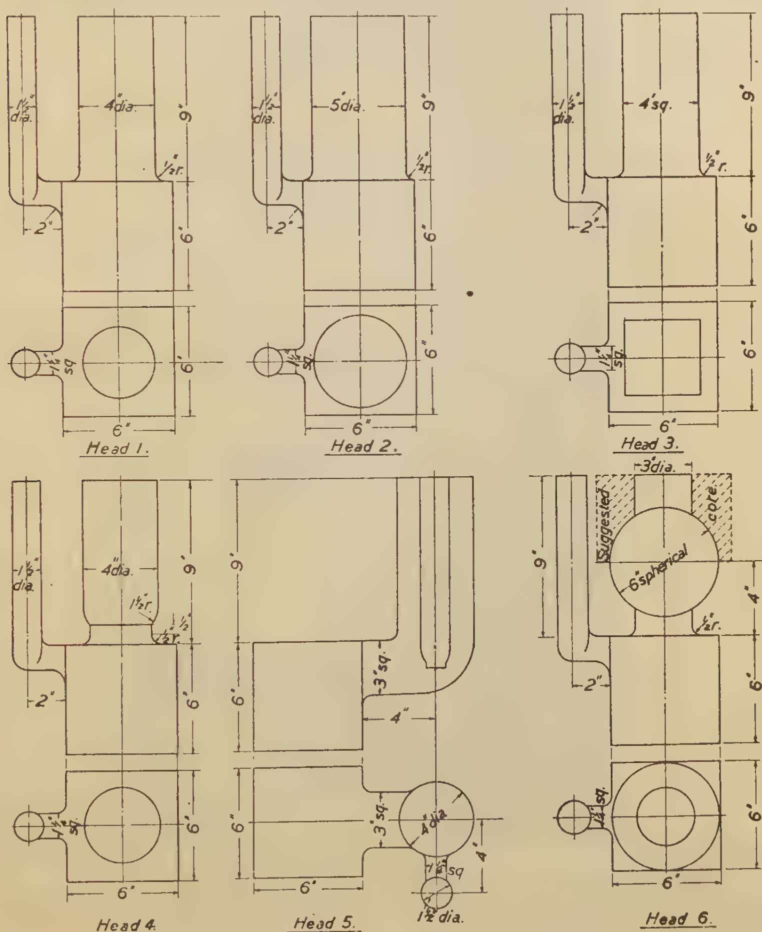

FIG. 1.—Designs of Heads, 6-in. cubes, group 1.

gives the dimensions and positions of the runner and ingate. All heads were 9 in. high and may be described as follows :

(1) Cylindrical of 4 in. dia.
(2) Cylindrical of 5 in. dia.
(3) Square section of 4 in. side.
(4) Cylindrical of 4 in. dia. but reduced at the bottom.
(5) Cylindrical of 4 in. dia. but L-shaped.
(6) Spherical of 6 in. dia., necked to 3 in. dia. at the junction with the cube.

One series of these cubes complete with heads is shown in Fig. 2. One complete set of six cubes with different heads was made in each of four foundries, the six cubes being cast in succession from a single heat. Observations were made at the melting furnace and in the foundry, and have been tabulated.

Table I. records the nature and weight of the furnace charge, and the times occupied by different stages during the heats.

TABLE I.—*Steelmaking Data for* 6-*in. Cubes, Group* 1.

Foundry.	Nominal Capacity of Furnace. Tons.	Charge.		Power on to Slag off.		Slag off to White Slag.	White Slag to Tap.	Total.	
				Hr.	Min.	Min.	Min.	Hr.	Min.
A	$3\frac{1}{2}$	Carbon steel turnings	60·7%	3	10	35	35	4	20
		Carbon steel scrap	37·5%						
		Manganese steel scrap	1·8%						
		Total, 76 cwt.	100%						
B	4	Carbon steel scrap, 82 cwt.	100%	2	19	49	90	4	38
C	3	Carbon steel scrap	96·7%	2	47	43	15	3	45
		Manganese steel scrap	3·3%						
		Total, 60 cwt.	100%						
D	$3\frac{1}{2}$	Foundry scrap	35·5%	2	40	25	12	3	17
		Turnings	21·5%						
		Miscellaneous	43·0%						
		Total, 70 cwt.	100%						

Table II. lists the chemical analyses of metal samples taken during the heats and also of the finished steels.

Table III. gives the results of temperature determinations at the furnace and in the foundry, the size of nozzle used and the time taken to cast each cube.

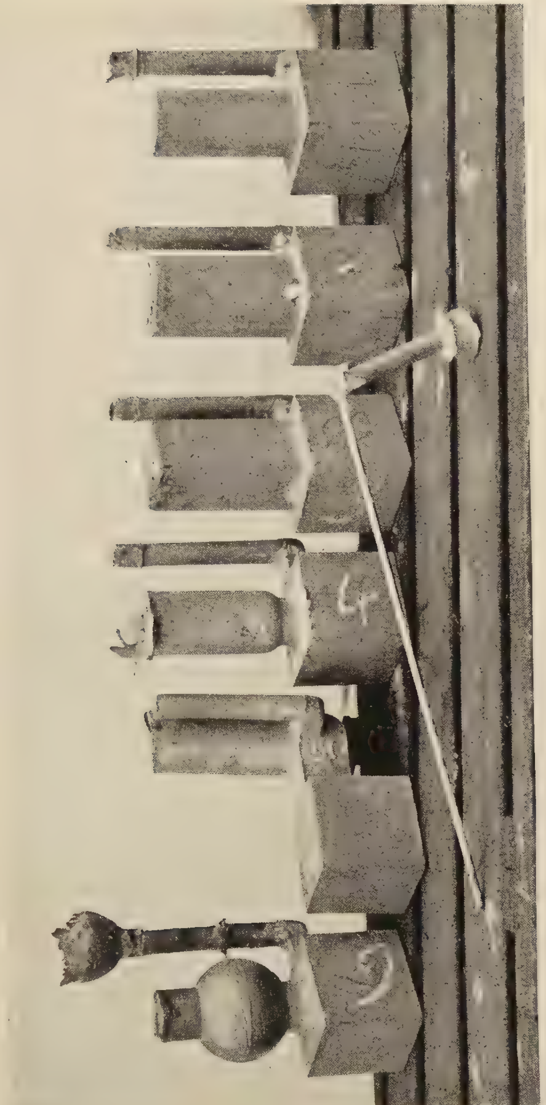

Fig. 2.—6-in. Cube Castings with Different Heads, as follows (reading from right to left): (1) Cylindrical, 4 in. dia. (2) Cylindrical, 5 in. dia. (3) Square, 4 in. side. (4) Cylindrical, 4 in. dia., reduced at bottom. (5) Cylindrical 4 in. dia., L-shaped. (6) Spherical, 6 in. dia., necked to 3 in. dia. at junction with cube.

TABLE II.—*Analyses of Metal Samples, 6-in. Cubes, Group* 1.

Foundry.	Sample.	C. %.	Si. %.	Mn. %.	S. %.	P. %.	Ni. %.	Cr. %.
A	Melted	0·83	...	...	...	0·035	0·11	...
	Slag off	0·21	...	...	...	...	...	...
	White slag	0·135	...	0·40	...	...	...	...
	Finished steel	0·19	0·41	0·90	0·018	0·030	0·20	0·09
B	Melted	0·18	...	...	...	0·014	...	...
	White slag	0·11	...	0·44	...	...	...	...
	Finished steel	0·245	0·19	1·14	0·017	0·022	0·10	0·12
C	Melted	0·05	0·04	0·08	0·040	0·018	...	...
	Before tap	0·15	0·33	0·80	0·020	0·023	...	...
	Finished steel	0·19	0·29	0·74	0·019	0·023	Nil	0·03
D	Melted	0·17	...	...	...	...	...	...
	White slag	0·20	...	...	...	...	...	...
	Finished steel	0·25	0·48	0·79	0·036	0·035	Nil	Trace

TABLE III.—*Casting Data for 6-in. Cubes, Group* 1.

Foundry :	*A.*	*B.*	*C.*	*D.*
Apparent temp. before tapping. ° C.	1485	1485	...	...
Ruff test, length of rod. In.	14, 14¼	7	27½ *	14
Apparent tapping temp. ° C.	1510–1520	1480	1530	...
Foundry remarks	Normal heat, slightly sticky.	Cold.	Nice medium heat, high fluidity, ran clean from nozzle with no "tear."	Hot, fluid.
Dia. of nozzle. In.	1¼	1½	1½	1⅜
Time between tapping and casting first cube. Min.	11	7½	4	7–10
Weight cast before first cube. Cwt.	15	21	15	15
Apparent casting temp. for cubes. ° C.	1435–1425	1405–1395	1510 (corrected)	...
Time to cast cube—				
No. 1. Sec.	10	5	10	...
No. 2. Sec.	30	8	10	...
No. 3. Sec.	22¼	9	10	...
No. 4. Sec.	14	6	10	...
No. 5. Sec.	14	6	10	...

* Test not under standard conditions.

Table IV. contains the weight of each cube and head, and the weight of each head expressed as a percentage of the weight of a solid 6-in. cube ($61\frac{1}{4}$ lb.). As all the cubes showed some secondary piping, the percentages quoted are somewhat lower than they should be. This Table also gives the rate of pouring the castings.

TABLE IV.—*Weights of 6-in. Cubes and Heads, Group 1.*

Foundry :		*A.*	*B.*	*C.*	*D.*
Weight of cube plus head after removal of runner. Lb.	Cube No. 1	$88\frac{1}{2}$	$92\frac{1}{2}$	$87\frac{1}{4}$	84
	„ 2	$106\frac{1}{2}$	$106\frac{3}{4}$	104	105
	„ 3	98	98	$96\frac{1}{2}$	89
	„ 4	$87\frac{1}{2}$	$89\frac{1}{2}$	$87\frac{1}{2}$	83
	„ 5	103	$105\frac{1}{4}$	$103\frac{1}{2}$	101
	„ 6	$94\frac{1}{2}$	$99\frac{1}{4}$	$95\frac{1}{2}$	87
Weight of head as percentage of solid cube of $61\frac{1}{4}$ lb. %	Cube No. 1	44·5	51	42·5	37·2
	„ 2	74	74·5	70	71·6
	„ 3	60	60	57·6	45·3
	„ 4	42·5	46·2	42·9	35·6
	„ 5	68·1	71·7	69·1	65
	„ 6	54·4	62	56	42·1
Rate of casting. Lb. per sec.	Cube No. 1	8·85	18·5	8·7	...
	„ 2	3·55	13·3	10·4	...
	„ 3	4·40	10·9	9·6	...
	„ 4	6·29	14·9	8·7	...
	„ 5	7·36	17·5	10·3	...
	„ 6	5·56	12·4	9·5	...

Table V. gives information regarding the moulding boxes used and the drying operation. Ample room was allowed for the castings, to eliminate any variations due to the use of too small boxes. A wide difference in the times between drying the moulds and casting will be noticed.

TABLE V.—*Moulding and Drying of 6-in. Cubes, Group 1.*

Foundry.	Moulding Boxes.	Drying Time. Hr.	Drying Temp. °C.	Period Elapsed before Casting.
A	24 in. sq., 10 in. deep (hand-rammed).	20	183–205 (chimney end) 288–330 (fire-box end)	3 days $20\frac{1}{2}$ hr.
B	20 in. dia. inside, top 9 in. deep, bottom 12 in. deep.	...	350–370	2 days $5\frac{1}{2}$ hr. (torch dried)
C	16 in. square.	...	425 (for 2 hr.)	5 hr.
D	In one large box, 9 in. between each casting and next.	...	390	$5\frac{1}{2}$ hr.

Table VI. shows the order of merit of the six castings in each series, so far as they could be arranged. For ready reference the conditions of steel temperature and pouring speed are included.

TABLE VI.—*Order of Merit, 6-in. Cubes, Group* 1.

Head No.	Head Volume as Percentage of Cube Volume. %	Order of Merit in Foundry—			
		A.	*B.*	*C.*	*D.**
1. (4 in. dia.)	52	5	4	4	...
2. (5 in. dia.)	82	1	1	1	1
3. (4 in. sq.)	67	2	2	2	...
4. (4 in. dia., reduced at base)	50	4	5	6	...
5. (4 in. dia., L)	74	6	6	5	6
6. (6 in. dia., spherical)	65	3	3	3	...
Summary of Casting Conditions.					
Steel		Warm.	Cold.	Warm.	Hot.
Pouring		Slow.	Fast.	Fairly fast.	...

* Intermediate order cannot be given, as thick saw-cut removed some cavities.

The diagrams in Fig. 3 show the sectioned cubes. There is good general agreement in the results obtained in the four different foundries, but in no case was a completely sound cube without pipe obtained, whether the cubes were cast hot or cold, and fast or comparatively slowly. The L-shaped head appeared to be inefficient in all cases. There was some indication that, with a fixed height of head, feeding was more efficient with heads of large cross-sectional area. Constricting the head at its base appeared to be disadvantageous.

In view of these results the Sub-Committee decided to carry out experiments with other shapes of head. Subsequently, the effects on pipe formation of additions of feeding compounds in the head, and of using insulating sand mixtures to line the head, were investigated. Other trials were made incorporating miscellaneous methods of gating and casting, and some attempt was also made to discover evidence bearing on commonly held views regarding the differences in feeding qualities of steel made by various processes, such as basic electric, acid open-hearth and converter steel. Finally, a few experiments were made to find if cubes of other sizes and masses (4-in. and 9-in. side) gave results substantially different from those obtained from 6-in. cubes cast under similar conditions. In some of these later experiments two other foundries participated, and these are referred to in diagrams and Tables under the letters *E* and *F*.

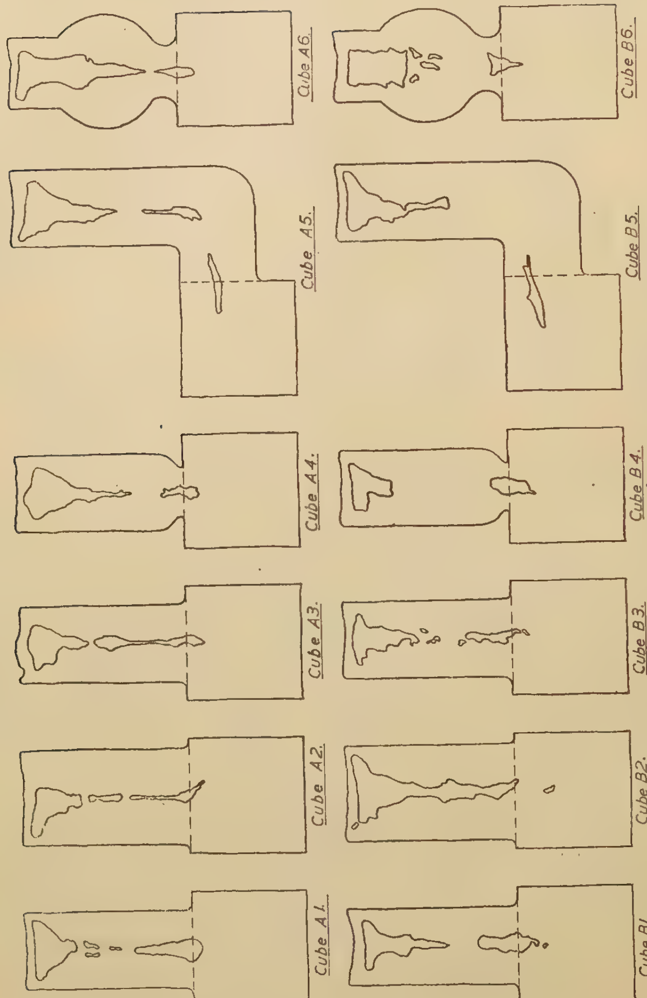

FIG. 3A (*see* Fig. 3B).—Sections of 6-in. Cubes, group 1.

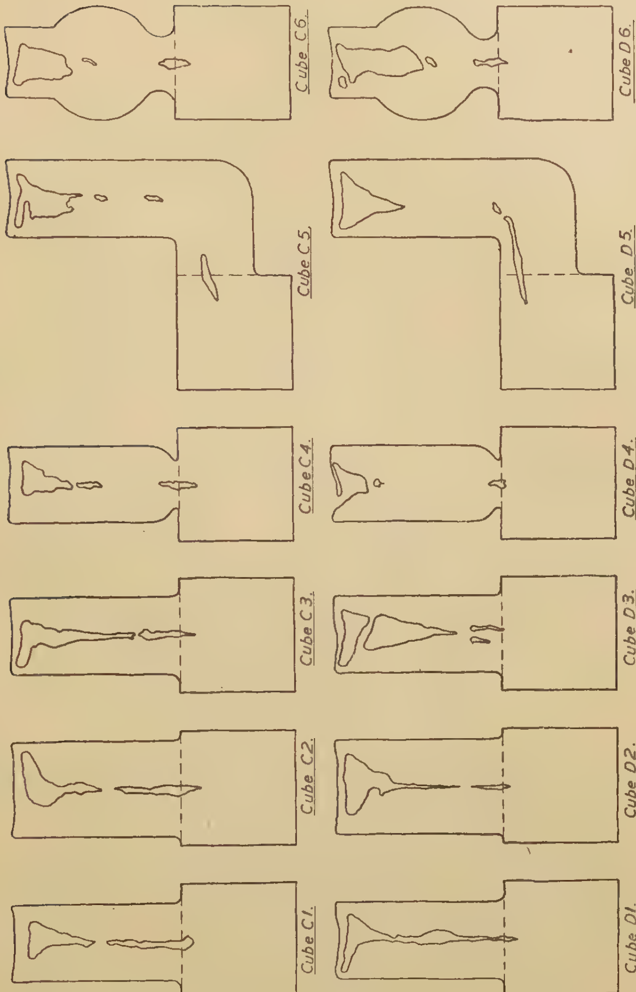

FIG. 3B (*see* Fig. 3A).—Sections of 6-in. Cubes, group 1.

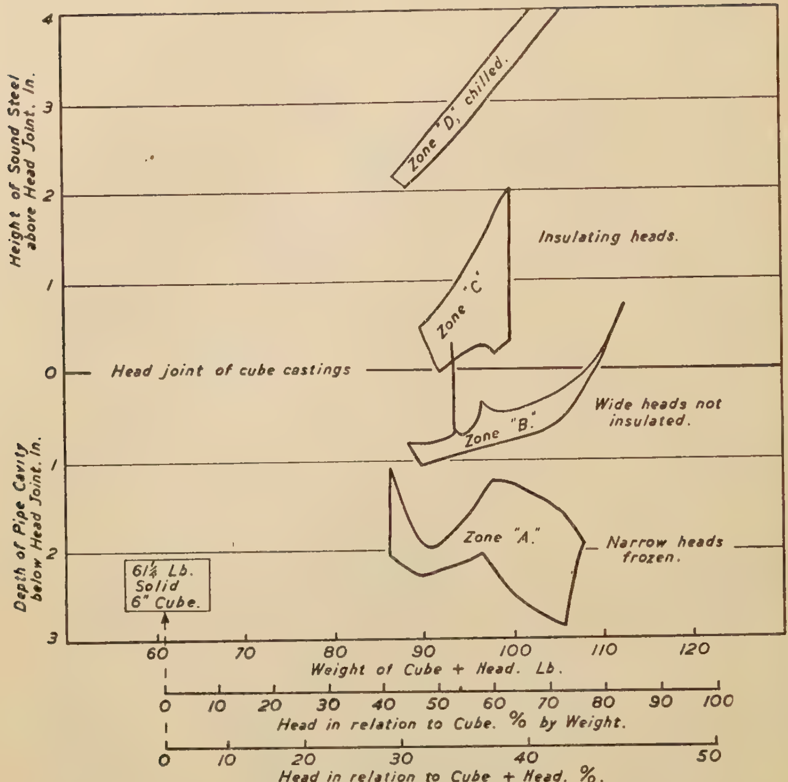

Fig. 4.—Diagrammatic Grouping of 6-in. Cubes.

Many of the results are summarised graphically in Fig. 4, where the length of sound material above the cube, or the depth of piping into the cube, is plotted against :

(*A*) The weight in pounds of the cube and head combined,
(*B*) the percentage weight of head in relation to a solid 6-in. cast-steel cube, weighing 61¼ lb., and
(*C*) the percentage weight of head in relation to the weight of the cube and head combined.

It will be seen that the first series of cube castings, referred to above as group 1, fall into zone *A*. These have pipe cavities varying in depth from 1 in. to 2 in. All these castings were made in dry sand, and no feeding compound was used in the heads after casting.

The chemical compositions of the steels used for different cubes are stated in Table VII. Other particulars relating to the cubes

in groups 2 to 7 are given in the Tables and diagrams referred to below under each group.

TABLE VII.—*Chemical Composition of the Steels Used for the Cubes.*

Cube Nos. and Dimensions.	Process.	Analysis. C. %.	Si. %.	Mn. %.	S. %.	P. %.
*A*1 to *A*6 (6 in.)	Basic electric	0·19	0·41	0·90	0·018	0·030
*A*7 to *A*12 (6 in.)	Basic electric	0·225	0·42	0·79	0·019	0·032
*A*13 and *A*14 (6 in.)	Basic electric	0·205	0·45	0·76	0·036	0·029
*A*15 to *A*26 (6 in.)	Basic electric	0·215	0·35	0·86	0·015	0·032
*A*27 to *A*30 (6 in.)	Basic electric	0·23	0·36	0·79	0·020	0·036
*A*31 and *A*32 (6 in.)	Acid electric	0·17	0·28	0·61	0·036	0·044
*A*33 (4 in.) and *A*34 (9 in.)	Basic electric	0·21	0·32	0·74	0·036	0·039
*B*1 to *B*6 (6 in.)	Basic electric	0·25	0·19	1·14	0·017	0·022
*B*7 to *B*10 (6 in.)	Basic electric	0·19	0·08	0·69	0·027	0·022
*B*11 and *B*12 (6 in.)	High-frequency	0·23	0·16	0·57	...	...
*B*13 to *B*16 (6 in.)	Basic electric	0·27	0·25	1·41	0·027	0·035
*B*17 (4 in.) and *B*18 (9 in.)	Basic electric	0·21	0·11	0·74	0·021	0·027
*C*1 to *C*6 (6 in.)	Basic electric	0·19	0·29	0·74	0·019	0·023
*C*7 and *C*8 (6 in.)	Basic electric	0·19	0·48	0·75	0·010	0·035
*C*9 to *C*12 (6 in.)	Basic electric	0·23	0·40	0·65	0·018	0·040
*C*13 to *C*16 (6 in.)	Basic electric	0·19	0·52	0·70	0·010	0·041
*C*17 and *C*18 (6 in.)	Basic electric	0·20	0·35	0·67	0·026	0·040
*C*19 (9 in.) and *C*20 (4 in.)	Basic electric	0·20	0·56	0·67	0·028	0·028
*C*21 (9 in.) and *C*22 (4 in.)	Basic electric	0·21	0·39	0·65	0·020	0·026
*D*1 to *D*6 (6 in.)	Basic electric	0·25	0·48	0·79	0·036	0·035
*D*7 and *D*8(2) (6 in.)	Basic electric	0·26	0·32	0·68	0·035	0·017
*D*8 and *D*9 (6 in.)	Basic electric	0·29	0·35	0·75	0·024	0·013
*D*10 (4 in.) and *D*11 (9 in.)	Basic electric	0·21	0·24	0·58	0·028	0·034
*D*12 (4 in.) and *D*13 (9 in.)	Basic electric	0·28	0·27	0·74	0·024	0·037
*E*1 to *E*3 (6 in.)	Robert converter (acid)	0·21	0·40	0·87	0·030	0·035
*F*1 to *F*4 (6 in.)	Acid open-hearth	0·23	0·37	1·05	0·040	0·037

Group 2.—6-*in. Cubes with Wider Heads of Different Shapes.*

Apart from the heads, the conditions of casting these cubes were the same as for group 1. Table VIII. records the type of head, head sand, the percentage weight of the head and the soundness, the apparent casting temperature and the rate of pouring. The results of sectioning these cubes are shown in Fig. 5 and fall into zone *B* of Fig. 4. They confirm the deduction from group 1 that efficient feeding is promoted by increasing the cross-sectional area of the head.

Group 3.—6-*in. Cubes with Ingot-Type Heads.*

In view of the success normally obtained by the use of ingot heads on the tops of steel ingots, the Sub-Committee decided to cast some cubes with this type of head. The dimensions of head

TABLE VIII.—*Data for 6-in. Cubes, Group 2.*

No. of Cube.	Type of Head.	Head Sand.	Feeding Compound Used?	Weight of Head. %.	Sawn Sections.		Apparent Casting Temp. ° C.	Casting Time. Sec.	Casting Rate. Lb. per sec.	Remarks.
					Soundness in Head. In.	Pipe Cavity in Cube. In.				
*A*7	5 in. dia., lower 2 in. tapering to 6 in. dia. (6 in. deep)	Dry	No	47·0	Nil	1·06	1430–1425	11	8·2	
*A*8		Dry	No	46·2	Nil	0·835		9	10·0	
*A*9	6 in. dia., 6 in. deep	Dry	No	67·5	Nil	0·395		12	8·5	
*A*10		Dry	No	69·0	Nil	0·665		12	8·6	
*C*7	5 in. dia., 5 in. deep	Dry	No	...	Nil	0·67	...	...	...	Cast direct down the head
*C*8	5 in. dia., 8 in. deep	Dry	No	...	Nil	1·25	...	...	...	
*D*7	6⅞ in. dia. (necked), 5⅞ in. deep	Skin-dried synthetic	No	76·5	Just sound		...	...	...	

adopted were as follows: 4 in. square at the top, tapering to 6 in. at the bottom where the head joins the 6-in. cube casting, the height of the heads being 6 in. or 9 in. Details of the cubes are given in Table IX., and their sections are shown in Fig. 6. When a dry-sand mould and head are used, without feeding compound, the ingot type of head appears to be the most economical for feeding a 6-in. cube. Cubes *A*13 and *A*14 were cast very slowly, a factor which probably improved the results. Cube *A*15, with a dry-sand head but no feeding compound, and cube *A*16, with a dry-sand head and an addition of feeding compound, were made to compare their results with those obtained from cubes in group 5.

Group 4.—6-*in. Cubes with External Chills.*

The use of external chills in controlling the differential cooling of castings is well known. In order to demonstrate their effect in the solidification of 6-in. cubes, a number of castings were made, of which particulars are given in Table X., the corresponding sections, showing the positions of the chills, being reproduced in Fig. 7. As was expected, the height of solid casting increases with the amount of chilling taking place. The feeding in cube *B*8, of which the head was lined with insulating mixture and which had one chill, was substantially less than that in *B*9, where there was no insulation in the head but two chills were used. The results for the chilled cubes fall into zone *D* of Fig. 4.

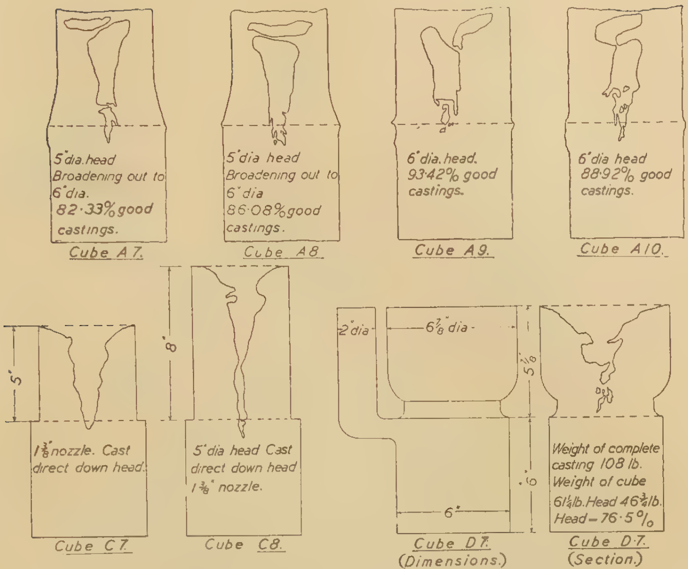

FIG. 5.—Sections of 6-in. Cubes, group 2.

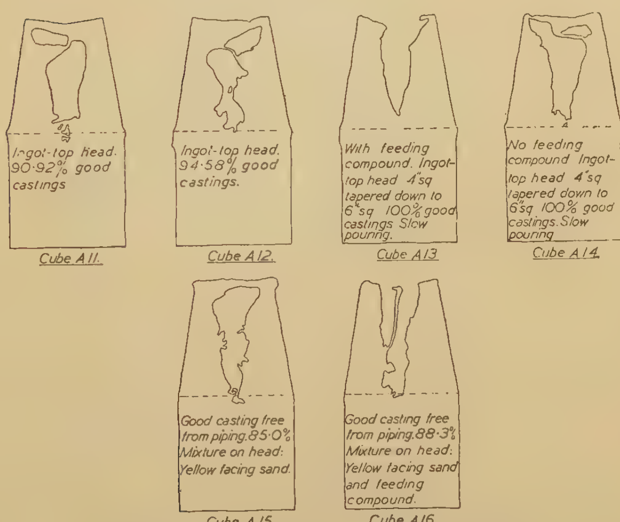

FIG. 6.—Sections of 6-in. Cubes, group 3.

TABLE IX.—*Data for 6-in. Cubes, Group 3.*

No. of Cube.	Type of Head.	Head Sand.	Feeding Compound Used?	Weight of Head. %.	Sawn Sections. Soundness in Head. In.	Sawn Sections. Pipe Cavity in Cube. In.	Apparent Casting Temp. ° C.	Casting Time. Sec.	Casting Rate. Lb. per sec.	Remarks.
*A*11	Ingot type, 6 in. deep	Dry	No	57·7	Nil	0·545	1430–1425	9½	10·2	
*A*12	Ingot type, 6 in. deep	Dry	No	57·7	Nil	0·325	1430–1425	10½	9·2	
*A*13	Ingot type, 6 in. deep	Dry	Yes	54·4	½	Sound	1450	17	5·55	Hot steel poured more slowly than *A*11 and *A*12. Gave less piping.
*A*14	Ingot type, 6 in. deep	Dry	No	53·6	Just sound	Just sound	1450	14	6·72	Hot steel poured more slowly than *A*11 and *A*12. Gave less piping.
*A*15	Ingot type, 6 in. deep	Dry	No	57·7	Nil	0·90	1435–1425	8	12·1	
*A*16	Ingot type, 6 in. deep	Dry	Yes	55·3	Nil	0·70	1435–1425	8	11·9	
*A*31	Ingot type, in. deep	Dry	No	79·4	Nil	0·375	...	9	12·2	Gated as usual. Acid electric steel.
*F*1	Ingot type, 6 in. deep	Dry	No	58·5	Nil	1·16	1425	9	10·8	
*F*2	Ingot type, 6 in. deep	Dry	No	58·0	Nil	0·88	1425	11	8·9	

TABLE X.—*Data for 6-in. Cubes, Group 4.*

No. of Cube.	Type of Head.	Head Sand.	Feeding Compound Used?	Weight of Head. %.	Sawn Sections. Soundness in Head. In.	Sawn Sections. Pipe Cavity in Cube. In.	Apparent Casting Temp. ° C.	Casting Time. Sec.	Casting Rate. Lb. per sec.	Remarks.
*B*8	4 in. dia., 12 in. deep	Insulating (sawdust and coaldust)	Yes	63·6	0·35	Sound	1435	6	16·7	Chilled at bottom of cube.
*B*9	5 in. dia., 8½ in. deep	Dry	No	70·0	4·50	Sound	1430	7	14·9	Chilled on three sides. Half chill at runner side. No chill at bottom. Run at bottom of side of cube.
*B*10	4 in. dia., 8 in. deep	Dry	Yes	42·0	2·25	Sound	1435	6	14·5	Chilled at bottom and sides of cube.

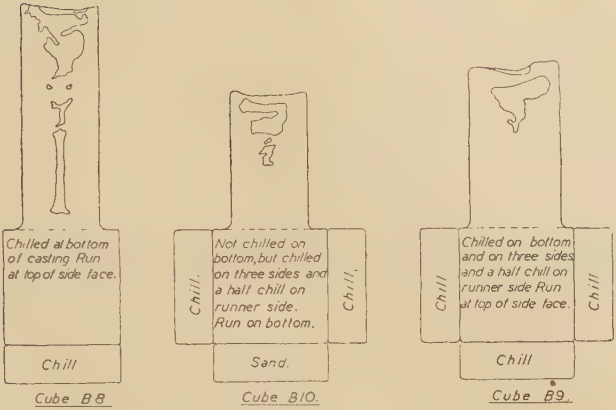

FIG. 7.—Sections of 6-in. Cubes, group 4.

Group 5.—6-*in. Cubes, Heads Lined with Insulating Mixtures.*

The data regarding the cubes in this series will be found in Table XI., and the sections are illustrated in Figs. 8 to 11. Two

FIG. 8 (*see also* Figs. 9, 10 and 11).—Sections of 6-in. Cubes, group 5.

cubes, *B*11 and *B*12 (Fig. 8), were cast from a high-frequency furnace charge, incorporating the use of a wide shallow head lined

TABLE XI.—*Data for 6-in. Cubes, Group 5.*

No. of Cube.	Type of Head.	Head Sand.	Feeding Compound Used?	Weight of Head. %.	Sawn Sections. Soundness in Head. In.	Sawn Sections. Pipe Cavity in Cube. In.	Apparent Casting Temp. ° C.	Casting Time. Sec.	Casting Rate. Lb. per Sec.	Remarks.
*A*17	Ingot type, 6 in. deep	Insulating (sawdust and coaldust)	No	52·7	1·0	Sound	1435–1425	6	15·6	
*A*18			Yes	57·7	1·0	Sound		8	12·1	
*A*19		Insulating (chopped straw and coaldust)	No	59·3	0·2	Sound		9	10·8	
*A*20			Yes	58·4	0·3	Sound		7	13·9	
*A*21		Insulating (coaldust and aluminium powder)	No	55·3	0·25	Sound		8	11·9	
*A*22			Yes	52·0	Just sound			11	8·95	
*A*23		Dry	No	57·7	Nil	0·80		$7\frac{1}{2}$	12·9	Risers painted with aluminium-coaldust paint.
*A*24			Yes	55·3	Nil	0·70		6	15·9	
*A*25		Insulating (sawdust and coaldust)	No	57·7	0·30	Sound		5	19·3	
*A*26			Yes	56·0	0·65	Sound		6	15·9	
*A*27	5 in. dia., lower 2 in. tapering to 6 in. dia. (6 in. deep)	Insulating, standard	No	43·7	Nil	0·25	1450	$10\frac{1}{2}$	8·4	Compare with *A*7 and *A*8.
*A*28	5 in. dia., lower 2 in. tapering to 6 in. dia. (9 in. deep)		No	72·4	0·75	Sound		$13\frac{1}{2}$	7·8	
*A*29	Ingot type, 6 in. deep		No	56·0	0·50	Sound		10	9·6	Compare with *A*11 to *A*26.
*A*30	Ingot type, 9 in. deep	Insulating (sawdust and coaldust)	No	86·0	1·875	Sound	1505–1500	14	8·1	High-frequency steel. Cast hot and slowly
*B*11	6 in. dia., $3\frac{3}{4}$ in. deep		Yes	26·3	0·50	Sound		20	4·48	
*B*12	5·3 in. sq., $4\frac{1}{2}$ in. deep		Yes	59·3	1·875	Sound		30	3·26	
*B*13	6 in. dia., $8\frac{1}{4}$ in. deep	Insulating, standard	No	111·0	0·50	Sound	1420	18	8·1	
*B*14	5·3 in. sq., $8\frac{1}{2}$ in. deep		No	109·0	1·50	Sound		15	10·2	
*B*15	5·3 in. sq., 6 in. deep		No	78·4	Nil	0·25		14	9·5	
*B*16	6 in. dia., 6 in. deep		No	80·0	0·50	Sound	1500 (corrected)	16	8·2	
*C*17	4 in. dia., $8\frac{3}{4}$ in. deep		Yes	45·45	1·00	Sound		14	6·85	
*D*8	6 in. dia. (necked), $5\frac{1}{4}$ in. deep		No	70·0	Nil	0·50	...	4	26·0	
*D*9	6 in. dia. (necked), $8\frac{1}{2}$ in. deep		No	128·0	3·0	Sound	...	5	28·0	
*E*1	Ingot type, 6 in. deep		No	63·2	Nil	0·375	1480	28	3·57	Converter steel. Cast slowly.
*E*2	Ingot type, 9 in. deep		No	92·6	0·562	Sound	1475	31	3·80	
*E*3	5 in. dia., 8 in. deep		No	66·4	Nil	0·562		30	3·40	Acid O.H. steels.
*F*3	Ingot type, 6 in. deep		No	60·0	Nil	0·15	1425	11	8·9	
*F*4			No	62·0	Nil	0·38		8	12·4	
*C*13	4 in. dia., 6 in. deep		No	25·9	Nil	1·50	1500 (corrected)	9	8·1	Compare with *C*1, *A*1 and *B*1.
*C*14	4 in. dia., 9 in. deep		No	42·5	Nil	0·75		10	8·4	
*C*15	5 in. dia., 6 in. deep		No	45·8	Nil	0·50		...	...	
*C*16	5 in. dia., 9 in. deep		No	68·0	Nil	1·50		...	...	

with insulating sand, a low rate of casting, and the addition of feeding compound. The results were very good and showed that under these conditions a sound cube could be obtained by using a head weighing less than 50% of the weight of the cube, without resorting to the use of chills. The heads used were of the same sectional area, one being round and 6 in. in dia., the other square and of 5·3-in. side. The results did not, however, point to a definite advantage of one shape of head over the other. It was then decided to investigate the advantage *per se* as a single factor of retaining the heat in the head by using a sand mixture of low thermal conductivity, all other factors being kept constant.

The mixture selected for use as an initial standard was made up as follows:

	Weight-%.
Arnold's silica sand No. 52	65
Sawdust through $\frac{1}{8}$-in. square mesh	10
Bituminous coaldust	10
Fireclay	10
Bentonite	5

The moisture content of the mixture as milled was 13–15%, and the mixture was thoroughly dried at 200–250° C. before casting. In order to avoid damage to the weak insulating sand during the closing of the moulds, the bottom 1 in. of the riser was rammed up in the usual dry sand.

TABLE XII.—*Insulating Materials Used as Facing Sands on Experimental Heads.*

Mixture.		Mixture.	
A	72·5% Old sand 10·0% Sawdust 10·0% Coaldust 7·5% Fireclay	*B*(2)	75·0% Old sand 5·0% Chopped straw 10·0% Coaldust 10·0% Fireclay
B(1)	72·5% Old sand 10·0% Chopped straw 10·0% Coaldust 7·5% Fireclay	*C*	72·5% Old sand 10·0% Aluminium powder 10·0% Coaldust 7·5% Fireclay

Aluminium/Coaldust Paint.

This paint was prepared as follows:

210 g. Aluminium powder
210 g. Coaldust
15 g. Bentonite
} mixed thoroughly.

720 c.c. Water added to give a cream.

Following mixture added to above:

50 g. Core cream.
25 g. Sulphite lye.
120 c.c. Water.

This paint has a thick consistency, and dries quite hard.

Fig. 8 also shows the sections of four cubes, *B*13 to *B*16, cast with this type of insulating sand round the heads. Two of the heads were 6 in. in dia., and the other two were square of 5·3-in. side. Again there were no particular indications of the superiority of either shape when the cross-sectional areas were the same. Other cubes were made for comparison purposes with insulating mixtures differing from the above; their constitution is given in Table XII. In mixture *B*1 the sawdust of mixture *A* was replaced by an equal amount of chopped straw. This amount, however, was found too much for the mixture to be easily workable, and it was consequently cut down in mixture *B*2. Some of the heads in these experiments

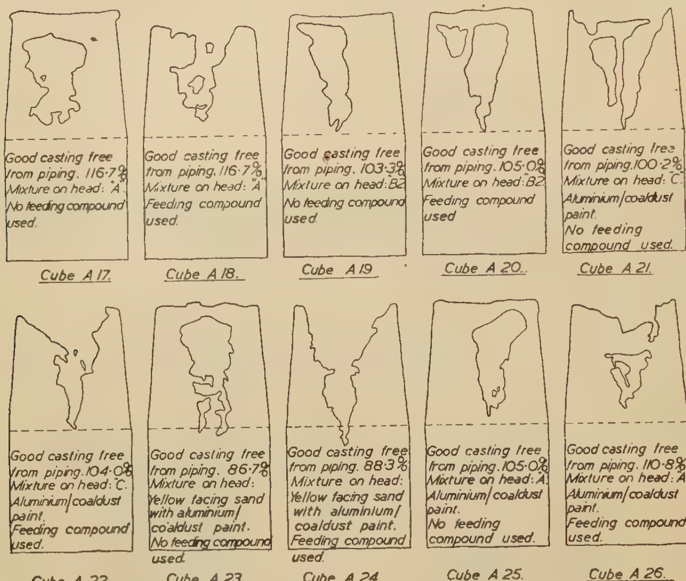

FIG. 9 (*see also* Figs. 8, 10 and 11).—Sections of 6-in. Cubes, group 5.

were painted with a paint containing powdered aluminium and coaldust, and in several instances pairs of cubes were made, feeding compound being added to one and not to the other. Sections of a number of these cubes are illustrated in Fig. 9. Sections of other cubes are reproduced in Fig. 10. The results for cubes *A*27 to *A*30 confirm that it is advantageous to use the ingot type of head. For cubes *C*13 to *C*16 no feeding compound was used. Cubes *C*13 and *C*15 had heads 6 in. high and 4 in. and 5 in. in dia., respectively, whilst *C*14 and *C*16 had corresponding heads but 9 in. high. For cubes *C*17 and *C*18 (duplicates) the use of standard insulating sand and feeding compound was combined.

A study of the results obtained in this group shows that the

mixture containing sawdust and coaldust was the most effective of those tried, and that the use of the aluminium/coaldust paint did not appreciably increase the soundness of the casting. From this point of view the use of insulating material in the head is of greater value than that of a feeding compound.

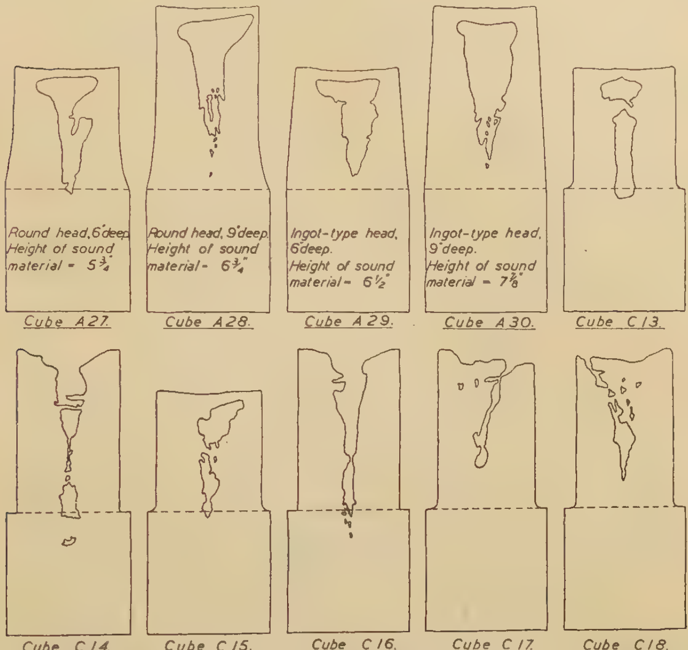

FIG. 10 (*see also* Figs. 8, 9 and 11).—Sections of 6-in. Cubes, group 5.

At this stage it was decided to cast a few cubes in steel made by the side-blown converter and by the open-hearth process, to compare the results with those obtained from similar cubes in basic electric steel. Three cubes, $E1$ to $E3$, were made by a fifth foundry in acid converter steel, with ingot-type and round heads, insulating sand being used in all cases. Four other cubes, $F1$ to $F4$, were made by the acid open-hearth process in a sixth foundry. The heads on $F1$ and $F2$ were of the ingot type in dry sand, and details of these are given in Table IX. (group 3), whilst those of cubes $F3$ and $F4$ were of the same type in the standard insulating sand mixture. Sections of these cubes are shown in Fig. 11. A comparison of the results given by similar cubes made by the three steel processes is given in Table XIII. They tend to show that steel made by the basic electric process suffers rather less piping

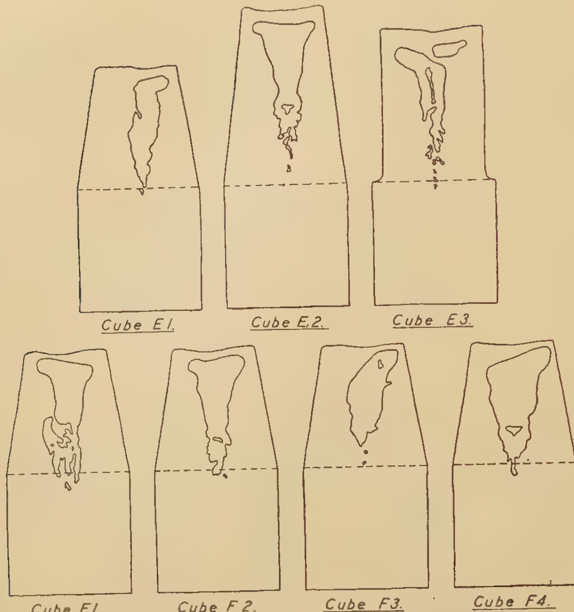

FIG. 11 (*see also* Figs. 8, 9 and 10).—Sections of 6-in. Cubes, group 5.

TABLE XIII.—*Effect of Steelmaking Process on Soundness of 6-in. Cube Castings.*

Ingot-type heads, 6 in. deep; no feeding compound.

Process.	Head Sand.	Cube No.	Soundness above Head. In.	Depth of Pipe in Cube. In.	Apparent Casting Temp. ° C.	Casting Rate. Lb. per sec.
Basic electric	Dry	*A*11	Nil	0·54	1430–1420	10·2
		*A*12	Nil	0·33	1430–1420	9·2
		*A*14	Nil	Just sound	1450	6·72
		*A*15	Nil	0·9	1435	12·1
		*A*23	Nil	0·8	1430	12·9
				Av'ge 0·51		
Acid O.H.	Dry	*F*1	Nil	1·16	1425	10·8
		*F*2	Nil	0·88	1425	8·9
				Av'ge 1·02		
Basic electric	Insulating	*A*17	1·0	Sound	1435	15·6
		*A*25	0·3	Sound	1425	19·3
		*A*29	0·5	Sound	1450	9·6
			Av'ge 0·6			
Acid converter	Insulating	*E*1	Nil	0·375	1480	3·57
Acid O.H.	Insulating	*F*3	Nil	0·15	1425	8·9
		*F*4	Nil	0·38	1425	12·4
				Av'ge 0·27		

than that made by the two acid processes under the conditions of these particular experiments.

The four sections reproduced in Fig. 12 are interesting as illustrating the effect of the use of insulating sand in the heads put on commercial steel castings. Heads Nos. 1 and 2 were in dry sand, whilst the similar heads Nos. 3 and 4 were lined with an insulating sand mixture. Feeding compound was used in all four cases. The increased feeding with the insulated heads is clearly seen.

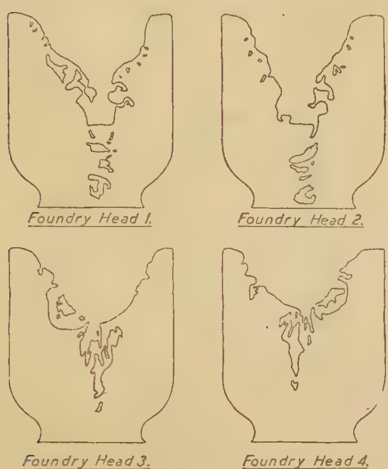

Fig. 12.—Sections of Foundry Heads.

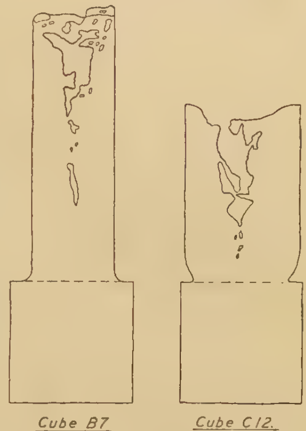

Fig. 13.—Sections of 6-in. Cubes, group 6.

Group 6.—6-*in. Cubes Run and Gated in Different Ways.*

A number of cube castings were made under conditions which varied from those taken as standard for the purpose of the present investigation. Particulars of these are given in Table XIV. Sections of cubes *B*7 and *C*12 are shown in Fig. 13. Cube *A*32 (not illustrated), with an ingot type of head 9 in. high in dry sand, was run at the bottom of the side. This gave results inferior to those of cube *A*31, which was otherwise similar but was run in the usual position at the top of the side of the cube. Both *A*31 and *A*32 were cast in acid electric steel. Cube *B*7 had a 4-in. dia. head 12 in. high in insulating sand, and feeding compound was added. This also was run at the bottom of the side and gave a sound cube. It was made to compare its results with those of cubes cast with chills. Cube *C*12 had a 5-in. dia. head in dry sand, necked at its base. It was run into the side of the head and feeding compound was used. The results are similar to those obtained from a cube run in the usual way with an insulated head and without the use of feeding compound.

TABLE XIV.—*Data for 6-in. Cubes, Group 6.*

No. of Cube.	Type of Head.	Head Sand.	Feeding Compound Used?	Weight of Head. %.	Sawn Sections. Soundness in Head. In.	Sawn Sections. Pipe Cavity in Cube. In.	Apparent Casting Temp. ° C.	Casting Time. Sec.	Casting Rate. Lb. per sec.	Remarks.
*A*32	Ingot type, 9 in. deep	Dry	No	79·4	Nil	0·750	...	10½	10·5	Gated into bottom of side of the cube. Acid electric steel.
*B*7	4 in. dia., 12 in. deep	Insulating (sawdust and coaldust)	Yes	63·6	2·00	Sound	1440	6	16·7	Run at bottom of side of cube.
*C*12	5 in. dia., 8 in. deep	Dry	Yes	80·0	0·90	Sound	...	...	...	Runner into the side of the head.
*D*8(2)	Two cubes with one central head 8 in. in dia. by 10 in. deep	Insulating, standard	No	92·8 for each cube	5¼	Sound	...	...	...	*See* Fig. 14.

TABLE XV.—*Data for 4-in. and 9-in. Cubes, Group 7.*

No. of Cube.	Side of Cube. In.	Type of Head.	Head Sand.	Weight of Head. %.	Sawn Sections. Soundness in Head. In.	Sawn Sections. Pipe Cavity in Cube. In.	Apparent Casting Temp. ° C.	Casting Time. Sec.	Casting Rate. Lb. per sec.
*A*33	4	Ingot type, 2·67 in. sq. at top, tapering to 4 in. sq. at the bottom, 4 in. deep	Dry	68·6	Nil	23/32	1410	9·2	3·3
*D*10	4			54·7	Nil	3/8	...	6	4·7
*B*17	4		Insulating, standard	60·3	½	Sound	1440	6	4·8
*C*20	4			76·7	Nil	½	1520 (corrected)	6	5·3
*C*22	4	3 in. dia., 5 in. deep	Insulating, standard and feeding compound	60·3	Nil	1·0	1525 (corrected)	5	5·8
*D*12	4	Ingot type, 2·67 in. sq. at top, tapering to 4 in. sq. at the bottom, 4 in. deep	Insulating, standard	49·2	Nil	¼	...	4	6·75
*A*34	9	6 in. sq. at top, tapering to 9 in. sq. at the bottom, 9 in. deep	Dry	61·6	Nil	31/32	1410	16·6	20·0
*D*11	9			61·5	Nil	¾	...	15	22·1
*B*18	9		Insulating, standard	64·7	1¼	Sound	1440	24	14·1
*C*19	9			63·9	Nil	½	1520 (corrected)	35	9·6
*C*21	9	6 in. dia., 11 in. deep	Insulating, standard and feeding compound	36·0	1·0	Nil	1525 (corrected)	32	9·1
*D*13	9	6 in. sq. at top, tapering to 9 in. sq. at the bottom, 9 in. deep	Insulating, standard	59·6	Nil	1¼	...	17	19·3

The results of another experiment are illustrated in Fig. 14. Two cubes, *D*8(2), were run from a single large, central reservoir and both were sound. The proportion of head for each cube was 92·8%, but the cross-section of the cast cubes suggests that three or four sound cubes could have been run from the same head.

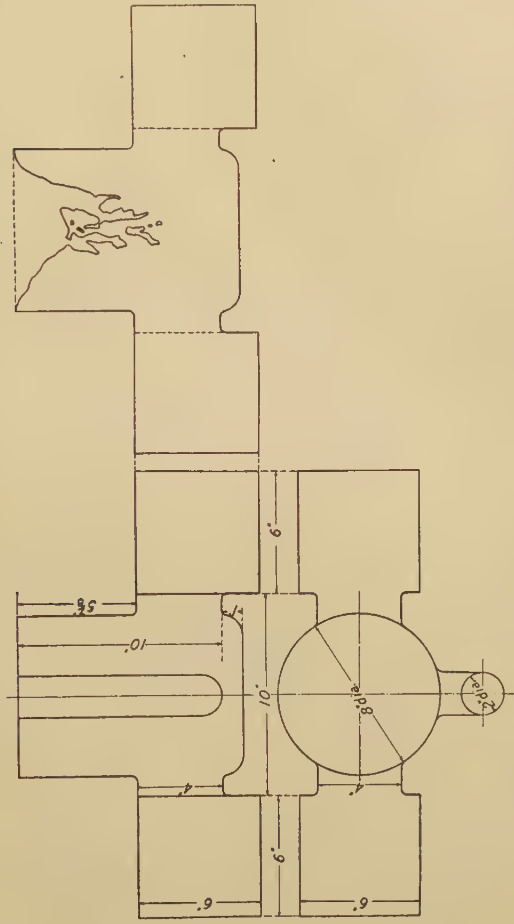

Fig. 14.—Design and Section of Two 6-in. Cubes Cast with One Head.

Group 7.—*Comparison of* 4-*in. and* 9-*in. with* 6-*in. Cubes.*

It was felt desirable to discover how the results obtained from 6-in. cubes would be modified if smaller or larger cubes were made on similar lines, and a number of cubes with 4-in. and 9-in. sides were made in four different foundries, some with insulating heads

and some with the usual dry-sand head. The bulk of these were fitted with heads of the ingot type, but two of them were made with round heads. Particulars of these cubes are given in Table XV. and sections are illustrated in Fig. 15.

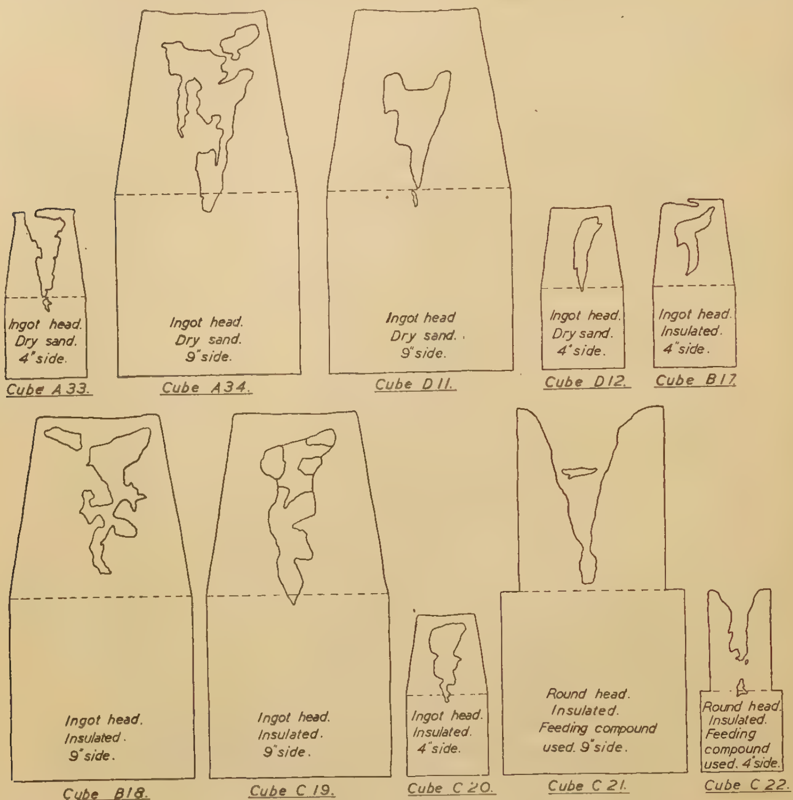

FIG. 15.—Sections of 4-in. and 9-in. Cubes, group 7.

The results obtained, which did not, however, aim at a complete duplication of all the work that had been done on 6-in. cubes, did not show that there was any significant difference in the feeding characteristics of heads of the same geometrical shape, when used on cubes of 4-in., 6-in. or 9-in. side. The advantage in feeding efficiency of insulating heads over dry-sand heads, found with 6-in. cubes, was again evident in the results from 4-in. and 9-in cubes.

General Survey.

The Members of the Sub-Committee realise that the experiments so far carried out cover only part of the field to be in-

vestigated before the rules relating to running and feeding, referred to in the Sub-Committee's terms of reference, can be enunciated so as to be of general application. The results so far obtained, however, are, in the main, consistent and enable the Sub-Committee to summarise their observations as follows.

In making 6-in. cubes, to obtain the most economical use of the steel, the area of the bottom of the feeder head should be as large as the area of the upper surface of the casting. The Sub-Committee realise that in making commercial foundry castings a head covering the entire upper surface of a casting is seldom possible or practicable, so that horizontal feeding becomes necessary. This aspect of the problem received attention in further experiments reported in Part 1 (*b*) of this Report.

Enlarging the bottom of the head where it joins the casting gives better results, with a casting such as a cube, than constricting the head at its junction with the casting. The shape of head normally used to feed steel ingots appears to be very effective.

It is advantageous to secure a differential rate of cooling in the casting and in the head. This has been obtained :

(1) By lining the head with a layer, $1\frac{1}{2}$ in. thick, of material having a low thermal conductivity, and thus reducing the rate of cooling in the head, and

(2) by the use of external chills, to increase the rate of cooling in the body of the casting.

In this connection it is to be noted that the Sub-Committee have not investigated the effect of internal chills in the form of horse-nails and other metallic shapes. They consider that insulating heads can be used with advantage much more widely than at present, so long as certain practical precautions are taken; for example, it must be remembered that insulating sand mixtures are weaker than ordinary dry sand, and also that insulated heads must not be heated to such a high temperature during the drying process that the coaldust is burnt out or the sawdust charred. The use of insulating sand heads appears to be at least as efficient as the use of feeding compound.

There is some suggestion in the results that hot steel poured slowly gives less piping than cold steel poured quickly.

Most of the cubes examined were made by the basic electric-arc process. In some cases variations in feeding have been found in comparing the cubes made in different foundries. This may be connected with the degree of deoxidation of the steel concerned. A study of the steel analyses reveals that steels with a higher silicon content give deeper piping than those containing less silicon. To compare with these cubes, a number of others were made in acid converter and in acid open-hearth steels. Under the conditions of the present experiments, the steel made by both acid processes gave somewhat greater piping than basic electric-arc steel.

The Sub-Committee wish to make it plain that, in such investigations as are described above, as many conditions as possible must be made standard and easy of attainment. Some of them therefore do not correspond to actual foundry practice, in which the efficiency of feeding is not the sole aim and may not even be the first consideration. There are many types of steel castings that are regularly made in substantial quantities by different foundries where successful results are consistently obtained by methods which, whilst standardised in a particular foundry, vary considerably between one foundry and another. This is usually due to exigencies of local equipment and conditions, and it does not follow therefore that a method successful in one foundry will be equally successful in another, where the materials available and the foundry conditions are different. Some of the factors involved in selecting practical procedure in a given foundry are very briefly mentioned below.

The orientation of the pattern in the mould is influenced by the incidence of machined surfaces, the size of mould box available, the ease with which cores may be supported, the disposition of the thicknesses which demand feeding, and so on. The orientation selected is based on a compromise between the varying demands, and to obtain efficient feeding it may be necessary, for example, to put at the top of the mould surfaces which must afterwards be machined.

The position of the ingate may be influenced, or even decided, by the properties of the mould and core materials used. It may, for example, have to be placed in a particular position to avoid damage to the mould or cores during pouring, especially if only comparatively large ladles are available. This position, however, may not give the best distribution of temperature to promote efficient feeding or minimise the danger of cracks.

The provision of internal or external chills to obtain differential cooling in the mould, and of brackets or fillets to prevent hot tears, may interfere with both the running of the casting and the feeding of the other portions.

These and similar factors affecting foundry procedure constitute the reasons why, as mentioned earlier in the present Report, foundry practice in moulding, gating and feeding is based on practical experience and works tradition. When castings of new design are called for, the cutting-up and examination of successive initial samples before satisfactory results are obtained are expensive and, what is often more important, waste a good deal of time.

Whilst it is evident that the investigation of running and feeding methods will not of itself solve the problems encountered in every foundry, there is no doubt that the establishment in this field of definite principles which could be applied to steel castings generally would be of considerable assistance and benefit to all technical steel-foundry staff, since it would help to remove some of the uncertainties which constantly confront them.

(b) *Castings* 6-*in. Square by* 12 *in.*, 18 *in. and* 24 *in. Long.*

The work reported in Part 1 (*a*) showed that in order to use the fluid steel in the most economical manner to produce sound 6-in. cube castings in sand, without the aid of (*a*) external chilling of the mould as can be done in the case of ingots, (*b*) insulation of the head, (*c*) the use of feeding compounds, and (*d*) "after-feeding" with liquid steel, the "ingot-type" head (4-in. sq. at the top tapering to 6 in. square at the lower end, by 6 in. deep) was not quite sufficient to give a sound cube, but that when the depth of the head was increased to 9 in. a sound cube casting was secured.

To those accustomed to ingot practice it may seem extraordinary that in casting a 6-in. cube (216 cu. in.) an ideally shaped head of 150 cu. in. (70%) was proved to be inadequate, and that it was found necessary to employ a head of 225 cu. in. (over 100%) to

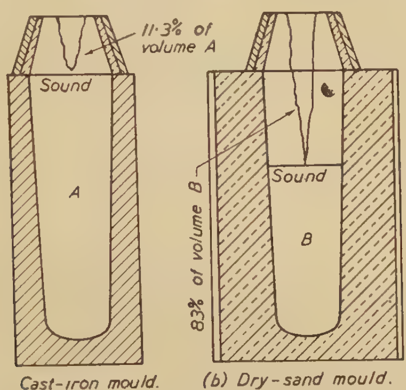

Fig. 16.—Comparison of Shrinkage Cavities in 11-in. Square Ingots.

ensure the absence of a piping cavity at the head joint when casting into dry-sand moulds. However, by a comparison of Figs. 1, 21 and 23 in a recent paper on "The Influence of Turbulence upon the Structure and Properties of Steel Ingots" by Northcott,[1] confirmation was obtained, as shown in Fig. 16. An 11-in. square ingot cast into a standard ingot mould (Fig. 16 (*a*)) was sound with 11·3% of head, whereas a similar ingot cast in dry sand (Fig. (16 (*b*)) showed a deep piping cavity extending 12½ in. below the head joint. In other words, to make a sound casting 11-in. square by 2 ft. 6 in. deep in sand it would be necessary to use a feeder head of 83% of the volume of the finished casting.

While it has been shown by the experiments that, in order to obtain the most economical use of the steel, the area of the bottom of the feeder on the 6-in. cube should be as large as the area of the

[1] L. Northcott, *Journal of The Iron and Steel Institute*, 1941, No. I., p. 49 P.

upper surface of the casting, in commercial castings a head covering the entire upper surface of the casting is seldom practicable, so that horizontal feeding becomes involved. This aspect of the problem was investigated by the Sub-Committee by making a series of blocks cast in dry sand, as follows:

Group 8.—6 in. square × 12 in. long.
,, 9.—6 in. square × 18 in. long.
,, 10.—6 in. square × 24 in. long.
,, 11.—6 in. square × 24 in. long with a larger head.

These castings are classified in Table XVI. and illustrated in Fig. 17.

TABLE XVI.—*Classification of the Castings in Groups 8 to 11.*

Group.	Casting.	Ingot-Type Head.
8(*a*)	6 in. sq. × 12 in. long, cast horizontally.	4 in. sq. at the top, tapering to 6 in. sq. at the bottom, by 9 in. deep. Head placed at the runner end of the casting.
8(*a*)1		As above, but using standard insulating sand round the heads.
8(*b*)		Head as 8(*a*), but placed central.
8(*b*)1		As 8(*b*), but using standard insulating sand round the heads.
8(*c*)		Head as for 8(*a*) and 8(*b*), but on end of the block cast vertically.
8(*c*)1		As above, but using standard insulating sand round the heads.
9	6 in. sq. × 18 in. long, cast horizontally.	Head as 8(*a*) and 8(*b*).
9(1)		As above, but using insulating sand round the head.
9(1)*L*		As 9(1), plus low-grade blacklead-type of feeding compound.
10	6 in. sq. × 24 in. long, cast horizontally.	Head as 8(*a*), 8(*b*) and 9.
10(1)		As above, but using insulating sand round the heads.
11		Larger head, 5½ in. × 4 in. at top, tapering to 8 in. × 6 in., by 12 in. deep.
11(1)		As above, but using insulating sand round the head.
11(1)*L*		As 11(1), plus low-grade blacklead-type of feeding compound.
11(*b*)		As 11, using 3 in. thickness of high-temperature insulating brick round the head, but no feeding compound.
11(*x*)		As 11, but run from the opposite end to the head in error.

The chemical compositions of the steels used for the different castings are stated in Table XVII. Other particulars relating to

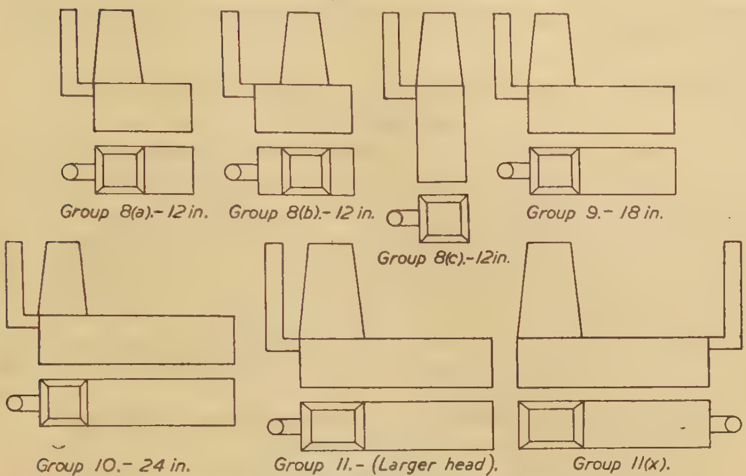

FIG. 17.—Methods of Running and Heading 6-in. Castings, groups 8 to 11.

TABLE XVII.—*Compositions of the Steels used for Groups* 8 *to* 11.

Group.	Cast-ing No.	Process.	Analysis.					Dimensions of Test Casting.
			C. %.	Si. %.	Mn. %.	S. %.	P. %.	
8(*a*)	*D*14		0·24	0·20	0·72	0·025	0·037	
	*A*36	Basic electric	0·215	0·44	0·82	0·030	0·035	
	*B*19		0·20	0·23	0·77	0·017	0·023	
	*C*23		0·19	0·36	0·67	0·014	0·032	
	*E*4	Acid converter (Robert)	0·20	0·40	0·95	...	...	
8(*b*)	*D*15		0·24	0·20	0·72	0·025	0·037	6 in. sq. × 12 in. long.
	*E*37	Basic electric	0·215	0·44	0·82	0·030	0·035	
	*B*20		0·20	0·23	0·77	0·017	0·023	
	*C*24		0·19	0·36	0·67	0·014	0·032	
	*E*5	Acid converter (Robert)	0·20	0·40	0·95	...	...	
8(*c*)	*A*35	Basic electric	0·27	0·40	0·80	0·014	0·029	6 in. sq. × 12 in. long (cast vertically).
	*A*38		0·215	0·44	0·82	0·030	0·035	
	*E*6	Acid converter (Robert)	0·20	0·40	0·95	...	...	
9	*C*25		0·21	0·42	0·84	0·015	0·033	
	*A*39	Basic electric	0·20	0·40	0·89	0·028	0·033	
	*B*21		0·20	0·23	0·84	0·016	0·014	
	*F*5	Acid O.H.	0·29	0·37	1·33	0·036	0·032	6 in. sq. × 18 in. long.
	*F*6		0·23	0·29	1·10	0·036	0·031	
	*B*22		0·20	0·23	0·84	0·016	0·014	
	*B*23	Basic electric	0·20	0·23	0·84	0·016	0·014	
10	*B*24		0·25	0·23	0·66	0·019	0·019	6 in. sq. × 24 in. long.
	*F*7	Acid O.H.	0·37	0·37	1·33	0·042	0·040	
	*B*25	Basic electric	0·25	0·23	0·66	0·010	0·019	
11	*F*9	Acid O.H.	0·33	0·43	1·42	0·036	0·046	
	*A*40		0·23	0·38	0·77	0·030	0·032	
	*B*26		0·23	0·28	0·59	0·008	0·016	6 in. sq. × 24 in. long (with larger head).
	*B*27		0·23	0·28	0·59	0·008	0·016	
	*B*28	Basic electric	0·23	0·28	0·59	0·008	0·016	
	*B*29		0·23	0·28	0·59	0·008	0·016	
	*A*41		...	...	...	...	...	

the castings in groups 8 to 11 are given in Tables XVIII. to XX. and Figs. 18 to 21.

Investigation of the Horizontal Feeding Characteristics of Steel in Dry-Sand Moulds.

In order to determine the most economical methods of feeding a casting such as a ring of which the section was 6 in. square, the Sub-Committee decided to extend the 6-in. cube casting in a horizontal direction in steps of 6 in. at a time, making castings 12 in., 18 in. and 24 in. long, respectively, and using in the first place the ingot type of head 9 in. deep, which was found to be adequate to make a sound 6-in. cube, and thus to determine the horizontal distance that such a head will feed. Unless otherwise stated, the castings were made in dry-sand moulds and heads and were run

Table XVIII.—*Results of Sectioning Castings in Group* 8.

6 in. square by 12 in. long.
8(*a*).—Cast horizontally, head at one end.
8(*b*).—Cast horizontally, central head.
8(*c*).—Cast vertically.

Casting No.	Head Sand Used.	Feeding Compound Used?	Steelmaking Process.	Casting Temp., App't. ° C.	Casting Time. Sec.	Result of Sectioning. Soundness in Head.	Result of Sectioning. Depth of Pipe Cavity into Casting.
			Group 8(a).—*Cast Horizontally. Head at one End.*				
*D*14	Dry sand	No	Basic electric	...	11	Nil	1 in. deep, plus secondary horizontal cavity 6½ in. long.
*A*36	Dry sand	No	Basic electric	1440	17	Nil	1⅞ in. pipe.
			Group 8(a)1.—*As for* 8(a), *but using Insulating Sand round the Heads.*				
*B*19	Insulating sand	No	Basic electric	1415	16	½ in.	Sound.
*C*23	Insulating sand	No	Basic electric	1520 (corrected)	10	Nil	¼-in. pipe.
*E*4	Insulating sand	No	Acid converter (Robert)	1465	26	Nil	⅜-in. pipe.
			Group 8(b).—*Cast Horizontally. Central Head.*				
*D*15	Dry sand	No	Basic arc	...	9	Nil	1-in. pipe.
*A*37	Dry sand	No	Basic arc	1440	14½	Nil	1⅛ in. pipe.
			Group 8(b)1.—*As for* 8(b), *but using Insulating Sand round the Heads.*				
*B*20	Insulating sand	No	Basic electric	1415	18	½ in.	Sound.
*C*24	Insulating sand	No	Basic electric	1520 (corrected)	12	Nil	½-in. pipe.
*E*5	Insulating sand	No	Acid converter (Robert)	1460	27	¼ in.	Sound.
			Group 8(c).—*Cast Vertically.*				
*A*35	Dry sand	No	Basic electric	1430	17	Nil	2¾-in. pipe.
*A*38	Dry sand	No	Basic electric	1440	15	Nil	3-in. pipe.
			Group 8(c)1.—*As for* 8(c), *but using Insulating Sand round the Heads.*				
*E*6	Insulating sand	No	Acid converter (Robert)	1460	28	Nil	⅝-in. pipe.

into the end of the block at the head end, and the molten steel was allowed to freeze over normally as it rose up the heads.

Group 8.—*Castings* 6 *in. Square by* 12 *in. Long.*

Head dimensions : Ingot type, 4 in. square at top, tapering to 6 in. square at bottom, by 9 in. deep.
Volume of casting : 432 cu. in.
Volume of head : 225 cu. in. = 52·4%.

(8*a*) *Cast in Dry-Sand Moulds and Dry-Sand Heads ; Head at one End* (*Cast Horizontally*).—The results of sectioning the castings are given in Table XVIII. and Fig. 18, and show that a head with

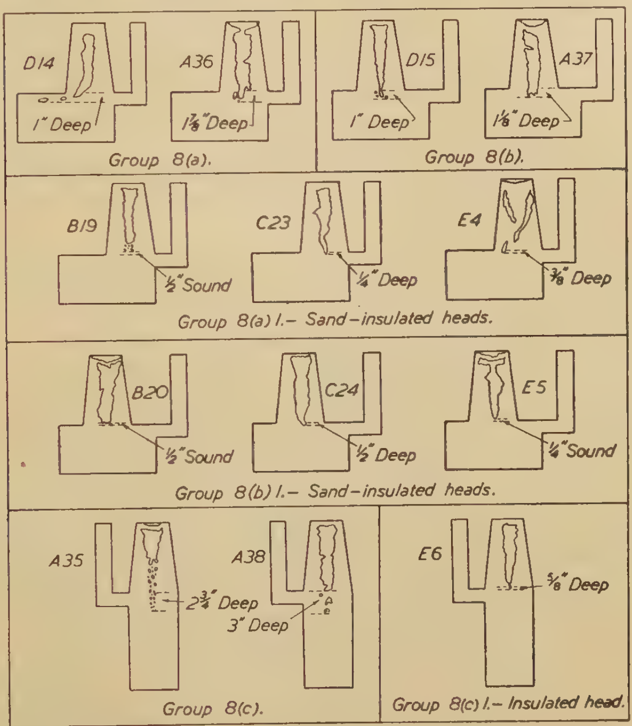

FIG. 18.—Sections of Group 8 Castings, 6 in. square by 12 in. long.

a volume of 52·4% of that of the block is insufficient. The use of insulating sand round the heads in group 8(*a*)1 reduced the depth of piping, and the castings were practically sound.

(8*b*) *With Central Head* (*Cast Horizontally*).—The use of a central head gives slightly less piping, as the steel is feeding in two directions horizontally, and again the castings in group 8(*b*)1, using insulating sand round the heads, reduced the depth of piping, and the castings

were practically sound. The ingate was apparently sufficiently near to the head to avoid the occurrence of a shrinkage cavity in front of the runner.

(8*c*) *Block Cast Vertically.*—When this block 12 in. long was cast vertically there was deeper piping, and it is interesting to notice the close comparison between these 6-in. square castings cast in sand and the 11-in. square ingot cast in sand reported by Dr. Northcott and illustrated in Fig. 16. The advantage of using insulating head sand is illustrated in group 8(*c*)1, in which casting *E*6 shows a reduction in the depth of piping of more than 2 in., so that in this casting the pipe extends only $\frac{5}{8}$ in. into the casting.

The conclusion from this group of tests is that in making a cast-steel block 6 in. square by 12 in. long the best position for the head is the central position 8(*b*) and that differential solidification obtained by the use of insulating head sand results in a sound casting with an ingot-type head 9 in. deep, having a volume of 52·4% of the volume of the block.

In commercial foundry practice, of course, the amount of head required could be reduced by the use of external chills, feeding compound and after-feeding.

Group 9.—*Castings* 6 *in. Square by* 18 *in. Long, Cast Horizontally in Dry Sand.*

Head dimensions: As for group 8.
Volume of casting: 648 cu. in.
Volume of head: 225 cu. in. ≐ 34·8%.

As it had been found that the 9-in. deep ingot-type head, which

TABLE XIX.—*Results of Sectioning Castings in Group* 9.

6 in. square by 18 in. long.

Casting No.	Head Sand Used.	Feeding Compound Used?	Steel-making Process.	Casting Temp., App't. °C.	Casting Time. Sec.	Result of Sectioning. Soundness in Head.	Result of Sectioning. Depth of Pipe Cavity in Casting.
			Group 9.—*Cast Horizontally. Head at one End.*				
*C*25	Dry sand	No	Basic arc	1520 (corrected)	25	Nil	1¾ in. deep and large secondary pipe 5½ in. long (probably gated at opposite end to head in error).
*A*39	Dry sand	No	Basic arc	1465	...	Nil	2¼-in. pipe.
*B*21	Dry sand	No	Basic arc	1437	17	Nil	2⅜-in. pipe.
*F*5	Dry sand	No	Acid O.H.	1420	16	Nil	2¼-in. pipe.
*F*6	Dry sand	No	Acid O.H.	1395	17	Nil	2-in. pipe.
			Group 9(1).—*As Group* 9, *but using Insulating Sand round the Heads.*				
*B*22	Insulating sand	No	Basic arc	1435	14	Just sound	Nil
			Group 9(1)L.—*As for Group* 9(1), *but using Feeding Compound in Addition to Insulation.*				
*B*23	Insulating sand	Yes	Basic arc	1430	14	½ in. sound	Nil

was sufficient to feed the 6-in. cube, was just capable of feeding the 12 in. long casting when insulating sand was used round the head, the same type of head was used for the 18 in. long castings, and was placed at the runner ends of the casting.

The results of sectioning are given in Table XIX. and illustrated in Fig. 19. All the castings without insulation show deeper pipes

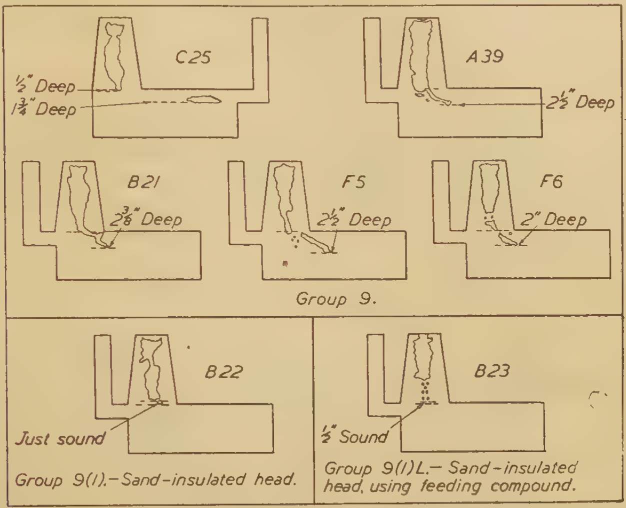

FIG. 19.—Sections of Group 9 Castings, 6 in. square by 18 in. long.

than those in the 12 in. long castings, but when insulation was used this head was just sufficient to give a sound casting (*B*22), and when a low-grade blacklead type of feeding compound was used in addition, there was ½ in. of sound steel in the head (*B*23). There is an indication that casting *C*25 was gated in error at the opposite end to the head, and this opinion was confirmed by section *A*41 (24 in. long casting) and by sulphur prints which are not reproduced in this Report.

Group 10.—*Castings* 6 *in. Square by* 24 *in. Long, Cast Horizontally in Dry Sand.*

Head dimensions : As for groups 8 and 9.
Volume of casting : 864 cu. in.
Volume of head : 225 cu. in. = 26·1%.

As the limit of the capacity of the 9-in. ingot-type head to feed in a horizontal direction had not yet been reached, the same head was used for the 24 in. long casting in group 10. The results shown in Table XX. and Fig. 20 illustrate that the limit had now been passed, and all the castings, even with insulating sand, had

a pipe cavity extending into the casting, so that 26·1% of volume of head was insufficient to feed a casting made under these conditions.

Group 11.—*Castings* 6 *in. Square by* 24 *in. Long, Cast Horizontally in Dry Sand, with Larger Head.*

Head dimensions: Ingot type, 5½ in. × 4 in. at the top, tapering to 8 in. × 6 in. at the bottom, by 12 in. deep.
Volume of casting: 864 cu. in.
Volume of head: 410 cu. in. = 47·5%.

At this point in the series of experiments a decision had to be taken as to whether to use two separate 9 in. deep ingot-type heads on the 24-in. length of casting, in order to investigate the effect of feeding in two horizontal directions, or to increase the dimensions

TABLE XX.—*Results of Sectioning Castings in Groups* 10 *and* 11.

6 in. square by 24 in. long.

Casting No.	Head Sand Used.	Feeding Compound Used?	Steel-making Process.	Casting Temp. App't. °C.	Casting Time. Sec.	Result of Sectioning.	
						Soundness in Head.	Depth of Pipe Cavity into Casting.
Group 10.—*Cast Horizontally, with* 4–6-*in. square Head,* 9 *in. Deep as used for the Smaller Castings.*							
*B*24	Dry sand	No	Basic arc	1435	...	Nil	2¾ in. deep.
*F*7	Dry sand	No	Acid O.H.	1415	30	Nil	2½-in. pipe with signs of secondary piping.
Group 10(1).—*As Group* 10, *with Insulating Sand round the Head.*							
*B*25	Insulating sand	No	Basic arc	1435	...	Nil	2-in. pipe.
Group 11.—*Cast Horizontally, with Larger Head,* 5½ *in.* × 4 *in. at the Top, Tapering to* 8 *in.* × 6 *in. at the Bottom by* 12 *in. Deep.*							
*F*9	Dry sand	No	Acid O.H.	1420	17	Nil	1½-in. pipe.
*A*40	Dry sand	No	Basic arc	1450	...	Nil	1½-in. pipe.
*B*26	Dry sand	No	Basic arc	1447	...	Nil	½-in. pipe
Group 11(1).—*As Group* 11, *but with Insulating Sand round the Head.*							
*B*27	Insulating sand	No	Basic arc	1447	...	Just sound	Nil.
Group 11(1)L.—*As Group* 11(1), *but with Addition of Feeding Compound as well as Insulation.*							
*B*28	Insulating sand	Yes	Basic arc	1447	...	1½ in.	Sound.
Group 11(b).—*As Group* 11, *but with* 3 *in. Thickness of Insulating Brick round the Head.*							
*B*29	Insulating brick	No	Basic arc	1447	...	2¾ in.	Sound.
Group 11(x).—*As Group* 11, *but Run from the Opposite End to the Head in Error.*							
*A*41	Dry sand	No	Basic arc	...	...	Nil	½-in. pipe, but large shrinkage cavity 6 in. long at "hot spot" in front of ingate.

of the head and use only one head. As it was the opinion of the Sub-Committee that in casting the ring of 6 in. square section, which is the ultimate aim of this part of the work, larger and elongated heads would be required, it was decided to increase the head dimensions along the length of the casting and also to increase the depth from 9 in. to 12 in.

The results of sectioning the castings is shown in Table XX. and Fig. 20, and it is apparent that with dry-sand moulds and heads

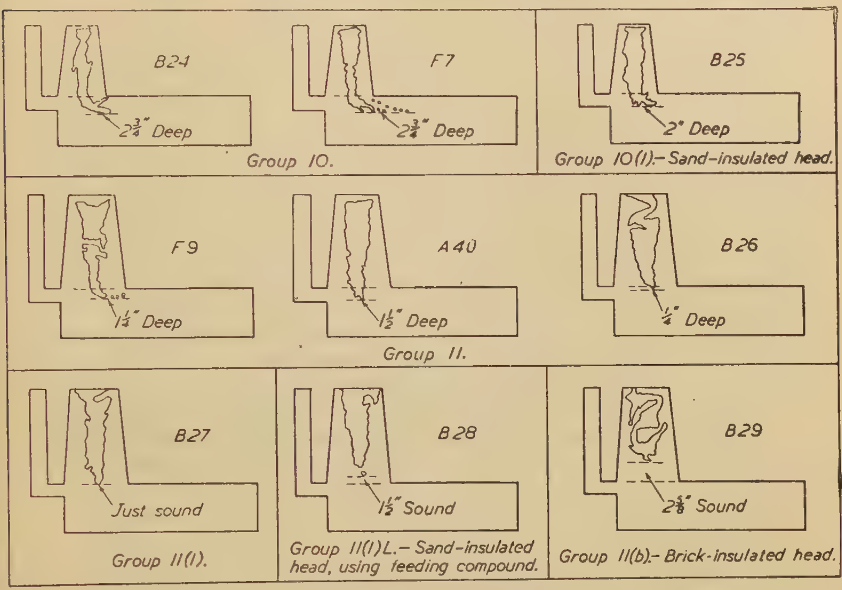

FIG. 20.—Sections of Groups 10 and 11 Castings, 6 in. square by 24 in. long.

the larger head (47·5% of the volume of the casting) is inadequate, but that by insulating the head and using feeding compound a sound casting can be made.

Casting *B*29 is interesting, because, instead of the rather friable 1½-in. thickness of standard insulating sand used on *B*27, the head was lined with a 3-in. thickness of high-temperature insulating brick. Even without the use of feeding compound this casting shows a depth of 2⅝ in. of sound steel in the head.

Incidentally, another casting of this type was made, using a low-temperature diatomaceous-earth type of insulating brick, but the fluid steel melted the brick and so the results have not been reported.

Inadvertently, one of these 24 in. long castings (*A*41) (*see*

Fig. 21) was run at the opposite end in error; the results have been included in this Report as a matter of interest, because, although the section shows a pipe cavity only ¼ in. under the head, there is a large shrinkage cavity 6 in. long in the "hot spot" in front of the runner. This example affords proof of the wisdom of the well-known foundry method of gating into the side of a head.

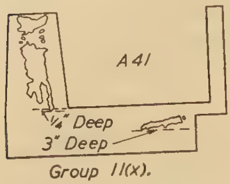

Fig. 21.—Section of a Casting Run at the Wrong End.

Remarks.

The Sub-Committee set out to find how far the head which would feed a 6-in. cube casting would feed in a horizontal direction longer castings of similar cross-section, and it was shown that satisfactory horizontal feeding did take place with the same type of head up to a length of 18 in. When the length of the casting was increased to 24 in. long it was necessary to use a larger head.

The effect of horizontal feeding is that the longer the casting in relation to its cross-section, the less is the percentage of head required when the head is placed at one end, the runner end. The Sub-Committee have not had time to determine the effect of varying the position of the head on the longer castings.

It has been confirmed that insulation round the heads and the use of feeding compounds makes a marked difference to the results.

The work has been completed as far as possible and reported, so that the investigations can be resumed when circumstances are more favourable.

Part 2.—The "Whirl-Gate" Head and the "Atmospheric" Head.

The Sub-Committee consider that the steel-foundry industry is greatly indebted to Mr. Basil Gray, of the English Steel Corporation, Ltd., for the pioneer work that he has done in developing the so-called "whirl-gate" head method of feeding castings. Mr. Gray wrote a confidential brochure which was circulated by the Institute to the industry, and he suggested that the Sub-Committee should co-operate in pooling their information and ideas on the application of this method of running and feeding, so as to provide up-to-date examples now they had had an extended period of trial.

The whirl-gate head, as it is now understood, is something more than a spinning ingate. It should consist of a head so placed and with such a short connection to the casting that the relative feed through this neck is increased by the heating effect on the sand caused by the proximity of the head to the casting. To this effect is added that of spinning the steel at the bottom of the head by a

whirling or tangential ingate. As much of the value of this form of head comes from its careful placing as from the spinning ingate.

One can divide the responsibility for the stages of development of the whirl-gate head as follows :

(*a*) The tangential, spinning or whirling ingate was applied to non-ferrous and malleable castings some years ago, particularly with a view to keeping slag and dirt from entering the moulds.

(*b*) This type of spinning ingate was applied to ordinary L-heads by Messrs. Hopkinsons, Ltd. This method of side-heading is familiar to all steel foundries, but it should be noted that this form of head *does not* have a restricted opening to the casting. Examples have been shown already in Fig. 1 (head No. 5), Fig. 3A (cube *B*5), &c., and an example with two cubes run off one head is given in Fig. 14.

(*c*) Heads with restricted openings, but so placed in close proximity to the casting that the opening became hot, providing a relative feed far in excess of what the actual size of the opening would give, were probably first introduced for steel castings in the United States of America. The feed bobs, on malleable and non-ferrous castings, however, again bore some relationship. It is certain that heads placed on the *upper* surfaces of steel castings with a very thin core giving entrance to the casting, much smaller than the head itself, were used in America before 1937. This type of head was often referred to as the "Williams" riser and the thin core was termed a "cut-off core."

(*d*) Whilst it is true that this type of head was recently applied to the side of castings in America, using thin dividing cores, it is certain that Mr. Gray was not aware of this work. It is known, however, that he suggested its application a few years ago with a view to improving the soundness of certain flat cast-steel test plates. After he had seen the spinning gate at Messrs. Hopkinsons applied to their normal L-heads, he tried it in combination with a *closely placed* head, and proceeded with his colleagues at the English Steel Corporation, Ltd., and later with the collaboration of various research committees, to pursue its development.

The Sub-Committee agree that the whirl-gate head method of feeding, correctly applied, with its consequent saving in dressing cost, is one of the outstanding advances in steel-foundry technique for many years. The bad results which many people get with the whirl-gate head are due to lack of knowledge as to (*a*) how close the head should be to the casting, and (*b*) the height of the connecting portion between the head and the casting.

The earliest co-operative work on whirl-gate heads by the Members of the Sub-Committee was carried out in 1942 to compare the

TABLE XXI.—*Summary of Results of Sectioning 6-in. Cube Castings made with a Whirl-Gate Head, Compared with Normal (Gravity) and L-Shaped Heads.*

No. of Cubes Fed from One Head.	Type of Head.	Dimensions of Head.	Dimensions of Ingate (*i.e.*, Area to be sawn through to remove Head).	Sawn Section of Cube.	Volume.	
					Head.	Cube.
1	Normal head over cube.	Ingot type, 4 in. sq. at top tapering to 6 in. sq. at bottom × 9 in. deep.	6 in. sq. covering the entire upper surface of the 6 in. cube.	Sound.	225	216
1	Normal head over cube.	Ditto, but only 6 in. deep.	6 in. sq. covering the entire upper surface of the 6 in. cube.	½-in. deep pipe (average).	150	216
1	L-type head.	4 in. dia. × 9 in. deep leading to 3 in. sq. into side of cube.	3 in. sq. at top centre of side face of cube.	Deep cavity 2–3 in. long.	168	216
1	L-type head.	4 in. dia. × 9 in. deep leading to 3 in. sq. into side of cube.	3 in. sq. at top centre of side face of cube.	Deep cavity 2–3 in. long.	168	216
1	Whirl-gate head.	4 in. dia. × 11 in. deep.	2 in. sq. at lower centre of side face of cube.	No cavity visible at ingate, but large cavity, 2 × 2 × ⅝ in., 1 in. below upper surface of cube.	138	216
1	Whirl-gate head.	4 in. dia. tapering to 5 in. dia. × 10½ in. deep with 1 in. well.	3 in. sq. at lower centre of side face of cube.	Small shrinkage spot at ingate and 2 bad internal defects 1⅜ in. below upper surface.	182	216
1	Whirl-gate head.	4 in. dia. tapering to 5 in. dia. × 10½ in. deep with 1 in. well.	6 in. sq. on side of cube.	Small shrinkage spot at ingate and large internal cavity 1⅜ in. below upper surface.	182	216
					1 Head to 3 Cubes.	
2	Central L-head feeding 2 cubes.	8 in. dia. × 10 in. deep.	4 in. sq. at top centre of side face of each cube.	Sound.	500	432
					1 Head to 2 Cubes.	
3	Central whirl-gate head feeding 3 cubes each with a different ingate area.	7 in. dia. at top tapering to 6 in. dia. at bottom × 18 in. deep.	2 in. sq. 3 in. sq. 4 in. sq. Central on side face, but 1 in. above bottom edge.	Sound. Sound. Sound.	600	648
3	Central whirl-gate head feeding 3 cubes each with a different ingate area.	7 in. dia. at top tapering to 6 in. dia. at bottom × 18 in. deep.	2 in. sq. 3 in. sq. 4 in. sq. Central on side face, but 1 in. above bottom edge.	Small cavity in centre of cube. Sound.	600	648
	Central whirl-gate head feeding 3 cubes each with a different ingate area.	Ditto, but only 12 in. depth of head.	2 in. sq. 3 in. sq. 4 in. sq. Central on side face, but 1 in. above bottom edge.	Sound. Sound. Sound. Sound.	395	648

new method of gating and heading on the 6-in. cube casting described in Part 1 of this Report.

Table XXI. gives the results of a series of castings which were gated in different ways and sectioned for examination.

In the first place single cubes were cast, and it was apparent that a 4-in. dia. side-head was insufficient to give a sound 6-in. cube casting. It was evident that if a larger head were used it would be possible to feed 3–6-in. cubes from one central head, and, by varying the size of the ingating where the head joins the casting, it might have been possible to determine critical dimensions for these conditions. The ingates were placed 1 in. above the bottom edge of the cube, and three decreasing sizes were employed, namely, 4 in. sq., 3 in. sq. and 2 in. sq., with a closely placed central head 7 in. in dia. at the top tapering to 6 in. in dia. at the bottom.

As sound sections were obtained with even the 2-in. sq. ingate, the limiting dimension had not been reached, but the fact that a 6-in. sq. section could be fed through a 2-in. sq. section meant that a great saving in burning-off time could be achieved. This was the first advantage which became apparent when whirl-gate heads were used.

The second advantage is that, as these castings are virtually "bottom-run," there is less sand erosion and the castings strip more easily than similar castings which have been "top-run" in the same sand.

In certain cases it will be seen that larger moulding boxes might be required to accommodate the side-feeding heads, and when casting single castings run off one whirl-gate head no saving in fluid weight of steel could be claimed. On the question of the reduction of the weight of fluid steel, however, two other factors emerge. The first is the possibility of running a number of castings from a common feeder head, and the second factor was the introduction in America of the "blind head," using an atmospheric core. By both or either of these methods it is possible to reduce the amount of feeder head required for certain castings.

Another advantage of side-feeding becomes apparent with designs in which the upper surface is of varying contour. Top-feeding heads would necessitate a considerable amount of burning or machining to shape, whereas with the side-feeding head the upper surface is not disfigured by the feeder head.

Some examples of the commercial application of side-feeder heads are given in the illustrations enumerated below:

Fig. 22 shows the application of an open side-feeder head 5 in. in dia. to a flywheel $22\frac{5}{8}$ in. in dia., giving a yield of 69·87%.

Fig. 23 is a similar casting $20\frac{5}{8}$ in. in dia. fed with an open side-head $5\frac{1}{2}$ in. in dia. In this case the yield was 54·72%.

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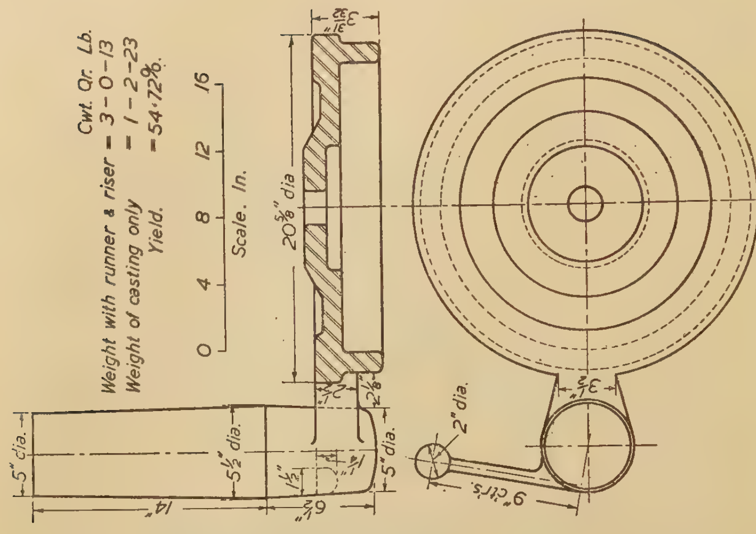

FIG. 23.—Whirl-Gate Feeder for Steel Flywheel Casting $20\frac{5}{8}$ in. in dia.

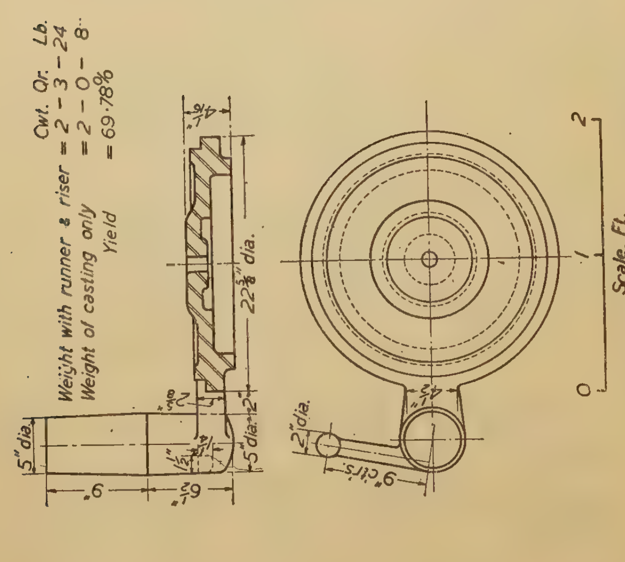

FIG. 22.—Whirl-Gate Feeder for Steel Flywheel Casting $22\frac{5}{8}$ in. in dia.

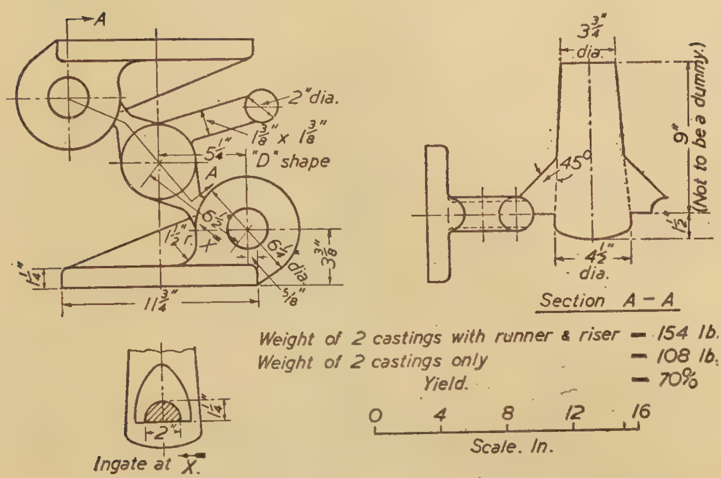

FIG. 24.—Whirl-Gate Feeder for Two 2-in. Link Plates.

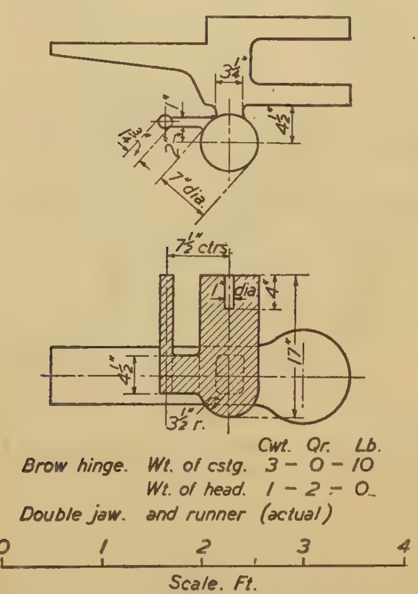

FIG. 25.—Atmospheric Head applied to a Brow Hinge (double jaw).

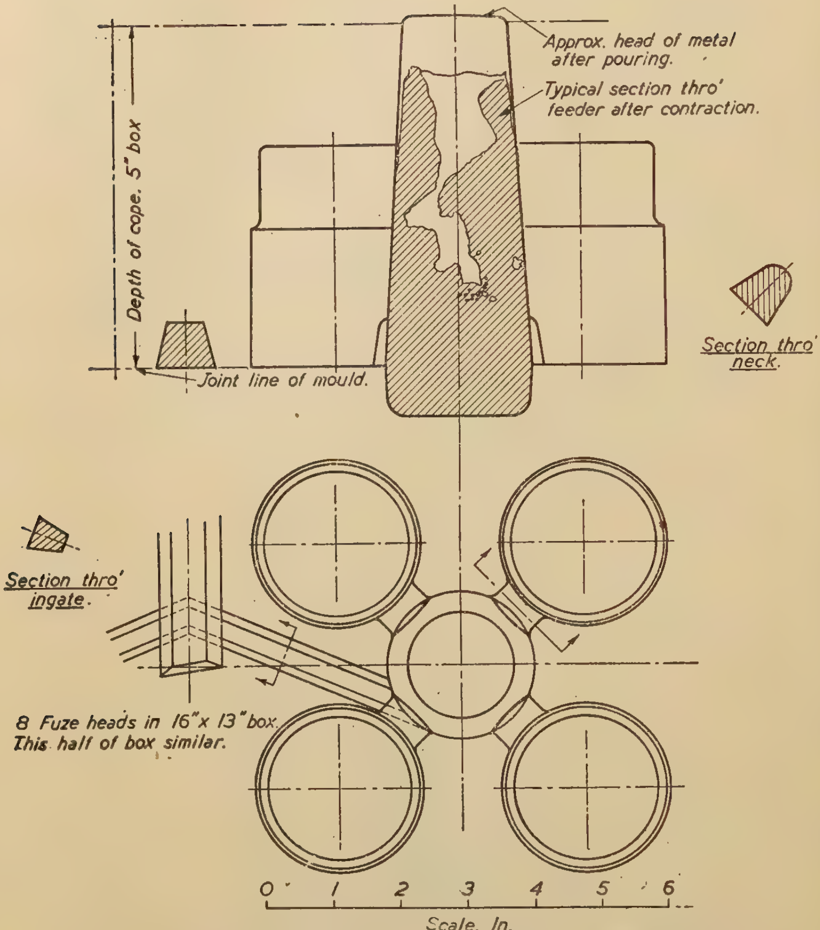

FIG. 26.—Whirl-Gate Head applied to Fuze Heads in Cast Iron, showing how atmospheric pressure can balance the liquid pressure in a casting.

Fig. 24 is an example of a centrally placed head feeding two cast-steel link plates. The yield in this case was 70%.

The above three examples were supplied by Mr. D. W. L. Menzies, of The North British Steel Foundry, Ltd., Leith, who comments that the real benefit is the saving of fettling and grinding costs.

Fig. 25 illustrates the use of the atmospheric core in a blind head feeding a brow hinge made by Messrs. Davy and United Roll Foundry, Ltd.

Fig. 26, supplied by Mr. Kain, of Messrs. Lake and Elliot, Ltd., Braintree, shows four fuze-head castings fed from one central open side-head.

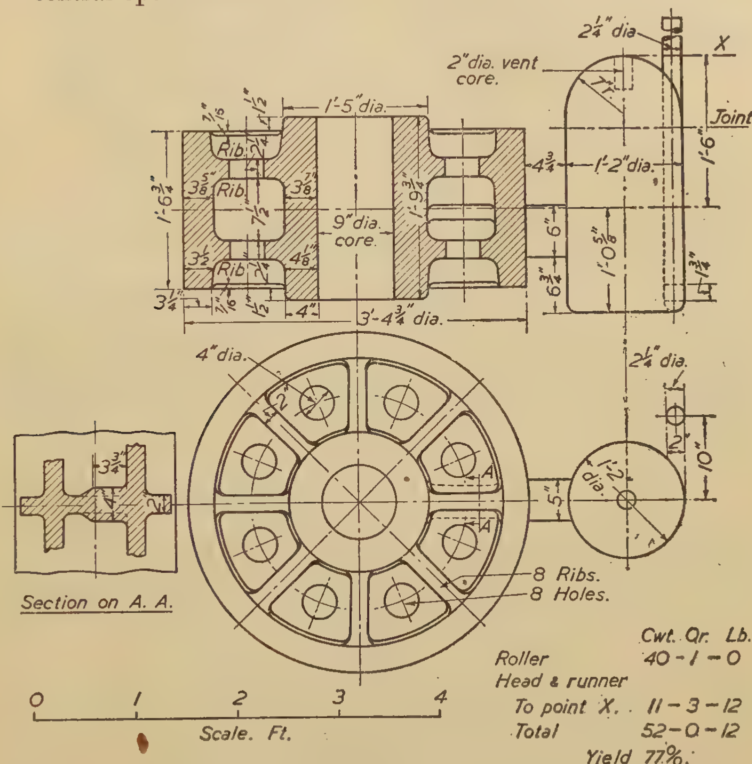

Fig. 27.—Cast-Steel Roller with Atmospheric Head.

Fig. 27 is an example, by Messrs. Hadfields, Ltd., where a considerable saving in fluid steel weight was made when an atmospheric core was used in conjunction with a blind head to feed a roller weighing 40 cwt.

Fig. 28 shows the use of open side-feeder heads on a heavy steel casting. This example of a 37¾-ton casting was furnished by the English Steel Corporation, Ltd.

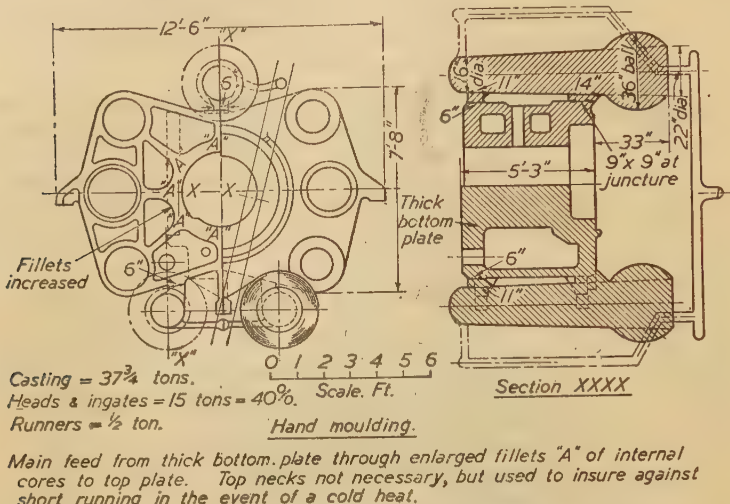

Fig. 28.—Open Side-Feeder Heads on a Heavy Steel Casting.

The above examples have been selected to illustrate the main factors of the new forms of side-feeding heads. A note on the influence of the mechanism of freezing by Mr. Basil Gray has been published by The Iron and Steel Institute.[1]

Part 3.—Feeding Compounds.

The majority of the castings of simple shape reported in Part 1 of this Report were made without the advantage of the use of feeding compounds on the fluid steel as it rose into the heads. This procedure was carried out deliberately, because when the experiments were in their early stages (in 1938 and 1939) a variety of British and foreign feeding compounds was being used by steel foundries, and the Sub-Committee did not wish to confuse the issue by introducing another variable into the study of the feeding characteristics of differently shaped heads.

However, after the early castings had been made, Members of the Sub-Committee undertook to carry out carefully controlled foundry trials to check the efficiency of proprietary and synthetic feeding compounds, with the object of determining the best type

[1] "The Side Feeding of Steel Castings. A Note on the Influence of the Mechanism of Freezing," *Journal of The Iron and Steel Institute*, 1944, No. II., p. 5 P.

of compound, and to discuss the reasons for the behaviour of the different types.

Chemical analyses of trade compounds showed wide variations in the ideas of the manufacturers, and steel foundrymen had likewise exercised their ingenuity to find the best compound for their own use. The various types used may be summarised as follows :

(1) Heat-evolving types containing aluminium powder and iron oxide with an inert filler powder (some of the low-melting-point mixtures giving fluid slags).

(2) Similar types to (1) with increasing amounts of carbonaceous matter.

(3) Finely ground or "falling" electric furnace slag, with pieces of charcoal robbled into the slag after casting.

(4) Papier-maché shapes (may be satisfactory for ingots, but quite useless for castings, where the heads are usually smaller and of various sizes and shapes).

(5) Chopped straw.

(6) Impure blacklead, with about 40–50% of ash.

(7) Finely ground coke dust or coal dust with adulterant.

The Sub-Committee's test results showed that the best type of feeding compound was No. 6, and the reason for its success was that as the carbon burned away slowly it left on the surface of the fluid steel a light flocculent type of ash, which formed an ideal insulation of the surface of the steel in the head. Compounds containing thermite (No. 1 type) produced a narrower and deeper pipe which tended to extend into the casting, but it was noticed that as the percentage of carbonaceous matter was increased the efficiency of the compound improved. Attempts to simulate the No. 6 type by making No. 3 (charcoal and slag) or No. 7 (coke dust and fine dust) did not succeed, because the ash layer left on the surface was not as insulating as that left when an impure blacklead burns slowly.

Chopped straw is used successfully on large heads and depends for its efficiency on the same effect, that is, it chars slowly and leaves a light insulating layer on the steel in the head.

Part 4.—Hot Tears.

(1) Cracks which occur in steel castings during manufacture may be divided into two classes, namely :

Class 1.—*Cracks controlled by Composition, Fluidity and Design.*

(*a*) Shrinkage cracks, hot tears or pulls, visible on the surface when the casting is removed from the mould and shot-blasted.

(*b*) Internal shrinkage fissures, which may or may not become visible after the surface has been scaled in heat treatment.

(c) Internal laps, which spread outwards and appear as horizontal cracks on the vertical faces of castings after heat treatment, particularly treatments involving quenching. They are seldom seen in carbon-steel castings.

Class 2.—Cracks controlled by Composition, Design and After-Treatment—commonly called "Clinks."

(a) Clinks caused by the release of casting stress or heat-treatment stress.

(b) Clinks developed by local applications of heat, as in uneven or too-rapid heating for annealing, &c., burning-off operations and welding.

(c) Clinks caused by the release of stress set up by heavy machining.

(2) This Part of the Report deals with the class 1(a) type of crack, better known by the foundry terms "hot tears" or "pulls," and distinguishable from the cracks of class 2 by the nature of the path of the crack across the casting. A hot tear follows a tortuous path across the face of the casting and the crack has jagged edges, whereas a "clink" follows a straight path.

(3) All steel foundrymen know that hot tears can be avoided by design modifications, the choice of low-sulphur-content steel, adequate fluidity of the steel, the use of green sand in cores or in portions of moulds which would offer resistance to the free contraction of flanged portions of castings, or by digging out intervening sand (easing) to allow the casting to contract without restriction immediately after solidification has taken place. The problem that was set to the Sub-Committee, however, was to formulate methods of a practical nature which would furnish a numerical index of the tendency of a given steel to hot-tear in moulds.

(4) In the First Report of the Steel Castings Research Committee [1] a difficult branch pipe casting (287 lb. weight) was made in various foundries and sixteen castings were critically examined by the Research Department, Woolwich. Table XXII. illustrates the influence of the sulphur content on the hot tears found on these intricate castings.

TABLE XXII.—*Influence of Sulphur Content on Incidence of Hot Tears.*

Incidence of Cracks.	Sulphur. %.	
	Basic Electric Steel.	Acid O.H. Steel.
No cracks . .	0·002, 0·003, 0·003, 0·008, 0·008	0·024
Slight cracks . .	0·015, 0·016, 0·024, 0·026, 0·028	0·024
Serious cracks .	0·020, 0·026, 0·028	0·033

[1] *The Iron and Steel Institute*, 1933, *Special Report No.* 3.

The fact that basic electric steel with less than 0·008% of sulphur was needed to avoid pulls, while a casting made in acid open-hearth steel with 0·024% of sulphur did not tear, illustrates why foundrymen favour acid open-hearth steel for castings.

(5) In 1936 the Committee designed a wheel, 18 in. in dia. with the rim $1\frac{1}{2}$ in. thick × 4 in. deep, and four straight spokes varying from $\frac{1}{2}$ in. to 2 in. thick, but even when the spokes were reduced to $\frac{1}{4}$ in. thickness, green-sand and dry-sand castings in carbon steel did not crack. In the following year, foundry trials were carried out with a multi-section design, a box casting, 18 in. × 12 in. × 4 in. deep, open at the top and bottom, with outer walls varying from $\frac{3}{4}$ in. to $1\frac{1}{2}$ in. thick and ribs $\frac{3}{4}$ in. and $\frac{1}{2}$ in. thick. The casting was fed from a large L-head at one end and made in open-hearth and Tropenas steels in two foundries for comparison. The results showed that lower sulphur content reduced the tendency to hot tear, and if the steel casting was free to move during contraction no pulls occurred. There were no pulls in green-sand, but numerous pulls in the dry-sand castings. The design could not be put forward as a standard test to obtain the desired numerical index.

(6) After devising a testing machine to determine the tensile properties of steel while cooling down from solidification,[1] Hall made a noteworthy contribution to the problem in developing a test with multiple flanged bars cast vertically with an enlarged middle portion or bulb, which, by reason of its delayed cooling, predisposed this part of the bar to failure.[2] His patterns were used in commercial foundries by Members of the Foundry Practice Sub-Committee in order to extend the range of steels tested.

At the ends of the test bars were flanges 2 in. in dia. with brackets attached. The total length between the flanges was 12 in., of which the bulb occupied 4 in., thus leaving a 4-in. length of shank above and below the bulb. Eight bulb diameters were employed, namely, 0·52 in., 0·60 in., 0·69 in., 0·72 in., 0·80 in., 0·88 in., 0·96 in. and 1·03 in. The 0·52-in. diameter represents a parallel bar. Each bar was made in two half oil-sand cores, split longitudinally, gummed together and assembled vertically around a central down-runner $1\frac{1}{2}$ in. in dia., from which projected eight ingates $1\frac{1}{2}$ in. × $\frac{1}{2}$ in., one to the bottom flange of each test bar. Each group of castings was poured from a 1-cwt. shank, and the steel remaining in the shank was used to obtain an immersion-pyrometer reading. The steels tested in the production foundry are shown in Table XXIII. and the results in Table XXIV.

It was expected that for each nest of eight test bars, one or more of the smaller-diameter bulbs would be sound and that all

[1] Second Report of the Steel Castings Research Committee, Section IV., *The Iron and Steel Institute*, 1936, *Special Report No.* 15.

[2] Third Report of the Steel Castings Research Committee, Section IV., *The Iron and Steel Institute*, 1938, *Special Report No.* 23.

TABLE XXIII.—*Analyses of Steels used for Multiple Test Bars.*

Steel.	Process.	C. %.	Si. %.	S. %.	P. %.	Mn. %.	Mo. %.	Cu. %.
A	H.F. furnace.	0·22	0·16	0·027	0·023	0·63	...	...
B	Converter.	0·20	0·20	0·058	0·051	0·58	...	...
C	H.F. furnace.	0·46	0·45	0·037	0·030	0·95	...	...
D	Converter.	0·37	0·37	0·065	0·051	0·78	...	...
E	H.F. furnace.	0·28	0·20	0·026	0·024	1·67	0·32	...
F	,, ,,	0·21	1·10	0·023	0·026	1·53	...	1·90

larger bulbs above a certain " critical diameter " would contain a hot tear. The results in Table XXIV. show that, whilst this ideal

TABLE XXIV.—*Examination of Bulbs after Casting.*

Steel.	Casting Temp. (Immersion). ° C.	S. %.	Bulb Diameter. In.							
			Parallel.	0·60.	0·69.	0·72.	0·80.	0·88.	0·96.	1·03.
A	1707	0·027	S	S	S	S	S	S	C	S
B	1570	0·058	S	PB	S	PB	PB	PB	PB	PB
C	1595	0·037	S	S	S	S	PB	PB	S	S
D	1600	0·065	S	PS	S	PS	PB	PB	S	PB
E	1660	0·026	S	S	S	S	S	S	S	S
F	1620	0·023	S	S	S	S	S	S	S	S

S = Sound. C = Slight crack on bulb found on shot-blasting. PB = Bulb pulled apart. PS = Shank pulled apart.

was not realised, steels *B* and *D* were inferior, and steels *A*, *E* and *F* were superior, with steel *C* possessing intermediate resistance to the formation of hot tears.

If the results are regrouped in order of casting temperature as in Table XXV., the formation of hot tears in the bulb diminishes as the casting temperature is raised, despite the wide variation of chemical composition, which suggests that this form of test may be more of the nature of a fluidity test than of a true hot-tear test.

TABLE XXV.—*Relation between Casting Temperature (Immersion) and the Number of Test Bars showing Hot Tears.*

Casting Temp. (Immersion). ° C.	S. %.	Number of Bars Pulled.	Steel.
1707	0·027	None (1 cracked)	0·22% carbon.
1660	0·026	None	Manganese-molybdenum.
1620	0·023	None	1·9% copper.
1595	0·037	2	0·46% carbon.
1600	0·065	5	0·37% carbon.
1570	0·058	6	0·20% carbon.

The patterns and special moulding boxes were used by another foundry, where a set of test bars was cast in acid open-hearth steel

and oil-sand cores, but only one bulb showing a hot tear was found.

The problem has had to be left in abeyance for the time being, as the method of moulding is very expensive as regards labour, and at the present juncture in the war there is no spare labour available to continue the work.

There are three points which require elucidation :

(1) Does the test give reproducible results on the same steel cast at the same temperature, and has the "rate of rise" in the moulds any influence on the results?

(2) What is the influence of casting temperature?

(3) Does the use of a less collapsible sand increase the sensitivity of the test?

(7) An interesting sidelight on hot-tearing was given by Andrew and Protheroe,[1] who investigated the influence of mould friction and showed that long steel castings 4 ft. long × $1\frac{1}{2}$ in. wide × $\frac{3}{8}$ in. thick made in coarse sand did not tear any more severely than castings made in fine sand. The degree of tearing varied with the sand bulk between the runner and riser, being more severe as the bulk of sand increased. They showed that, by using sloping runners, the slight ease given to the contracting casting against the bulk resistance of the sand was sufficient to prevent tearing.

(8) Members of the Sub-Committee engaged on the production on a large scale of certain thin-walled cylindrical castings have supplied records of the percentage of castings showing hot tears from a large number of heats made under practical conditions; these are summarised in Table XXVI.

Table XXVI.—*Relation between the Composition and the Percentage of Thin-Walled Castings showing Hot Tears.*

Number of Castings in Group.	Analysis.		Castings Torn. %.
	S. %.	P. %.	
34	0·038–0·055	0·051–0·058	Nil
30	0·054–0·062	0·055–0·066	Nil
20	0·037–0·055	0·047–0·109	Nil
8	0·050	0·087	Nil
305	0·066–0·077	0·072–0·080	10
24	0·061–0·080	0·061–0·063	17
6	0·089	0·060	100

Another foundry making a larger casting of the same type has reported a similar relation between the sulphur content of the steel and the tendency of the castings to show hot tears.

The general view of the Members of the Sub-Committee is that, after all the skill of the foundryman has been exercised, the degree

[1] *Journal of The Iron and Steel Institute*, 1942, No. I., p. 101 P.

of hot tearing on a particular design can be minimised or often eliminated by using steel with a lower sulphur content. With normal designs there should be little trouble if the sulphur content is below 0·035%, but for more intricate designs it may be necessary to go to still lower limits.

Part 5.—War-Time Activities.

(1) The long-term programme of foundry research envisaged by the Sub-Committee had to be curtailed during the busiest years of the war period, as increasing calls were made on the Members for their attendance on other committees set up to deal with problems of design modifications and production schemes—work which cannot be reported to the Institute but which served the useful purpose of bringing together, for the common good, steel-foundry executives from all parts of the country. One interesting example of the spirit of co-operation which has grown up is the scheme whereby a foundry making satisfactory samples of a new design would prepare and circulate drawings showing full moulding details for the assistance of other steel foundries called upon to make the same casting.

(2) For a time the meetings of the Sub-Committee had to be discontinued, except when foundry questions of a general nature were submitted by Government Departments for concerted action. Many of the smaller steel foundries were being called upon to produce castings of which they had had no experience, and one problem which was quickly solved from the experience of Members of the Sub-Committee without the necessity of carrying out foundry experiments was the replacement of cast-iron solid practice shells, 8 in. and 16 in. in dia., by cast steel to a rigid mechanical-test specification. The foundry had been attempting to produce these solid shells in sand moulds, and had encountered circumferential and longitudinal cracks, revealed when the surface was removed in machining. Other foundries were able to report that success had been achieved by casting into ingot moulds of special shape, and it was recommended that the new foundries should be put into contact with those steelmakers who had made these shells successfully. It was further recommended that any steelmaker undertaking the manufacture of these shells should have facilities for machining the casting, or be able to keep in close touch with the machining results during the early stages of manufacture, in order that steelmaking variables, casting temperatures and pouring speeds could be correlated with the incidence of defects and the necessary steps taken to build up a satisfactory technique from this experience.

(3) Another urgent problem investigated by the Sub-Committee was the possibility of providing cast-steel target plates, 4 ft. 0 in. square by 10 mm. thick, which must be uniformly sound throughout

to test small-arms ammunition. Sample plates 2 ft. 0 in. square were cast in carbon steel in two thicknesses, 10 mm. and 20 mm., and X-ray examinations were made to determine the influence of different methods of gating and heading. The conclusion was reached that, whilst it seemed difficult to obtain sound 10-mm. plate of this area, the problem was simpler in the case of the 20-mm. plate. The X-ray examinations showed that the fluidity of the steel was of great importance in obtaining soundness. Adequate soundness in a 10-mm. thick plate was obtained only by running 10° uphill, with four good heads on the face of the plate. When vertical or horizontal casting was attempted with heads at the edges of the plate, it was apparent that the 10-mm. section was too thin to permit satisfactory feeding to overcome centre-line weakness.

(4) The use of substitute moulding materials and the increasing interest in methods of desulphurisation and dephosphorisation following the deterioration in the qualities of raw materials available for steelmaking have been discussed by the Members in collaboration with the relevant Sub-Committees of the Institute.

Part 6.—Fundamental Casting Problems.

With a return to more normal conditions the Members of the Sub-Committee will be able to devote their attention to the co-operative study under practical foundry conditions of the programme set out in Part 1 of this Report. Whilst it is true that every new design must be dealt with separately and different foundries often mould and run similar castings in different ways to suit the available equipment, it is hoped that the work will tend to clarify the principles involved in producing sound steel castings in the most economical fashion.

Investigations dealing with the effects of variation in the rates of pouring castings in different foundries and with methods of feeding 100-mm. thick plates are already in hand. In addition, the Sub-Committee are preparing a pamphlet on the "Design of Steel Castings," which, it is hoped, will have a wide circulation in the engineering industry and serve the purpose of bringing together the foundry executives and the designers while designs of steel castings are still on the drawing board.

The future progress of the steel castings industry depends on the co-operation of the foundry executives, who have already given so much of their time and experience, and on the generosity of individual manufacturers, who have carried out the practical experiments recorded in this Report at their own expense.

DISCUSSION.

This Report was presented at the Annual General Meeting of the Institute in London on 11th July, 1945.

Dr. H. T. ANGUS (British Cast Iron Research Association, Birmingham): We must congratulate the Sub-Committee on commencing this work on a fundamental basis by attempting to study the feeding of steel castings in its simplest aspect. It is important to realise that in the work described the only cavities discussed are the gross cavities; as all steel founders are aware, the finer degrees of porosity are concealed by saw-cuts rather than revealed. Satisfactory examination by means of fracture or deep etching is often very difficult, and I think it would be desirable, in an extension of the work, for the Sub-Committee to consider the use of density determinations as well as saw-cuts, X-rays, &c. The advantage of determining the soundness of a casting by its density is that a numerical figure can be given; of course, the estimation should be carried out before any sectioning is undertaken. Shrinkage defects which reach the surface will obviously have to be stopped off before the density determinations are made, but a complete examination of the results obtained both by the final sectioning and by the density determinations will give considerably more value to the work which will be done.

Some preliminary work which I carried out two or three years ago indicated that even in apparently sound castings (sound as far as X-ray examination could determine it) substantial variations in density occurred, both with weight of feeding head and with pouring temperature for otherwise identical castings.

Dr. L. NORTHCOTT (Teddington): This Report is, in effect, a presentation of experimental data. There is little in the way of theory or speculation involved, so that it does not call for very much comment in that respect. I feel that the Sub-Committee might have made some more definite comment on p. 355 P, where they say: "There is some suggestion in the results that hot steel poured slowly gives less piping than cold steel poured quickly." As far as my experience goes, both with steels and with non-ferrous alloys of all types, what is wanted in a finished casting is something rather different from what is readily obtainable practically, for the simple reason that to get good surface quality it is desirable to pour as quickly and as hot as possible, whereas to obtain adequate soundness it is preferable to pour as slowly and as cold as possible. There are, of course, limits to the control available, and in practice some reasonable balance has to be maintained.

The main point to be mentioned is one raised by Mr. Angus just now, namely, testing for soundness or unsoundness. In some somewhat similar work being carried out on various alloys in order to determine the correct methods of pouring and correct moulding conditions, it has been found very useful to section castings, to etch and in particular to take samples for mechanical tests. It is very disconcerting, when you get what appears to the eye to be a reasonably sound test-bar, to obtain an elongation figure of 2% or 3%, and sometimes nothing at all. Conversely, there is the opposite experience of having what appears to be an obviously unsound test-bar, and obtaining quite reasonable elongation and mechanical-strength figures. The effect is due, of course, to the form in which the porosity occurs, and the finer types of porosity, the intergranular and interdendritic type, are not very well shown by radiographic work but are brought out very well by mechanical tests. There is no need to carry out

mechanical tests on the simpler shapes, and I would suggest, in view of the amount of work which the Sub-Committee have to do, that mechanical tests should be restricted to the more complicated shapes of castings.

In Fig. 16 the Sub-Committee make a comparison of shrinkage cavities in 11-in. square ingots, chill-cast and sand-cast, and there is one point to be mentioned there. It is probable that a chill-cast steel ingot is slightly smaller than a sand-cast ingot when the two are originally intended to be of the same size. A chill-cast ingot rapidly forms a thin skin which very soon shrinks away from the mould, whereas that occurs less with a sand casting of any size.

I should like to suggest, for further work, that the Sub-Committee should give us the benefit of their experience in chilling or slow pouring or after-feeding. On certain types of castings where one attempts to get preferential solidification or directional solidification by having a chill at the bottom of the mould, it is possible to obtain a relatively sound casting, using not more than about 20% excess metal. I should like to see the Sub-Committee's results on any work that they do on a larger scale; I have been limited to ingots about 5 in. in dia.

I should also like to ask whether the Sub-Committee have had any experience of special methods of casting. The methods referred to in the Report are entirely orthodox, and one would be glad to see any results obtained by any unorthodox method of casting, such as, in particular, the Durville method, where the pouring can be done at only 2° or 3° C. above what is normally considered the liquidus. There is also the possibility, for special castings, of continuous feeding. In the non-ferrous metal industry continuous feeding is used a great deal—not necessarily strictly continuous, but after pouring a series of castings the ladle is returned and runs down the line a second and perhaps a third time.

This Report had previously been discussed at a Joint Meeting of The Iron and Steel Institute and the Sheffield Branch of The Institute of British Foundrymen to which the Sheffield Society of Engineers and Metallurgists and the Sheffield Metallurgical Association were invited, held at the Royal Victoria Hotel, Sheffield, on Monday, 26th March, 1945, at 7 P.M., with Dr. C. J. Dadswell (English Steel Corporation, Ltd., Sheffield), Past-President of the Sheffield Branch of The Institute of British Foundrymen, in the Chair. The following is a summary of the discussion.

Mr. F. COUSANS (Messrs. Catton & Co., Ltd., Leeds), Chairman of the Foundry Practice Sub-Committee, summarised the Report.

Mr. J. G. PEARCE (British Cast Iron Research Association, Birmingham), in congratulating Mr. Cousans on the work of the Sub-Committee, said that the recording for comparison purposes of the method of making standard steel castings was a very difficult task, particularly with regard to the steel condition. He noticed that no data were given regarding the gas content of the steels; his opinion was that the gas content of the fluid steel had a pronounced effect on hot-tearing and feeding. Further, he thought that more detailed information should be sought or made available regarding the types of inclusion and their influence on hot-tears. Work in the United States of America had shown that the distribution had a distinct effect on the tendency to hot-tear.

Mr. COUSANS agreed with Mr. Pearce and referred to Mr. Protheroe's paper for the Steel Castings Research Committee in which three different methods of deoxidation were employed.* In any future work the Sub-

* *Journal of The Iron and Steel Institute*, 1944, No. II., p. 157 P.

Committee would endeavour to standardize the methods of deoxidation in order to control the types of inclusion in the steel. It was unfortunate that no analyses of the gas contents of the various steels examined were available, but it was a point which would be taken care of in future work.

Mr. Howard Biers (Electro-Metallurgical Company of Canada, Ltd., Grimsby, Ontario, Canada) agreed with the previous speaker that he would have liked to have seen greater attention paid to the question of inclusions, which his own work in Canada had shown to be of great importance in controlling the formation of hot-tears.

Mr. W. H. Salmon (Messrs. Hadfields, Ltd., Sheffield), Secretary of the Sub-Committee, said that the most tantalising problem that they had been set was to devise a foundry method which would give a numerical index to compare the ability of various steels to resist the formation of hot-tears. The fluidity of the steel, the casting temperature and the moulding operations were variable factors which had to be taken into consideration in discussing the results; he had recently been interested to hear of some practical work which might throw some light on the problem. Messrs. Millspaughs, Ltd., were centrifugally casting hollow cylinders in steel and had trouble with hot-tears. Nothing could be done to effect any improvement until the silicon content of the steel was raised to about 0·60%. Immediately, 33 castings were produced free from hot-tears. The steel was basic electric, low in sulphur and phosphorus. He wondered if the Sub-Committee would approach the problem of obtaining a "numerical index" by using a small centrifugal casting machine, in which a definite stress could be imposed on the solidifying steel, and by making a standard cylindrical casting, in various steels, record the hot-tears found. Alternatively a critical speed might be found, which would vary with the capability of the steel to resist the centrifugal stress imposed.

Mr. Cousans said that the possibilities of this method of investigation would be considered by the Sub-Committee.

Mr. W. C. Meredith (Messrs. Hadfields, Ltd., Sheffield), referring to the last sentence on p. 355 P, which said that under the conditions of the present experiments steels made by the acid process gave greater piping than basis electric arc steels, commented that in his experience basic electric steel gave him more trouble with shrinkage cavities than acid steel. Castings which had been made successfully in acid steel had needed larger feeder heads when the steel was changed to basic.

Mr. Cousans reminded Mr. Meredith that the Sub-Committee were only reporting their experience with a particular casting of simple shape, and it had been considered wise to insert the phrase "under the conditions of the present experiment." Since the early work had been done we were a lot wiser; we know a lot more about the vagaries of our optical pyrometers than we did a few years ago. There might have been differences in the casting temperatures, and he hoped that in future work it would be possible to use the more accurate immersion pyrometer entirely.

Dr. Dadswell (Chairman) interposed that the method of deoxidation might have some bearing on this question of feeding.

Mr. T. R. Walker (English Steel Corporation, Ltd., Sheffield) said that it might be wondered why a Sub-Committee formed in 1938 had not reported until 1945. As a matter of fact this was the third time that this

Report had been written. They could not get permission to publish the results earlier as it was thought that some of the information might be of use to the enemy. He recalled that at the inaugural meeting six foundry managers and himself were present. They decided that they should have a standard design, a 6-in. cube, and each of the foundry managers had a different opinion as to the type of head that he would use. That was the state of the art in 1938, and he did not think it was much better now. In many foundries, when an order was received for a non-standard casting, the foundry people would remember that a similar casting had been made many years before. Written records of the results obtained were, however, not usuallly available, so that the problems of moulding, gating and feeding had to be solved anew. He considered it very desirable for every foundry to keep detailed records of the methods used to manufacture each type of casting, so that, when orders for the same or similar castings were received, the experience previously gained could be utilised. One valuable contribution of the Sub-Committee had been that accurate observations had been made throughout all their experiments, and the records in detail were available as a permanent guide to all those interested in the problems involved. There must be hundreds of similar experiments yet to be made before they could give the steel foundryman all the information he required regarding running and heading.

Mr. E. J. Brown (Messrs. Hadfields, Ltd., Sheffield) said he was aware of the greater use that was being made of technical resources nowadays compared with 1938 and would have liked to see references to the type of inclusion in the steels examined. He wondered why the cube castings were not examined by X-rays rather than by sawing. He would always look upon optical-pyrometer readings with a certain amount of doubt.

The casting of these experimental castings involved the question of differential freezing. If one could get a steep gradient, keeping the top hot and the bottom of the casting cool, one had a better chance of making a satisfactory casting.

He asked why the Sub-Committee had decided to cast the experimental castings without using a feeding compound. Without a feeding compound the steel froze over in the heads, and one got a " sucking-back " or drawing when the last metal froze.

With regard to the Ruff test, his experience was that great inaccuracies were caused by variations in the method of pouring, and a better fluidity test was required for steel. Another important test required by the steel founder was a yardstick for measuring hot-tearing.

Mr. Cousans replied that the Sub-Committee had decided to make these experimental castings with or without the use of feeding compounds. It was felt that the use of feeding compounds was a subject for a separate investigation, and the work of the Sub-Committee on these materials was summarised in the First Report. One point which had come out of the work was that insulating heads were of far greater value than feeding compounds.

Mr. J. G. Bailes (Messrs. Wm. Jessop & Sons, Ltd., Sheffield) was rather amazed at the boldness of the Sub-Committee in tackling so many variables at once. He suggested that one type of moulding sand should have been supplied by a central foundry and sent to all the foundries making the test castings. The other variable was the steel; each foundry used a carbon steel of their own.

With regard to inclusions, he suggested that the samples should have been examined for refractories and non-metallics.

He wondered whether the Sub-Committee would be prepared to specify a standard sand applicable to the whole country. Further, would they issue an encyclopaedia of steel castings and show which method of manufacture would give the best results for different types of castings. Would the Sub-Committee be bold enough to publish a correlated report, pooling the knowledge of its Members, and say that if one made a certain casting in this way or that way, with a specified type of sand, one could forget the steel aspect and produce a sound casting?

Mr. Cousans replied that the standardization of moulding sand was a many-sided problem. During the war, the Foundry Bonding Materials Control had been anxious that foundries should try to work with sands from near-by deposits, to save transport and the costs of transport.

The Moulding Materials Sub-Committee had done invaluable work in this connection and had made available to all steel founders very sound, fundamental knowledge of all the sources of moulding sands available. The Members had given freely of their experiences in sand mixtures and the behaviour of sands used for steel-casting production, so that, as far as the moulding problem was concerned, adequate information was readily available. With regard to methods of gating and heading, the steel foundryman must use his initiative, and decide what was best for his own conditions. The Sub-Committee had published the results of their experiments; it was for the foundryman to decide whether to use the published data. He would not agree that the effect of steelmaking variables should be excluded from the Sub-Committee's work. The steelmaking was an important side of the foundry; that was why so much information had been included in the Report. The Sub-Committee were anxious that the foundryman should have all the information about the steelmaking technique used in these experiments.

If the Sub-Committee were to attempt to show the best method for every design of steel casting going through the foundry, the examples would be so numerous that one would require a large number of brochures. The Sub-Committee, quite rightly, decided to commence with simple shapes and to build up to complicated shapes in logical stages.

At the present time the Sub-Committee were engaged on the preparation of a pamphlet on the "Designing of Steel Castings," which would have a wide circulation in the engineering industry and would serve the purpose of bringing together the foundry executives and the designers while new designs of steel castings were still on the drawing board. All the Sub-Committee could do was to determine the underlying principles governing the production of sound steel castings and leave it to the practical foundryman to endeavour to apply the results obtained.

Dr. A. H. B. Cross (The Brown-Firth Research Laboratories, Sheffield) remarked that the Report was an admirable attempt to elucidate some of the fundamental problems of steel-foundry practice. It was logical that the Sub-Committee should have commenced with a simple shape, and their results showed that the production of this simple shape was not such a simple matter. When a steel ingot was made one used a metal mould and a firebrick- or compo-lined head, and produced a sound ingot with the pipe confined to the head section. When a steel casting was made both the mould and the head portions were made of sand, and consequently other methods had to be adopted to produce a sound casting. The work reported by the Sub-Committee had tended to move in the only logical direction by increasing the insulation value of the head material.

In his opinion, the Sub-Committee were more than generous to the use

of feeding compounds when they said that they thought that insulation was at least as beneficial. Personally, he thought insulation more valuable than feeding compounds.

In the production of steel ingots there were various sources of inclusions —furnace slag, deoxidation products, reaction inclusions, ladle refractories and pit refractories. In steel castings there were all those inclusions and, in addition, the sand or compo with which the mould was made.

Mr. COUSANS was pleased to know that Dr. Cross was in agreement with the Sub-Committee in commencing with simple shapes, and that he looked with favour on their efforts to control directional solidification by using insulated feeding heads. There was undoubtedly a lot of work to be done on the study of inclusion types. Most foundrymen had encountered heats of steel of low ductility when the inclusions were of the grain-boundary type, and there was no doubt that a larger percentage of castings was rejected because of hot-tearing or pulls due to steel containing the unsatisfactory type of inclusion, than from steel of normal cleanliness and where the inclusions were of the spheroidal type.

Mr. WALKER said that in the Sub-Committee's records a large amount of information on the sands used had been reported. The Sub-Committee had summarised the data to provide the essential information for record purposes. A lot more information was available to support the conclusions given. He was afraid it would never be possible to say that such and such a sand was the best that any given foundry should employ. If Mr. Bailes took a ton of sand from another foundry and used it in his own foundry he would not get the same results. Every foundry would have to study its own conditions and decide which sand was the best for its own practice for any given casting.

Dr. DADSWELL said it was a good thing to keep careful records. Only a few months ago Mr. Basil Gray had been able to turn up the records of castings made several years ago by this Sub-Committee and use the information as a basis for further experiments.

CORRESPONDENCE.

Mr. G. M. MENZIES (North British Steel Foundry, Ltd., Bathgate, Scotland) wrote : I have read this Report with the greatest possible interest and think a very great deal of interesting work has been done.

I am particularly interested in the Parts on hot-tearing, although I think it is rather contradictory that it is admitted that the different processes may involve different sulphur limits for the same amount of "tearability," if there is such a word, while later on in Table XXVI. it is sought to find a relationship between the composition and the percentage of castings torn without stipulating what type of melting unit is utilised.

The Part on the spinner gate is extremely interesting. Concerning Part 3, dealing with feeding compounds, I am wondering if it would be possible to give information regarding the various proprietary compounds sold as feeding powders. We would very much like to try out impure blacklead with 40–50% of ash which is recommended, but as it is probably sold under some other name, there might be a very real difficulty in ascertaining the source of supply without direct communication with the Sub-Committee.

Dr. J. H. Whiteley (Consett Iron Co., Ltd., Consett, Co. Durham) wrote: In reading this Report I was particularly struck with Part 4, the section dealing with hot-tears, for it seems to have a bearing on the cognate problem of ingot cracking. To my mind Hall's * experiments were an illuminating contribution to that subject, since they demonstrated how seriously red-short steel can be at temperatures near the solidus. The Sub-Committee's view seems to be that the formation of hot-tears in the bulb diminishes as the casting temperature is raised, despite the wide variation of chemical composition, and that the test may be more of the nature of a fluidity test than a true hot-tear test. It is a well ascertained fact, however, that FeS in iron causes serious red-shortness, and the point should not be overlooked that a certain amount of that compound is always present in ordinary steel when in the liquidus–solidus range, even when the manganese is 1·0% or more. To observe the FeS a ladle sample should be taken, and it should be cooled to about 900° C. at a fairly rapid rate, especially when the manganese is high, for, given sufficient time in the temperature range between the solidus and about 900° C., all the FeS is converted to the dove-grey MnS or else to a somewhat lighter coloured dove-grey solid solution of FeS in MnS.

Now, it seems reasonable to infer that the actual amount of FeS present in the liquid steel and the rate at which it is subsequently changed to MnS will depend upon the sulphur and manganese contents; in other words, the speed of the reaction $FeS + Mn = MnS + Fe$ will be governed by the law of mass action and, of course, the diffusion rate of manganese. With that idea in mind I have calculated the ratios of manganese to sulphur in the steels given in Table XXIII. and have included them in ascending order in Table A. Although there is a minor discrepancy between steels *A* and *C*,

Table A.—*Relation of the Mn/S Ratio to Cracking.*

Steel.	Mn/S Ratio.	No. of Bars Pulled.
B	10	6
D	12	5
A	23·3	1, cracked
C	25·7	2
E	64·0	None
F	66·5	None

the Table indicates quite clearly that cracking is minimised as the Mn/S ratio increases. To put it in another way, cracking diminishes as the FeS present in the steel at and just below the solidus is reduced in quantity. The same thing appears when the results on the experimental casting † recorded by the main Committee are examined in that way. In the seven basic electric casts they used no cracking occurred when the Mn/S ratio was above 70. With the three acid open-hearth casts a ratio of about 45 was sufficient. This illustrates the Sub-Committee's observation that a much lower sulphur content is needed in basic electric than in acid open-hearth steel if pulls in casting are to be avoided, and I think it possible that this difference in behaviour may be due to the presence of a more refractory type of inclusion in the electric steel, for there is some reason to believe that such inclusions tend to induce cracking. For instance, Herty ‡ states

* Third Report of the Steel Castings Research Committee, Section IV., *The Iron and Steel Institute*, 1938, *Special Report No.* 23.

† First Report of the Steel Castings Research Committee, Tables V. and VI., *The Iron and Steel Institute*, 1933, *Special Report No.* 3.

‡ *U.S. Bureau of Mines*, 1931, *Technical Paper No.* 492.

that his experimental 3-in. square ingots of about 20 lb. weight were all extremely red-short when they contained silica as glassy inclusions—in fact, some of them could not be forged into 1-in. dia. bars—whereas those ingots in which the inclusions consisted of ferrous silicates high in FeO were quite satisfactory. Then, too, it is a common experience that the addition of aluminium or titanium to the ladle is liable to cause ingot cracking. There, again, refractory inclusions are formed.

Finally, I note that none of the castings with the 0·69-in. dia. bulb cracked. That, of course, may be a purely fortuitous result. On the other hand, it may be that there is an optimum rate of cooling whereby cracking is minimised, irrespective of the casting temperature. I have often observed that when ingots of different weights are teemed from the same ladle, one size may crack while another is relatively free from that defect.

SUB-COMMITTEE'S REPLY.

The Sub-Committee wrote in reply: The Members of the Sub-Committee are indebted to Mr. Menzies for his comments and are pleased to find that their work on hot-tearing has been of interest. A separate paper on the manufacture of some thin-walled castings and notes on the influence of pouring speeds has been prepared. This further work has shown that the susceptibility to hot-tear formation appears to be more directly related to the sulphur content than to the particular steelmaking process utilised. It will be appreciated that where the basic electric arc steelmaking process is used the resultant sulphur contents are likely to be much lower than in the case of steel made by the side-blown or acid open-hearth processes.

Dr. Whiteley's very interesting communication has been studied by the Members of the Sub-Committee. They hope to submit samples of castings containing serious hot-tears to the Metallurgical (Steel Castings) Sub-Committee of the Steel Castings Research Committee for examination and for the determination of types of inclusions. They are indebted to Dr. Whiteley for bringing out so clearly the relation of the manganese/sulphur ratio to cracking. They are very interested in the suggestion that it may be that there is an optimum rate of cooling whereby cracking is minimised irrespective of the casting temperature, and it is felt that this is of sufficient interest to justify the Sub-Committee doing some experimental work on it.

The comments of Dr. Angus are greatly appreciated by the Members of the Sub-Committee. They feel that a fracture, provided that it is obtained from the section and at the angle required, is equally as effective as an X-ray examination. X-ray examination is useful for the purpose of exploration, but, having found the defect, break-up shows the nature of the defect to best advantage. As far as the Members of the Sub-Committee are aware, no work has been reported on the variation of density, but it may be that a study of this aspect would be worth while and it may be possible to include this in a future investigation.

Dr. Northcott in his remarks raises some very interesting points. The Members of the Sub-Committee feel that, whilst Dr. Northcott's remarks regarding the effect of pouring speed on soundness may be valid as far as the pouring of ingots is concerned, they do not think that they have as much bearing on the soundness of steel castings. Provided that the feeding heads and runners are properly placed to obtain good directional solidification and freedom from sand erosion, the soundest castings should

be obtained from quickly poured steel. The pouring of steel castings slowly is almost bound to give rise to mis-running and cold laps. In the experience of the Members of the Sub-Committee many more defects are found in castings produced from steel lacking in fluidity or not poured quickly enough, than in castings made from hot steel or poured quickly.

The Members of the Sub-Committee have no experience in connection with the Durville method of casting and will certainly bear this process in mind in planning their future researches.

THE CONSTITUTION OF BASIC STEEL FURNACE SLAGS.*

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(Figs. 10 to 13 = Plates XLVI. to XLIX.)

Synopsis.

Systems of phase assemblages were deduced from the available phase-diagram data for basic electric reducing and oxidising slags and basic open-hearth slags. Accordingly, the constitutions of these slags were calculated and compared with the constitutions determined by the X-ray powder method. The agreement between the theoretical and observed results indicates that the systems of phase assemblages are essentially correct. Further confirmation was obtained by correlating the constitutions of a number of basic open-hearth slags as observed by Mason (*Journal of The Iron and Steel Institute*, 1944, No. II., p. 69 P) with the theoretical results deduced from his analysis. The slags are classified according to their constitutions, the chief characteristics discussed and the various phases described.

The occurrence of merwinite as the main phase in certain basic electric reducing slags and the mode of occurrence of fluorine and TiO_2 was also established. An X-ray method for the quantitative estimation of the amounts of phases was developed for reducing slags.

In electric oxidising slags the (Fe,Mn,Mg)O solid solution, the composition and amount of which vary from slag to slag, is a major phase. The lattice parameters of the RO phases were measured and correlated with their compositions, and the change of spacing during the oxidising stage was observed. An interesting feature is the appearance of a high-temperature form of $2CaO.SiO_2$, as distinct from the β form (denoted α'-$2CaO.SiO_2$), and of merwinite liable to contain manganese-merwinite in solution.

Basic open-hearth finishing slags from a number of steelworks were examined and their constitutions established. The chief phosphate phase is nagelschmidtite or fluorapatite, depending on whether fluorspar had been added or not. The solid solution (Fe,Mn,Mg)O occurs in all the slags and, in the more basic slags, co-exists with free (Ca,Mn)O. The co-existence of these two RO phases established by the X-ray method is believed to be a new discovery of considerable practical importance. The amount of the (Ca,Mn)O phase decreases with decreasing basicity until a limiting value is reached at which all the lime is in combination and only one RO phase (Fe,Mn,Mg)O remains. A practical X-ray method of slag control based upon the estimation of these two RO phases is proposed.

Introduction.

This paper is part of a major investigation, commenced in 1940, into the chemistry of the basic-steel melting processes involving the constitution of slags and refractory linings, the slag-metal, slag-refractory, and metal-refractory relationships. At this stage, however, the primary consideration is the constitution of the slags, and no attempt will be made to discuss the other aspects of the major investigation. It is considered that a knowledge of the chemistry of the crystallised slags is a primary essential to the study of the slags in the liquid state and to the major study of the chemistry of steelmaking. Furthermore, if the constitution of basic slags could be determined rapidly, a scientific control of the slag during the melting and refining process would be possible. A knowledge of the constitution of

* Received November 17, 1944.

basic slags would also render material assistance in the nationally important study of basic slags rich in phosphate.

When this investigation was approaching completion two papers dealing with the constitution of basic open-hearth slags were published. In the first White [2] surveyed the fundamental physico-chemical relationships governing the constitution and properties of open-hearth slags, and reviewed and attempted to correlate the available constitutional- and thermal-equilibrium diagrams with the available data on the constitution of slags. Consequently, reference should be made to White for previous publications on the constitution of slags, and only those phase diagrams will be discussed where the authors are not in agreement with White. In the second recent publication Mason [1] determined the constitution of fifteen basic open-hearth slags with low phosphate contents. The present investigation has a wider scope and provides much new information, since it deals with basic electric furnace slags from the Héroult and Greaves-Etchell types of furnaces, Bessemer slag and basic open-hearth slags. The basic open-hearth slags differ appreciably from those examined by Mason in that they contain more P_2O_3 and, in many cases, fluorine.

The Constituents of Basic Slags.

According to White [2] basic slag is derived principally from the oxidation of the charge, from the addition of slag-forming materials to the furnace and from the attack of the charge on the hearth refractory. The substances which find their way into the slag layer are CaO, FeO, MnO, MgO, SiO_2, P_2O_5, Fe_2O_3 and Al_2O_3. In addition, basic electric oxidising slags usually contain small amounts of Cr_2O_3. Slag chemistry is, therefore, concerned essentially with the relationships prevailing within systems of these oxides.

METHODS USED IN THE INVESTIGATION.

Chemical Analysis.

The chemical analyses of the various basic slags were carried out by the well-established procedure of silicate analysis, particular attention being paid to the influence of certain special constituents on the determination of others. Thus, in those slags containing fluorine the silica was determined from the various silica-bearing precipitates obtained during the determination of the fluorine by the method of Berzelius, as described by Hillebrand and Lundell,[3] with a modification introduced by Hoffman and Lundell;[4] the remainder of the constituents of these fluorine-bearing slags were determined after expulsion of the fluorine by fuming with perchloric acid in the usual way. The effect of the small amounts of sulphide sulphur on the ferrous-iron determination was not taken into account. Although a small amount of ferric iron is reduced under the conditions of the analysis, some ferrous iron is likely to become oxidised before the determination is completed. The error is therefore likely to fall more on the ferric than the ferrous side. All other constituents were determined in the usual way. Double dehydrations for the separation of the silica, and double precipitations for the ammonia group and for lime and magnesia were the rule throughout.

All the slag samples were crushed to pass a 90-mesh sieve and thoroughly cleaned with a magnet before analysis.

Methods for the Identification of Phases.

The majority of investigators have employed the petrological microscope for the identification of the phases in basic open-hearth slags. In his recent investigation Mason [1] employed a combination of methods, namely,

the examination of polished and thin sections, the determination of the refractive indices of silicate phases by the immersion method, the use of X-ray powder photographs and, in some cases, by the actual separation and chemical analysis of the phases present. The greater part of the present investigation was carried out with small spoon samples of slag, and, consequently, owing to their rapid rate of cooling, the slags were usually very fine-grained. The small crystal size and the intimate intertwining of the various crystal phases militated against the successful application of petrological methods.

The slags were therefore examined throughout by the X-ray method, which, in spite of certain limitations, was found extremely satisfactory. The technique was essentially the same as that employed previously by the authors in their study of dolomite refractories.[5] The chief limitation is the lower limit of visibility of X-rays, which, for phases producing faint patterns such as a number found in slags, can be relatively high. The authors do not, however, share Mason's view that identification becomes difficult or impossible in the presence of two or more phases with complex patterns, as many later examples will testify. Superposition of lines lowers the sensitivity of detection, but, provided that the relative intensities of standard patterns are estimated correctly (if necessary, by photometry), identification is generally possible. Measurement of interplanar spacings is often found valuable in cases of ambiguity but unnecessary in normal cases in which the constitution can be readily determined by matching against standard photographs.

All slags were examined by the Debye-Scherrer method, using a 9-cm. dia. camera and CoK_a radiation. The 19-cm. dia. camera was discarded in favour of the smaller one as the former's advantage of greater resolution was far outweighed by the latter's greater speed of operation and by the fact that all the information required for this survey could be obtained from the 9-cm. photographs. As some of the slags contained appreciable amounts of manganese, absorption had an adverse influence on the clarity of the photographs, but the CoK_a radiation had to be accepted as the best compromise between alternatives each presenting disadvantages. The specimens were prepared in the usual manner by mounting the powder on a hair by means of Canada balsam. The effect of absorption could often be reduced by dilution and by using a thin specimen. Experiments made to determine the optimum diameter for specimens showed that the other extreme, namely, a very thick (1¼ mm. instead of the usual ⅓ mm.) and slightly eccentric specimen, offers definite advantages by improving the detectability of weak lines at low orders. This method is not used, of course, for measuring accurate lattice parameters, and is only employed as a useful supplement to the normal method in a few special cases.

The information on slags obtained by the powder method includes the following:

(1) The physical state of the slag, *i.e.*, whether crystalline or amorphous. All the slags examined were largely crystalline and contained small amounts of glass.

(2) Identification of phases present.

(3) Relative amounts of phases. In the case of the basic electric reducing slags a quantitative method of estimation was developed.

(4) Lattice parameter measurements, in particular of the RO type of phases, may be correlated with the composition of the solid solution between the oxides FeO, MgO, MnO and CaO, assuming the validity of Vegard's law or an equivalent known calibration curve.

(5) The degree of homogeneity of the phases, or the range of composition, made manifest by the relative line breadth.

The X-ray results were correlated with the available phase-diagram data as in the dolomite investigation.[5]

Standard Compounds.

The preparations of $3CaO.SiO_2$, β- and γ-$2CaO.SiO_2$, $3CaO.Al_2O_3$, $5CaO.3Al_2O_3$, brownmillerite ($4CaO.Al_2O_3.Fe_2O_3$) and its isomorph $2CaO.Fe_2O_3$, CaO and MgO have been previously described.[5] The description of the preparation of standard compounds will be restricted to those compounds actually observed, but it should be emphasised that the slags were also examined for many other phases.

(1) *FeO and MnO.*—Although pure FeO does not exist, pure iron-rich wüstite solid solutions were obtained as the innermost layer of scale formed on Armco iron oxidised at 1000° C.[6]

MnO was prepared by heating manganese oxalate to 1000° C. in an atmosphere of hydrogen.

(2) *Spinel.*—Pure $MgO.Fe_2O_3$, $MgO.Al_2O_3$ and $MgO.Cr_2O_3$ were prepared by thoroughly mixing MgO with the correct proportions of Fe_2O_3, Al_2O_3 and Cr_2O_3, respectively, and firing to about 1600° C. The spinels observed in basic slags correspond, in general, to solid solutions of the type $(Fe,Mn,Mg)O.(Fe,Cr,Al)_2O_3$.

(3) *Calcium Fluoride.*—A sample of pure natural fluorspar was used. The spacing of the cubic lattice agreed satisfactorily with those found in the literature ($a_0 = 5{\cdot}453$ Å.).

(4) *Calcium Hydroxide.*—$Ca(OH)_2$ was prepared by hydrating pure lime.

(5) *Merwinite* ($3CaO.MgO.2SiO_2$).—Pure $CaCO_3.MgO$ and silica were mixed in the correct proportions and fired (*a*) at 1500° C. and water-quenched, (*b*) at 1500° C. and air-cooled and (*c*) at 1580° C. and water-quenched. The compound was formed in each case, the difference in the rate of cooling apparently having had no effect. Certain merwinitic basic electric reducing slags have identical X-ray patterns to these products, so that they too can, in practice, serve as standards.

The manganese analogue of merwinite—$3CaO.MnO.2SiO_2$—was prepared by firing a mixture of β-$2CaO.SiO_2$ and $2MnO.SiO_2$ at 1300° C. in hydrogen and rapidly cooling.[7]

(6) *Melilite.*—Melilite is the name given to a member of the isomorphous series with the end-members gehlenite ($2CaO.Al_2O_3.SiO_2$) and akermanite ($2CaO.MgO.2SiO_2$). The compound $2CaO.Al_2O_3.SiO_2$ was prepared by firing pure $CaCO_3$, Al_2O_3 and SiO_2 in the appropriate proportions to 1550° C.

(7) *Monticellite* ($CaO.MgO.SiO_2$).—A mixture of pure $CaCO_3$, MgO and SiO_2 was fired below 1500° C. The product was thoroughly crushed, moulded and fired below 1500° C., this process being repeated until the product was homogeneous.

(8) *Hilgenstockite* ($4CaO.P_2O_5$).—Pure $CaCO_3$ and P_2O_5 in the correct proportions were fired to 1700° C. and cooled rapidly.

(9) *Steadite* ($3CaO.P_2O_5$).—Commercial tricalcium phosphate was found to contain a large amount of hydroxy-apatite and only a small amount of $3CaO.P_2O_5$. Pure $3CaO.P_2O_5$ was prepared by firing a mixture of $CaCO_3$ and P_2O_5, in the ratio of three molecules of CaO to one of P_2O_5 at 1700° C., and cooling rapidly.

(10) *Nagelschmidtite* ($9CaO.P_2O_5.3SiO_2$–$7CaO.P_2O_5.2SiO_2$).—The compounds $9CaO.P_2O_5.3SiO_2$ and $7CaO.P_2O_5.2SiO_2$ were prepared by firing mixtures of $CaCO_3$, P_2O_5 and SiO_2 in the correct proportions to 1600° C. Their X-ray patterns were identical, a fact which was previously noted by Nagelschmidt.[8] There appears to be a solid-solution range of this structure

in the quasi-binary system $2CaO.SiO_2$–$3CaO.P_2O_5$, a recent paper by Barrett and McCaughey [9] providing confirmation.

(11) *Silico-carnotite* ($5CaO.P_2O_5.SiO_2$).—This compound was prepared by firing a mixture of $CaCO_3$, P_2O_5 and SiO_2 in the molecular proportions 5 : 1 : 1 to 1600° C.

(12) *Fluorapatite* ($9CaO.CaF_2.3P_2O_5$).—A pure natural apatite mineral was employed as a standard. The hydroxy-apatite is isomorphous with the fluorapatite, and the series may be written as follows :

Fluorapatite.—$Ca_{10}F_2(PO_4)_6$ or $9CaO.CaF_2.3P_2O_5$.
Hydroxy-apatite.—$Ca_{10}(OH)_2(PO_4)_6$ or $9CaO.Ca(OH)_2.3P_2O_5$.

Abbreviated Symbols Used in the Paper.

For convenience the following abbreviated symbols may be used in the text and diagrams to replace the more cumbersome formulæ for certain compounds :

Abbreviations for Chemical Formulæ.

C = CaO.
A = Al_2O_3.
S = SiO_2.
M = MgO.
P = P_2O_5.
F = Fe_2O_3.
RO = (Fe, Mg, Mn)O.
R_2O_3 = $(Fe, Al, Cr)_2O_3$.

Abbreviations in the Photometer Curves.

γ = γ-$2CaO.SiO_2$.
α = α-$2CaO.SiO_2$.
T = $3CaO.SiO_2$.
F = CaF_2.
M = MgO.
E = Merwinite.
L = Lime.
S = Spinel.
R = RO = (Fe, Mg, Mn)O.
H = $2CaO.Fe_2O_3$.
N = Nagelschmidtite.
O = Silico-carnotite.
P = Fluorapatite.

Basic Electric-Furnace Slags.

The usual basic electric steelmaking process is carried out in two distinct stages. In the first, or oxidising stage, the silicon in the charge is oxidised-the resulting SiO_2 combining with the added lime to form a calcium-silicate slag which contains appreciable amounts of iron oxides, manganese oxide, and smaller amounts of MgO, Al_2O_3 and P_2O_5. In addition, when alloy scrap is charged, the slag contains variable amounts of oxides of certain alloying elements, notably chromium. This oxidising slag is removed at the end of the first stage and the reducing or finishing slag is made by the addition of limestone (or lime), sand and fluorspar. The iron-oxide content is maintained at the minimum by the addition of anthracite and powdered ferro-silicon to the slag. This reducing slag consists mainly of CaO and SiO_2, with smaller amounts of MgO, Al_2O_3 and fluorine, and very small amounts of iron and manganese oxides and sulphur.

Although from the order in which the process is carried out it would be more logical to deal with the constitution of the oxidising slags before proceeding with the reducing slags, the latter slags are much less complex, and, accordingly, will be considered first.

The Constitution of Basic Electric Reducing Slags.

Table I. gives the chemical compositions of a considerable number of basic electric reducing slags from a variety of steel qualities. In

TABLE I.—*Chemical Compositions of*

Slag No.	Chemical Analysis.													
	SiO_2. %.	FeO. %.	Fe_2O_3. %.	Al_2O_3 and TiO_2. %.	MnO. %.	CaO. %.	MgO. %.	CaF_2. %.	S. %.	CaC_2. %.	CaO.	FeO.	MnO.	MgO.
1	}										Accurate analyses not obtained			
2	}													
3	22·10	‡	0·56	1·54	0·34	62·30	7·50	...	...	...	...	0·46	0·34	7·50
4	19·40	1·48	0·20	1·93 §	0·55	54·20	16·78	...	...	...	...	1·48	0·55	16·8
5 (*a*)	25·20	0·27	0·26	1·64	0·11	64·10	0·94	6·5	0·34	0·24	...	0·27	0·11	0·94
5 (*b*)	26·60	0·34	0·42	1·30	0·15	59·50	1·56	7·50	0·47	1·73	...	0·34	0·15	1·56
5 (*c*)	27·70	0·54	0·28	1·42	0·18	59·94	0·94	5·80	0·57	1·25	...	0·54	0·18	0·94
6 (*a*)	25·20	0·58	0·32	2·04	0·48	59·02	2·17	9·20	0·27	...	...	0·58	0·48	2·17
6 (*b*)	28·60	0·36	0·24	2·00	0·43	60·38	2·21	5·20	0·23	...	...	0·36	0·43	2·21
7	26·20	...	0·32	2·25	0·01	60·79	4·81	5·60	0·48	0·77	...	...	0·01	4·81
8	25·70	0·54	0·36	2·14	0·26	59·70	4·66	6·70	0·06	...	...	0·54	0·26	4·66
9 (*a*)	24·44	1·36	0·40	1·91	0·60	58·22	6·82	7·08	0·35	...	...	1·36	0 60	6·82
9 (*b*)	28·04	0·34	0·14	1·87	0·11	61·42	4·05	4·28	0·25	...	...	0·34	0·11	4·05
10	27·10	0·52	0·14	2·48	0·23	61·88	1·95	4·60	0·64	...	...	0·52	0·23	1·95
11 (*a*)	27·80	0·29	0·48	1·91	0·28	62·29	2·35	3·90	0·34	...	...	0·29	0·28	2·35
11 (*b*)	33·20	0·23	0·38	2·22	0·14	60·21	0·87	1·70	0·28	...	...	...	...	...
12	26·90	0·77	0·36	1·86	0·54	59·80	5·17	5·00	0·30	...	...	0·77	0·54	5·17
13 (*a*)	25·00	0·66	0·30	3·20	0·17	55·45	3·49	11·08	0·34	0·27	...	0·66	0·17	3·49
13 (*b*)	30·12	0·24	0·14	3·76	0·03	52·86	6·36	6·64	0·08	0·66	...	0·24	0·03	2·39
14	27·68	0·72	0·40	5·47	0·41	60·00	3·23	...	0·014	0·60	...	0·72	0·41	3·23
15 (*a*)	27·60	0·34	0·26	2·34	0·23	59·04	5·36	5·30	0·32	...	...	0·34	0·23	5·36
15 (*b*)	30·50	0·20	0·18	2·18	0·11	59·52	3·00	3·40	0·25	...	...	0·20	0·11	3·00
16	25·80	0·58	0·16	3·10	0·25	53·88	7·07	9·00	0·38	...	...	0·58	0·25	7·07
17 (*a*)	27·90	0·50	0·48	1·26	0·64	57·06	3·33	6·70	0·24	...	...	0·50	0·64	3·33
17 (*b*)	28·50	0·74	0·30	1·58	0·28	57·56	4·02	5·80	0·11	...	...	0·74	0·28	4·02
18	26·95	1·38	0·46	3·62	0·82	54·79	5·10	6·70	0·28	...	...	1·38	0·82	5·10
19	27·20	‡	1·20	2·00	0·44	54·60	9·48	4·60	...	...	...	1·00	0·44	9·48
20	30·60	0·36	0·24	1·16	0·38	61·00	1·67	4·50	0·21	...	...	0·36	0·38	1·67
21	27·00	1·12	0·36	2·30	0·44	51·03	10·5	6·40	0·27	...	...	1·12	0·44	9·52
22	29·70	1·14	0·40	1·72	0·58	56·48	5·54	3·80	0·62	...	...	1·14	0·58	4·76
23	27·90	1·15	0·56	1·96	0·52	52·48	8·80	5·70	0·22	...	...	1·15	0·52	7·90
24 (*a*)	27·40	0·61	0·20	2·36	0·58	54·13	6·22	7·70	0·32	...	...	0·61	0·58	6·22
24 (*b*)	29·40	1·13	0·26	2·62	0·67	55·20	6·01	4·00	0·19	...	...	1·13	0·67	4·91
25	28·20	1·49	0·64	1·68	0·72	52·30	10·75	3·40	0·15	...	...	1·49	0·72	9·94
26	29·20	0·86	0·20	1·04	2·44	52·93	6·66	5·60	0·23	...	...	0·86	2·49	4·75
27	32·00	0·65	0·08	2·40	0·87	57·54	3·00	3·40	0·08	...	...	0·65	0·87	0·33
28	30·10	1·42	0·84	1·90	0·40	53·99	6·12	4·00	0·12	...	...	1·42	0·40	1·63
29	30·20	1·12	0·30	2·58	0·43	53·57	9·14	3·00	0·16	...	...	1·12	0·43	5·83
30	31·80	1·33	0·28	1·58	0·43	56·15	7·75	...	0·09	...	...	1·33	0·43	4·73
31	32·00	0·92	0·18	2·25	0·92	55·50	5·07	3·40	0·15	...	...	0·92	0·92	0·88
32	33·30	0·70	0·48	1·82	0·47	56·73	5·53	...	0·10	...	...	0·70	0·47	0·67
33 (*a*)	34·60	1·15	0·16	3·15	0·74	44·28	14·0	...	Tr.	...	...	...	...	...
33 (*b*)	25·40	1·33	0·52	9·40 ‖	0·68	43·00	14·76	...	0·014	...	...	1·35	0·68	3·19
34	32·80	0·83	0·88	2·04	0·50	54·85	6·95	...	0·03	...	...	0·83	0·50	1·28
35	32·00	1·15	1·44	2·40	0·81	53·09	8·51	...	0·06	...	...	1·15	0·81	2·34
36	33·00	0·88	0·64	2·10	0·94	53·91	6·60	...	0·11	...	...	0·88	0·94	0·06
37 (*a*)	30·40	0·56	0·18	2·97	0·28	49·99	9·55	6·60	0·33	...	...	0·56	0·28	3·05
37 (*b*)	32·90	0·48	0·34	5·75	0·60	45·55	11·9	1·30	0·25	...	...	...	...	...
38 *	31·60	1·55	1·20	1·83	0·48	53·36	5·64	3·30	0·15	...	...	1·55	0·48	0·78
39 †	32·20	‡	1·60	1·40	1·47	51·90	3·87	7·80	Tr.	...	...	...	...	...
40	30·80	0·77	0·16	1·96	0·74	49·14	7·46	7·20	0·15	...	...	0·77	0·74	0·43
41	30·80	1·08	0·24	2·93	0·85	48·81	11·47	4·30	0·18	...	...	1·08	0·85	3·77
42	32·80	0·90	1·40	2·60	0·54	50·40	9·47	...	0·25	...	...	0·90	0·54	0·16
43	32·30	0·52	0·30	2 02	0·10	46·41	11·98	4·70	0·30	...	...	0·52	0·10	0·63
44	32·40	1·73	1·28	2 70	0·62	42·80	15·84	...	0·29	...	...	1·73	0·62	1·37

* $MgO.Cr_2O_3$, 1·20.

† $3CaO.MgO.2SiO_2$, 22·0%. $3CaO.MnO.2SiO_2$, 7·5%.

Basic Electric Reducing Slags.

Constitution. Weight % (calculated) of—															
$3CaO.SiO_2$.	$2CaO.SiO_2$.	$3CaO.MgO.2SiO_2$.	$3CaO.2TiO_2$.	$CaO.MgO.SiO_2$.	$2CaO.MgO.2SiO_2$.	$2CaO.Al_2O_3.SiO_2$.	$3CaO.Al_2O_3$.	$5CaO.3Al_2O_3$.	$MgO.Fe_2O_3$.	$MgO.Al_2O_3$.	$4CaO.Al_2O_3.Fe_2O_3$.	CaF_2.	CaS.	CaC_2.	CaO/SiO_2 Ratio.
owing to high CaF_2 content.															
74·9	6·9	...	...	...	...	...	3·97	...	...	...	0·30	5·6	...	...	2·82
60·1	10·3	...	...	...	...	...	4·77	...	...	...	0·60	8·0	...	...	2·80
55·8	30·2	...	...	...	...	...	3·89	...	...	...	0·79	6·5	0·74	0·24	2·55
27·4	35·6	...	...	...	...	...	2·73	...	...	...	1·28	7·5	1·05	1·73	2·23
19·5	64·7	...	...	...	...	...	3·28	...	...	...	0·85	5·8	1·27	1·25	2·15
32·6	47·7	...	...	...	...	...	4·87	...	...	...	0·97	9·2	0·61	...	2·35
13·0	72·2	...	...	...	...	...	4·90	...	...	...	0·73	5·2	0·51	...	2·10
29·3	53·1	...	...	...	...	...	5·43	...	...	...	0·97	5·6	1·08	0·77	2·32
33·1	48·6	...	...	...	...	...	5·06	...	...	...	1·09	6·7	0·13	...	2·30
35·4	43·4	...	...	...	...	...	4·37	...	...	...	1·22	7·1	0·78	...	2·30
22·4	63·5	...	...	...	...	...	4·72	...	...	...	0·43	5·6	0·55	...	2·19
27·1	57·3	...	...	...	...	...	5·68	...	...	...	0·43	4·6	1·44	...	2·26
26·3	59·9	...	...	...	...	...	4·24	...	...	...	1·46	3·9	0·76	...	2·24
...	84·0	3·75	...	...	2·79	5·97	...	...	...	...	...	1·7	0·63	...	1·81
23·8	59·2	...	...	...	...	...	4·32	...	...	...	1·09	5·0	0·68	...	2·22
10·9	63·4	...	...	...	...	...	8·13	...	...	...	0·85	11·1	0·76	0·27	2·22
...	65·2	20·1	...	...	...	...	...	...	0·18	5·25	...	6·6	0·18	0·66	1·75
...	79·4	...	...	...	...	...	10·8	2·2	...	...	1·22	...	0·03	0·60	2·16
12·2	70·0	...	...	...	...	...	5·8	...	...	...	0·78	5·3	0·72	...	2·14
...	87·5	...	...	...	...	...	...	3·97	...	...	0·54	3·4	0·55	...	1·95
...	73·9	...	...	...	...	...	7·95	...	...	...	0·48	9·0	0·85	...	2·09
9·35	72·9	...	...	...	...	...	2·54	...	...	...	1·45	6·7	0·54	...	2·05
5·82	77·3	...	...	...	...	...	3·70	...	...	...	0·91	5·8	0·25	...	2·02
...	77·3	...	...	...	...	...	0·53	6·0	...	...	1·39	6·7	0·63	...	2·02
1·63	76·8	...	...	...	...	...	5·03	...	...	...	0·60	4·6	...	...	2·00
6·10	83·1	...	...	...	...	...	2·67	...	...	...	0·73	4·5	0·46	...	1·99
...	77·4	...	...	...	...	...	...	...	0·33	3·21	0·30	6·4	0·61	...	1·90
...	85·1	...	...	...	...	...	...	...	0·50	2·40	...	3·8	1·39	...	1·90
...	80·0	...	...	...	...	...	...	...	0·70	2·73	...	5·7	0·49	...	...
...	78·6	...	...	...	...	...	...	4·30	...	...	0·61	7·7	0·72	...	1·98
...	84·3	...	...	...	...	...	...	...	0·33	3·65	...	4·0	0·42	...	1·88
...	80·3	...	...	...	...	...	...	...	0·80	2·33	...	3·4	0·33	...	1·84
...	71·4	11·8	...	...	...	...	...	...	0·25	1·45	...	5·6	0·53	...	1·81
...	77·2	13·9	...	...	...	...	...	...	0·10	3·35	...	3·4	0·18	...	1·80
...	71·2	16·2	...	...	...	...	...	...	1·05	2·65	...	4·0	0·27	...	1·79
...	67·6	18·1	...	...	...	...	...	...	0·37	3·60	...	3·0	0·36	...	1·77
...	71·3	19·0	...	...	...	...	...	...	0·35	2·20	...	...	0·19	...	1·77
...	64·0	26·5	...	...	...	...	...	...	0·23	3·14	...	3·4	0·34	...	1·73
...	61·1	32·8	...	...	...	...	...	...	0·60	2·53	...	...	0·22	...	1·70
...	...	60·0	...	24·1	3·4	8·5	...	...	...	...	...	...	...	...	1·28
...	6·75	63·0	9·92	...	...	...	...	...	0·65	13·4	...	...	0·03	...	1·69
...	54·4	37·9	...	...	...	...	...	...	1·10	2·85	...	...	0·07	...	1·67
...	50·2	39·6	...	...	...	...	...	...	1·80	3·35	...	...	0·13	...	1·66
...	47·2	45·2	...	...	...	...	...	...	0·80	2·93	...	...	0·24	...	1·64
...	42·1	43·1	...	...	...	...	...	...	0·23	4·14	...	6·6	0·74	...	1·64
...	...	65·6	...	14·8	...	15·2	...	...	...	0·40	...	...	0·55	...	1·38
...	55·9	33·2	...	...	...	...	...	...	1·50	2·54	...	3·3	0·33	...	1·61
...	49·6	29·5	...	...	8·02	3·76	...	...	...	...	...	7·8	...	...	1·61
...	35·1	51·0	...	...	...	...	...	...	0·20	2·73	...	7·2	0·35	...	1·60
...	32·9	52·9	...	...	...	...	...	...	0·30	4·08	...	4·3	0·40	...	1·58
...	24·9	65·8	...	...	...	...	...	...	1·75	3·63	...	...	0·56	...	1·54
...	4·0	84·6	...	...	...	...	...	...	0·38	2·82	...	4·7	0·67	...	1·43
...	...	70·6	...	17·2	...	...	...	...	1·60	3·76	...	...	0·65	...	1·32

‡ Total iron reported as Fe_2O_3.   § TiO_2 0·13%.   ‖ TiO_2 3·60%.

this and subsequent Tables letters suffixed to slag numbers indicate the following:

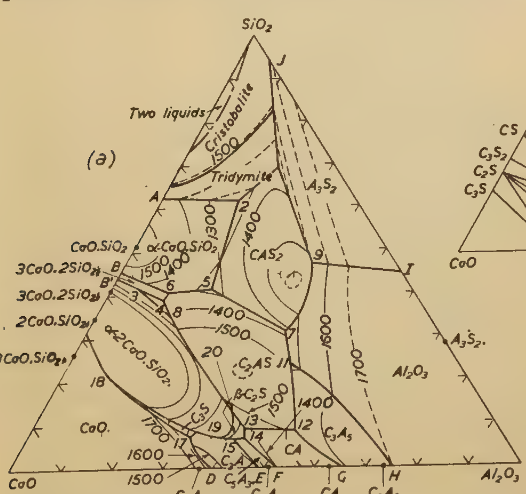
(a) Sample of slag taken at the melted stage.
(b) Sample of slag taken just prior to slag-off, when there are only two samples, or during the boil, when there are more than two samples.
(c) Sample of slag taken just prior to slag-off.

The main components of these slags are CaO and SiO_2, with small but significant amounts of MgO and Al_2O_3, these four oxides comprising more than 90% of the total oxides present. In addition, significant amounts of fluorine are present in many of the slags. The iron oxides, MnO and sulphur, are minor components. It is therefore possible to consider basic electric reducing slags as having compositions within the quaternary system $CaO–MgO–SiO_2–Al_2O_3$ when fluorspar is absent; as will be shown later, constitutions of finishing slags containing fluorspar may also be calculated from this quaternary system.

The Quaternary System $CaO–MgO–SiO_2–Al_2O_3$.

The phase equilibria in this quaternary system have not been worked out, although the bounding ternary systems and parts of the quaternary system have been determined. White[2] reviewed the available data and tentatively proposed a quaternary phase diagram, but one of the authors * of the present paper recently investigated part of this system and his phase diagram is not in agreement with that appearing in White's paper; the system will therefore be re-considered.

The complete equilibria for the ternary system $CaO–SiO_2–Al_2O_3$ were determined by Rankin and Wright.[10] The equilibrium diagram showing the primary fields of crystallisation and the several invariant points is



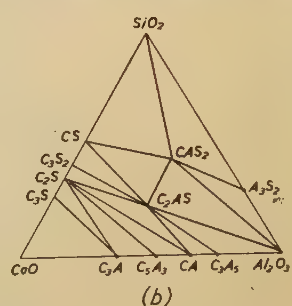


FIG. 1.—(a) Thermal Equilibrium Diagram of the System $CaO–SiO_2–Al_2O_3$, according to Rankin and Wright.[10] (b) Phase Distribution in the System $CaO–SiO_2–Al_2O_3$.

shown in Fig. 1. The $CaO–SiO_2–Al_2O_3$ system, which has frequently been described in connection with blast-furnace slags and cement chemistry,

* J. R. Rait, unpublished work.

need not be described in detail here. It is interesting to note that, although $3CaO.SiO_2$ is not a primary phase in the binary system $CaO–SiO_2$, it has a long thin area of primary crystallisation in the ternary, tapering to an invariant point at 1900° C., where it decomposes into CaO and $2CaO.SiO_2$. It also forms invariant points with CaO, $3CaO.Al_2O_3$ and with $2CaO.SiO_2$, $3CaO.Al_2O_3$ at 1470° C. and 1455° C., respectively. The ternary compound gehlenite ($2CaO.Al_2O_3.SiO_2$) melts congruently at 1590° C.

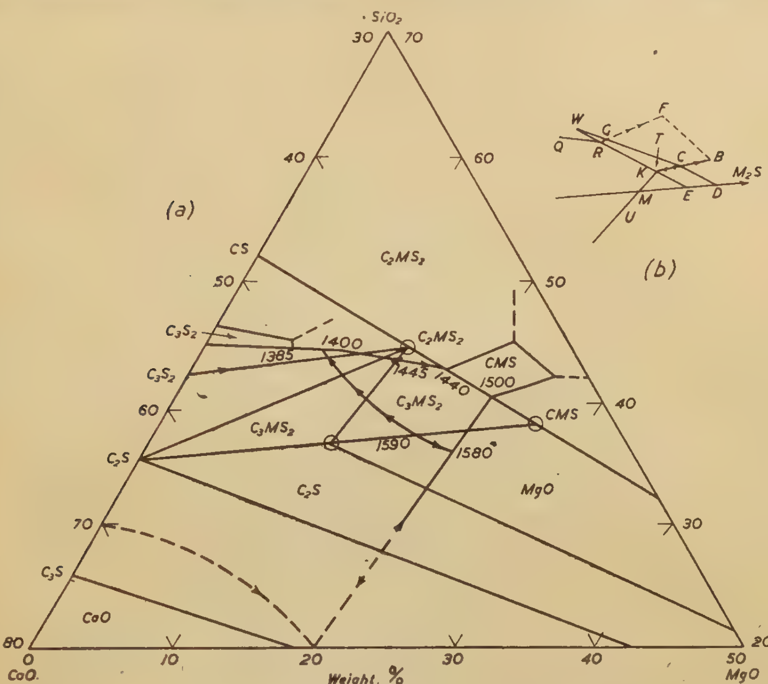


FIG. 2.—(*a*) Thermal Equilibrium Diagram of the System $CaO–MgO–SiO_2$, according to Ferguson and Merwin,[11] modified by Parker and Nurse.[13] (*b*) The Monticellite Field in the $CaO–MgO–SiO_2$ System, according to Ferguson and Merwin.[11]

The various groups of co-existing phases for crystallisation according to complete equilibrium are as follows :

(*a*) CaO, $3CaO.SiO_2$, $3CaO.Al_2O_3$.
(*b*) $3CaO.SiO_2$, $2CaO.SiO_2$, $3CaO.Al_2O_3$.
(*c*) $2CaO.SiO_2$, $3CaO.Al_2O_3$, $5CaO.3Al_2O_3$.
(*d*) $2CaO.SiO_2$, $5CaO.3Al_2O_3$, $CaO.Al_2O_3$.
(*e*) $2CaO.SiO_2$, $CaO.Al_2O_3$, $2CaO.Al_2O_3.SiO_2$.
(*f*) $2CaO.SiO_2$, $3CaO.2SiO_2$, $2CaO.Al_2O_3.SiO_2$.
(*g*) $3CaO.2SiO_2$, $CaO.SiO_2$, $2CaO.Al_2O_3.SiO_2$.

The ternary system $CaO–MgO–SiO_2$ is of considerable complexity. Ferguson and Merwin[11] reported no fewer than four ternary compounds, $CaO.MgO.2SiO_2$ (diopside), $CaO.MgO.SiO_2$ (monticellite), $2CaO.MgO.2SiO_2$ (akermanite) and $5CaO.MgO.2SiO_2$, six ternary eutectics and considerable solid solution. Recently, however, a fifth ternary compound

$3CaO.MgO.2SiO_2$ (merwinite) has been identified,[12] thus necessitating modifications in the original equilibrium diagram. Parker and Nurse[13] investigated the thermal behaviour of merwinite in the ternary system, and, accordingly, the diagram to be accepted is that due to Ferguson and Merwin[11] modified by the data of Parker and Nurse.* Consideration will only be given to the low-silica portion of the system appropriate to basic slags. The boundary curves and directions of falling temperature are shown in Fig. 2. The relationship of the merwinite field to the complete

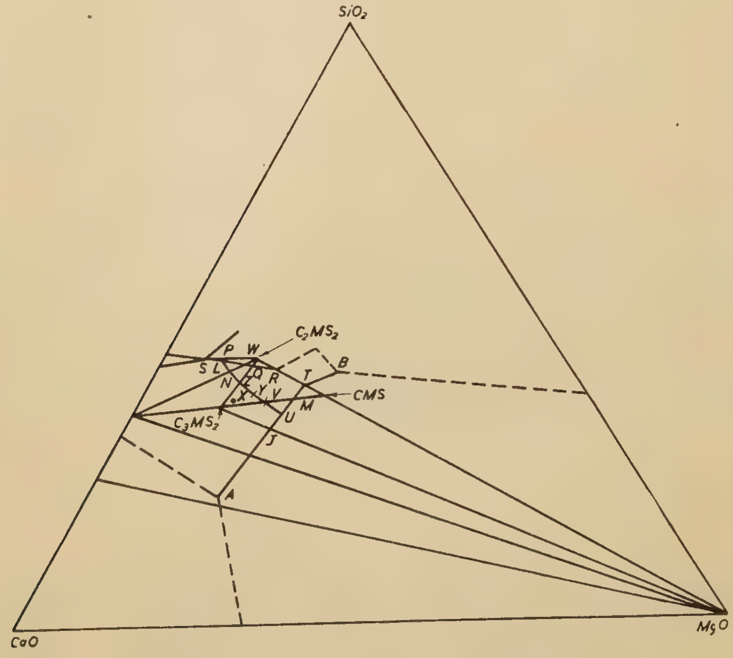


Fig. 3.—Phase Distribution in the System CaO–MgO–SiO_2.

system and the phase groupings for complete crystallisation are shown in Fig. 3: *A* is the ternary eutectic between CaO, MgO and α-$2CaO.SiO_2$; the temperature is not known accurately, but is higher than 1900° C. *B* is the invariant point between MgO, monticellite solid solution and $2MgO.SiO_2$ at 1502 ± 5° C. *R* is the ternary eutectic $2CaO.MgO.2SiO_2$, $3CaO.MgO.2SiO_2$ and monticellite solid solution at 1440° C. *T* is the invariant point MgO.3CaO, $MgO.2SiO_2$, monticellite solid solution at 1500° C. *P* is the invariant point $2CaO.MgO.2SiO_2$, $3CaO.MgO.2SiO_2$, $2CaO.SiO_2$ at 1400° C. *U* is the invariant point $3CaO.MgO.2SiO_2$, $2CaO.SiO_2$, MgO at 1580° C.

On the boundary curve between $3CaO.MgO.2SiO_2$ and $2CaO.MgO.2SiO_2$ there is a temperature maximum (1445° C.) at the intersection of the line

* Since the publication of the paper the authors' attention has been drawn to a later investigation of the CaO–MgO–SiO_2 system by Osborn (*Journal of the American Ceramic Society*, 1943, vol. 26, p. 321). The essential modification is in the position of the point *U*. Osborn's composition for *U* is CaO 43·0%, MgO 18·2%, SiO_2 38·8% at 1575 ± 5° C., and this has been accepted by Parker and Nurse.

between merwinite and akermanite at Q. On the boundary curve between $2CaO.SiO_2$ and $3CaO.MgO.2SiO_2$ a maximum occurs at the intersection of the projection of the line $2CaO.SiO_2$–$3CaO.MgO.2SiO_2$ at V, which gives the decomposition temperature of merwinite at 1590° C.

Ferguson and Merwin [11] explained the thermal behaviour of monticellite with the aid of the diagram shown in Fig. 2 (*b*). There is considerable solid solution of forsterite in monticellite (about 10%), and, on cooling a mix with the monticellite composition, instead of obtaining pure $CaO.MgO.SiO_2$ a mixture of monticellite solid solution and another phase were obtained. Ferguson and Merwin [11] considered this other phase to be α-$2CaO.SiO_2$, but according to the new data it is $3CaO.MgO.2SiO_2$.

The various groups of co-existing phases for crystallisation according to complete equilibrium are as follows :

(*a*) CaO, MgO, $3CaO.SiO_2$.
(*b*) $3CaO.SiO_2$, $2CaO.SiO_2$, MgO.
(*c*) $2CaO.SiO_2$, $3CaO.MgO.2SiO_2$, MgO.
(*d*) $3CaO.MgO.2SiO_2$, $CaO.MgO.SiO_2$, MgO.
(*e*) $CaO.MgO.SiO_2$, $2MgO.SiO_2$, MgO.
(*f*) $2CaO.SiO_2$, $3CaO.MgO.2SiO_2$, $2CaO.MgO.2SiO_2$.
(*g*) $3CaO.MgO.2SiO_2$, $CaO.MgO.SiO_2$, $2CaO.MgO.2SiO_2$.

There is no primary-phase field of $3CaO.SiO_2$ in this system owing to the temperature of the ternary eutectic A being higher than the dissociation temperature of $3CaO.SiO_2$. Parker and Nurse [13] pointed out that " disappearing phases " frequently occur in this system owing to the shape and position of the primary-phase fields in relation to the triangles representing co-existing phases on final crystallisation (Fig. 3). For example, compositions in the area C_3MS_2–N–V first crystallise $2CaO.SiO_2$ on cooling, this compound becoming a disappearing phase. Consider the behaviour of the composition X, Fig. 3, which has a final melting point of about 1750° C. On cooling below this temperature $2CaO.SiO_2$ begins to crystallise and the composition of the liquid will be given by the intersection of the line $2CaO.SiO_2$–X with the contour representing the temperature reached. At 1580° C. this interesection will be at Y, on the $2CaO.SiO_2$–$3CaO.MgO.2SiO_2$ boundary. On further cooling $2CaO.SiO_2$ will re-dissolve, while $3CaO.MgO.2SiO_2$ separates, the trace of the liquid composition being the boundary curve $2CaO.SiO_2$–$3CaO.MgO.2SiO_2$. The composition of the solid mixture of $2CaO.SiO_2$ and $3CaO.MgO.2SiO_2$ will be represented by points on the line $2CaO.SiO_2$–$3CaO.MgO.2SiO_2$ and determined by the point of intersection of the $2CaO.SiO_2$–$3CaO.MgO.2SiO_2$ line with a line drawn through the composition of the liquid and the original composition. At Z the line XZ passes through $3CaO.MgO.2SiO_2$; all the $2CaO.SiO_2$ will then have disappeared, only one solid phase, merwinite, remaining. The next phase change occurs when the line C_3MS_2–X produced intersects the boundary curve $3CaO.MgO.2SiO_2$–$2CaO.MgO.2SiO_2$. At this point merwinite and akermanite crystallise together, the liquid composition following the boundary curve until, at R, the melt becomes completely solid, the final composition thus being merwinite, akermanite and monticellite. Crystallisation paths may be followed out in a similar fashion for compositions within the various fields which provide further illustrations of the phenomenon of the disappearing phase. Further examples are $2CaO.SiO_2$ in the area $3CaO.MgO.2SiO_2$–U–MgO, MgO in the area $CaO.MgO.SiO_2$–T–M, $3CaO.MgO.2SiO_2$ in the area PWL, and others.

This type of crystallisation path (showing disappearing phases) may have an important influence on the resultant constitution of a slag, depending on its rate of cooling. Thus, a slag whose crystallisation path

has a disappearing phase can have different constitutions with different rates of cooling.

The equilibrium diagram of the ternary system $CaO–MgO–Al_2O_3$ shown in Fig. 4 is due to Rankin and Merwin.[14] There is no ternary compound in

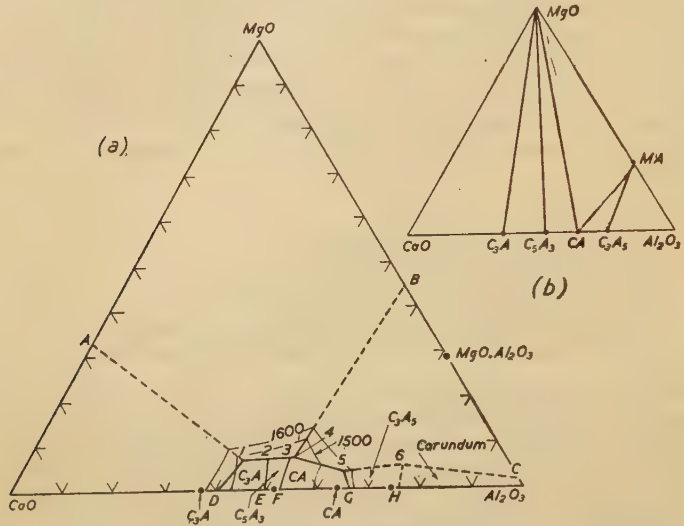


FIG. 4.—(a) Thermal Equilibrium Diagram of the System $CaO–MgO–Al_2O_3$, according to Rankin and Merwin.[14] (b) Phase Distribution in the System $CaO–MgO–Al_2O_3$.

this system. The phase regions for complete crystallisation are as follows:

(a) CaO, $3CaO.Al_2O_3$, MgO.
(b) $3CaO.Al_2O_3$, $5CaO.3Al_2O_3$, MgO.
(c) $5CaO.3Al_2O_3$, $CaO.Al_2O_3$, MgO.
(d) $CaO.Al_2O_3$, $MgO.Al_2O_3$, MgO.
(e) $CaO.Al_2O_3$, $3CaO.5Al_2O_3$, MgO.
(f) $3CaO.5Al_2O_3$, $MgO.Al_2O_3$, Al_2O_3.

The shape and position of some of the primary-phase fields in relation to the triangles which delimit the phases present on final crystallisation are such that the type of crystallisation path introduced a disappearing phase. For example, part of the primary field of lime projects into triangle $C_3A–C_5A_3$–MgO in Fig. 4(b), so that CaO is a disappearing phase in this area.

The equilibrium diagram of the ternary system $MgO–SiO_2–Al_2O_3$ investigated by Rankin and Merwin,[15] and later modified by Greig,[16] is reproduced in Fig. 5. There is one ternary compound, cordierite ($2MgO.2Al_2O_3.5SiO_2$), which is unlikely to be encountered in basic slags. The following are the phase regions appropriate to basic slags :

(a) MgO, $2MgO.SiO_2$, $MgO.Al_2O_3$.
(b) $2MgO.SiO_2$, $MgO.SiO_2$, $MgO.Al_2O_3$.

The binary solid-solution metasilicate series between diopside ($CaO.MgO.2SiO_2$) and clino-enstatite ($MgO.SiO_2$) has been described by Bowen,[17] and Ferguson and Buddington[18] have determined the thermal-equilibrium diagram of the melilite-solid-solution series between akermanite

($2CaO.MgO.2SiO_2$) and gehlenite ($2CaO.Al_2O_3.SiO_2$). The plane MgO–$2CaO$–SiO_2–$5CaO.3Al_2O_3$ was shown by Hansen[19] to be a simple ternary

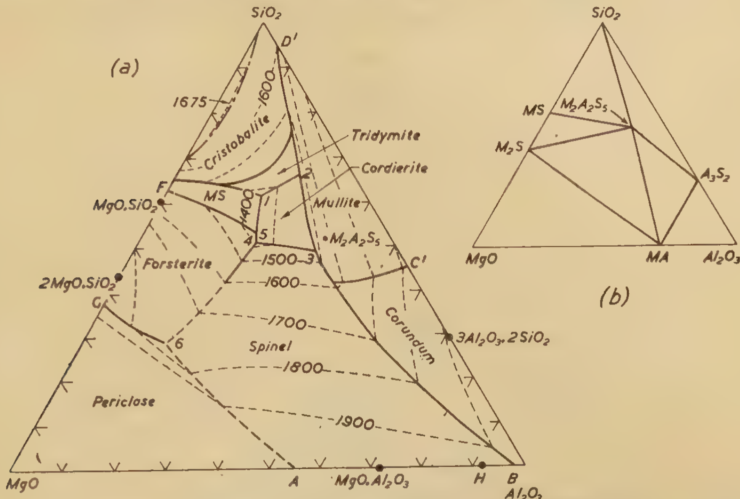


FIG. 5.—(*a*) Thermal Equilibrium Diagram of the System CaO–MgO–Al_2O_3, according to Rankin and Merwin.[15] (*b*) Phase Distribution in the System MgO–SiO_2–Al_2O_3.

eutectic system. The melting relationships in the partial quaternary system CaO–$2CaO$–SiO_2–$5CaO.3Al_2O_3$–MgO have been studied by McMurdie and Insley.[20] A noteworthy feature of this system (Fig. 6) is the occurrence of a volume of primary crystallisation of $3CaO.SiO_2$.

The phase equilibria of the complicated major quaternary system are not available, but an estimation of the smaller quaternary systems within the major system was given by McCaffery, Oesterle and Schapiro.[21] The compounds formed on the complete crystallisation of a slag depend upon the smaller quaternary system in which the slag composition falls, and it was shown[21] how a small change in composition may lead to a marked change in the constitution and viscosity of a slag.

White[2] applied the phase distribution proposed by McCaffery[21] in the study of the constitution of basic open-hearth slags, while Parker and Ryder[22] accepted this phase distribution, modified by the introduction of merwinite, in their investigations on "falling" blast-furnace slags.

In a separate investigation, one of the authors * of the present paper determined the phase distribution in part of the quaternary system as follows:

(*a*) CaO, MgO, $3CaO.SiO_2$, $3CaO.Al_2O_3$.
(*b*) $3CaO.SiO_2$, $2CaO.SiO_2$, $3CaO.Al_2O_3$, MgO.
(*c*) $2CaO.SiO_2$, $3CaO.Al_2O_3$, $5CaO.3Al_2O_3$, MgO.
(*d*) $2CaO.SiO_2$, $3CaO.MgO.2SiO_2$, $MgO.Al_2O_3$, MgO.
(*e*) $3CaO.MgO.2SiO_2$, $CaO.MgO.SiO_2$, $MgO.Al_2O_3$, MgO.
(*f*) $CaO.MgO.SiO_2$, $2MgO.SiO_2$, $MgO.Al_2O_3$, MgO.
(*g*) $2CaO.SiO_2$, $5CaO.3Al_2O_3$, $CaO.Al_2O_3$, MgO.
(*h*) $2CaO.SiO_2$, $CaO.Al_2O_3$, $2CaO.Al_2O_3.SiO_2$, $MgO.Al_2O_3$.
(*i*) $2CaO.SiO_2$, $3CaO.MgO.2SiO_2$, $2CaO.MgO.2SiO_2$, $2CaO.Al_2O_3.SiO_2$.

* J. R. Rait, unpublished work.

(j) $2CaO.SiO_2$, $2CaO.MgO.2SiO_2$, $CaO.SiO_2$, $2CaO.Al_2O_3.SiO_2$.
(k) $3CaO.MgO.2SiO_2$, $CaO.MgO.SiO_2$, $2CaO.Al_2O_3.SiO_2$, $MgO.Al_2O_3$.
(l) $3CaO.MgO.2SiO_2$, $CaO.MgO.SiO_2$, $2CaO.MgO.2SiO_2$, $2CaO.Al_2O_3.SiO_2$.

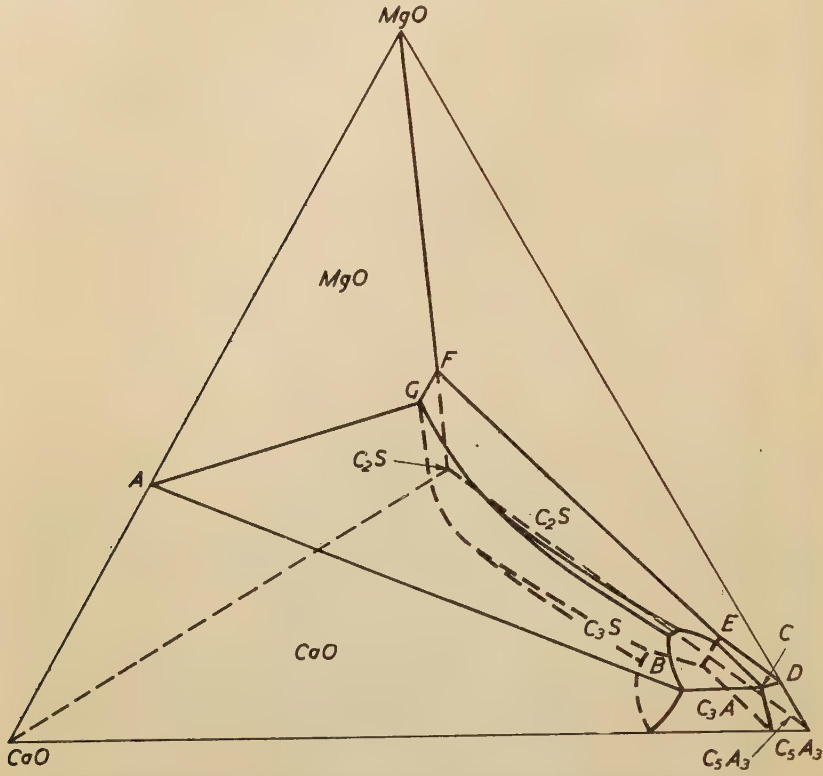


FIG. 6.—Thermal Equilibrium Diagram of the System $CaO–MgO–2CaO.SiO_2–5CaO.3Al_2O_3$, according to McMurdie and Insley.[20]

The outstanding differences between these findings and the phase distribution proposed by McCaffery [21] are as follows :

(1) Gehlenite does not co-exist with free MgO on complete crystallisation. Mixtures of MgO and gehlenite fired to their softening points " dusted " on cooling, and were found by the X-ray to consist of MgO, $MgO.Al_2O_3$ and $2CaO.SiO_2$. The tie-line between MgO and $2CaO.Al_2O_3.SiO_2$ shown in McCaffery's diagram does not exist; this involves considerable modification to the phase distribution in the system.

(2) McCaffery [21] has described a compound called "Madisonite," having the formula $2CaO.2MgO.Al_2O_3.3SiO_2$. Mixed oxides of this composition were fused and crystallised, and examined by the X-ray powder method. The material was found to be heterogeneous, consisting of spinel ($MgO.Al_2O_3$) and a pyroxene phase with a pattern similar to that of diopside. Parker and Ryder [22] also investigated

TABLE II.—*Composition and Viscosities of Blast-Furnace Slags Containing Fluorspar.*

Analysis.					Constitution. Weight (calculated), % of—											η (poise) at 1575° C.
SiO_2. %.	Al_2O_3. %.	CaO. %.	MgO. %.	F. %.	$3CaO.SiO_2$.	$2CaO.SiO_2$.	$3CaO.MgO.2SiO_2$.	$2CaO.Al_2O_3.SiO_2$.	$2CaO.MgO.2SiO_2$.	$3CaO.Al_2O_3$.	$5CaO.3Al_2O_3$.	$CaO.Al_2O_3$.	$MgO.Al_2O_3$.	MgO.	CaF_2.	
33·88	5·34	54·68	5·81	...	...	38·1	45·4	14·4	1·8	...	...	...	...	...	...	16·8
32·84	5·88	54·92	5·98	0·35	...	44·7	41·5	9·8	...	...	...	...	3·2	...	0·72	2·4
34·78	5·30	53·72	5·38	0·56	...	44·5	17·8	14·3	21·8	...	...	...	...	...	1·15	1·6
32·20	5·42	55·30	5·32	1·69	...	38·5	42·8	14·3	...	...	...	...	0·14	...	3·47	1·4
25·56	5·90	55·50	12·90	...	...	73·3	...	...	...	8·3	5·3	...	...	12·9	...	Solid
28·48	5·76	59·58	5·65	0·38	...	81·7	...	...	...	1·8	9·7	...	...	5·7	0·78	6·4
26·36	5·58	59·74	7·58	0·56	1·94	74·1	...	...	...	14·8	...	...	...	7·58	1·15	2·3
26·00	5·82	60·01	6·72	1·27	...	74·6	...	...	...	15·4	0·10	...	...	6·72	2·61	1·9
27·98	5·90	55·72	10·19	...	...	80·3	...	...	...	...	0·89	8·42	...	10·19	...	Solid
28·41	5·87	55·00	10·24	0·35	...	81·5	...	...	...	...	...	3·95	4·63	8·93	0·72	4·5
26·84	6·10	56·40	9·45	1·06	...	75·4	...	...	...	2·7	9·8	...	...	9·45	2·18	2·6

this compound and concluded that the madisonite composition is more likely to be simply a point in a pyroxene-solid-solution series than a distinct quaternary compound.

(3) Merwinite, which does not appear in McCaffery's system, is included.

The constitutions of basic electric-furnace reducing slags free from fluorine may be calculated from the above phase-diagram data, but, since fluorspar is added to a considerable number of these slags, it is necessary to consider its influence on the constitution.

Many opinions have been published concerning the behaviour of fluorspar in basic slags. Sisco [23] reported that fluorspar was not present as such in basic electric-furnace slags, but chiefly as a complex calcium silicofluoride. Analysis revealed that 6·32% of the fluorspar added to the furnace remained in the slag. A number of investigators [24] have shown that the following reaction occurs:

$$2CaF_2 + SiO_2 = SiF_4 + 2CaO.$$

Diepschlag and Brennecke [25] reported that the volatilisation of SiF_4 was checked to some extent when lime was in excess, while Blau, Silverman and Hicks [26] claimed similar stabilisation with alumina. Geiger [24] considered that the above reaction continued until all the fluorspar was used up, the silicon tetrafluoride passing out in the waste gases. Thus, the concentration of CaO increases and that of SiO_2 decreases.

Herty,[24] Schwerin [27] and Rait and Hay,[28] however, found that fluorspar lowered the viscosity of slags composed of CaO, MgO, SiO_2 and Al_2O_3, and that this effect persisted for some time. The change in composition calculated on the assumption that all the fluorine passed out as silicon tetrafluoride did not account for the reduced viscosity.

The compositions and viscosities of a selection of slags investigated by Schwerin [27] are shown in Table II. After additions of fluorspar the presence of fluorine in the slag was proved by chemical analysis and by the greatly reduced viscosity. The constitutions of these slags were calculated by the authors of the present paper from the phase-diagram data on the assumption that the fluorine occurs as CaF_2.

The stability of some of the silicates encountered in basic electric finishing slags in the presence of fluorspar was tested by firing a mixture of the compound with an equal weight of fluorspar at 1200° C. for 2 days, and submitting the resultant product to X-ray examination. The results are shown in Table III. It would appear that $2CaO.SiO_2$ is not decomposed by

TABLE III.—*Stability of Silicates in the Presence of Fluorspar.*

Mixture.	Constitution (X-ray).
50% $3CaO.SiO_2$, 50% CaF_2	Free lime.
50% CaO, 50% CaF_2	Free lime and CaF_2.
50% $2CaO.SiO_2$, 50% CaF_2	β-$2CaO.SiO_2$ and CaF_2.
50% $3CaO.MgO.2SiO_2$, 50% CaF_2	β-$2CaO.SiO_2$, MgO and CaF_2.

fluorspar, while $3CaO.SiO_2$ readily loses all its silica to form free lime and merwinite loses part of its silica by being converted to $2CaO.SiO_2$. These conclusions obtain confirmation from the results shown in Table II. The fluorine persists in all these slags, which are rich in $2CaO.SiO_2$.

It is important to note that CaF_2 does not form solid solutions with either CaO or $2CaO.SiO_2$. It must therefore form simple eutectic systems with these compounds.

The constitutions of the basic electric-furnace finishing slags (for analyses *see* Table I.) as determined by the X-ray powder method are given

in Table IV. Calcium silicofluorides as postulated by Sisco[23] were not found, and in each case where fluorine was present in sufficient amounts, CaF_2 was detected. Accordingly, the constitutions of these slags were calculated from the phase-diagram data, the fluorine being calculated as CaF_2. The calculated constitutions are shown in Table I.

The X-ray and calculated results show remarkably close agreement, discrepancies being chiefly due to the fact that the slags, because of their rapid cooling, probably do not completely crystallise, whereas the calculations were based on complete crystallisation under equilibrium conditions. The lack of thermal-equilibrium data for the complex system prevents the deduction of the effects of rapid cooling on the constitution, but this may be studied in the simpler case of a ternary system, such as the example given by Lea and Desch[29] for mixes in the $CaO–SiO_2–Al_2O_3$ system. Consider the crystallisation path of a ternary $CaO–SiO_2–Al_2O_3$ mix A (Fig. 7) on cooling from the liquid state, equilibrium being maintained during cooling. The first solid to separate out will be lime, the liquid composi-

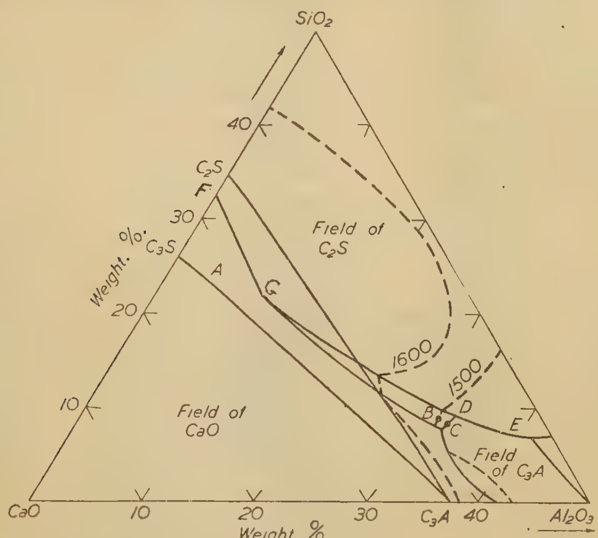


FIG. 7.—Field of $3CaO.SiO_2$ in the System $CaO–SiO_2–Al_2O_3$.

tion changing along the join from the CaO apex to A produced until it intersects FG. On further cooling, CaO and C_2S separate out, the composition of the liquid moving along FG. At G all the lime combines with the C_2S to form C_3S. At 1500° C. the slag will consist of solid $3CaO.SiO_2$ and a liquid B; the composition of B is obtained from where the line from $3CaO.SiO_2$ to A, produced, intersects the 1500° C. isotherm. On further cooling more $3CaO.SiO_2$ crystallises, the composition of the liquid moving along AB, produced, until the $3CaO.SiO_2–3CaO.Al_2O_3$ boundary is reached at C; $3CaO.Al_2O_3$ then begins to separate, and the liquid composition will follow CD until the invariant point D is reached, when complete crystallisation occurs. If, however, the rate of cooling is too rapid for equilibrium to be maintained, the liquid may be quenched to a glass, or crystallised separately, without regard to the solids already present. The liquid compositions B, C and D all lie outside the $3CaO.SiO_2–2CaO.SiO_2–3CaO.Al_2O_3$ triangle. Hence, such liquids alone could not crystallise to give these

TABLE IV.—*Mineralogical Constitutions of*

The approximate percentage by weight of each

Slag No.	CaO.	MgO.	$3CaO.SiO_2$.	$2CaO.SiO_2$.	$2CaO.SiO_2$.
1	Large (50).	...	Small (10).	...	...
2	Large (37).	Small (8).	Small to medium (15).	...	...
3	...	Small to medium (12).	Small (6–8).	...	Large (76).
4	...	Small (15).	Small to medium (15).	...	Large (60–65).
5 (*a*)	...	Absent.	Medium (30).	...	Large (55).
5 (*b*)	...	Absent.	Medium (15).	...	Large (70).
5 (*c*)	...	Absent.	Small to medium (10).	...	Large (85).
6 (*a*)	...	Small (5).	Small (10).	...	Large (70).
6 (*b*)	...	Small (3).	Small (12).	...	Large (75).
7	...	Small (5).	Medium (20).	...	Large (70).
8	...	Small (5).	Small (8–10).	...	Large (75).
9 (*a*)	...	Small (10).	Medium (20).	...	Large (60).
9 (*b*)	...	Small (5).	Medium (15).	...	Large (75).
10	...	...	Small (12).	...	Large (75).
11 (*a*)	...	Small (5).	Medium (15).	...	Large (75).
11 (*b*)	...	...	Small (8–10).	...	V. large (90).
12	...	Small (5).	Small (8).	...	Large (80).
13 (*a*)	...	Small (5).	Small (10).	...	Large (70).
13 (*b*)	...	Small.	...	...	Large.
14	...	Small (5).	Small (10).	...	Large (75).
15 (*a*)	...	Small (5).	Small (10).	...	Large (80).
15 (*b*)	...	...	Small (10).	...	Large (90).
16	...	Small (8).	Small (8).	...	...
17 (*a*)	...	Small (5).	Small (8).	...	Large (75).
17 (*b*)	...	Small (3).	Small (8).	...	Large (85).
18	...	Small (5).	Small (12).	...	Large (75).
19	...	Small to medium (12).	Small (10).	...	Large (70).
20	...	Small (3).	Small (8).	...	Large (85).
21	...	Small (8).	Small (8).	...	Large (75).
22	...	Small (5).	Small (10).	...	Large (80).
23	...	Small (8).	Small (8–10).	...	Large (75).
24 (*a*)	...	Small (8).	Small (10).	...	Large (75).
24 (*b*)	...	Small (8).	Small (10).	...	Large (75).
25	...	Small (15).	Small (10).	...	Large (65).
26	...	Trace (3–5).	Small (8–10).	...	Large (75).
27	...	...	...	Small.	Large.
28	...	...	Trace.	...	Large.
29	...	Small (5).	Small (8).	...	Large (80).
30	...	...	Small.	...	Large.
31	...	...	Trace.	...	Large.
32	...	...	...	Medium.	Small.
33 (*a*)	...	...	...	Trace.	...
33 (*b*)	...	...	...	...	...
34	...	...	...	...	Trace.
35	...	...	...	...	...
36	...	...	Trace.	Medium (25).	...
37 (*a*)	...	...	...	...	Medium.
37 (*b*)	...	...	...	...	...
38	...	...	Small.	Large.	Large.
39 *	...	...	...	Large.	Trace.
40	...	...	...	...	Large.
41	...	Trace.	...	...	Small.
42	...	...	...	...	Trace.
43	...	...	...	...	...
44	...	...	...	...	...

* Distorted merwinite structure.

Basic Electric Finishing Slags.

constituent is given in brackets in most cases.

$3CaO.MgO.2SiO_2$.	$3CaO.2TiO_2$.	$CaO.MgO.SiO_2$.	$2CaO.MgO.2SiO_2$.	$2CaO.Al_2O_3.SiO_2$.	$MgO.Fe_2O_3$.	$MgO.Al_2O_3$.	CaF_2.
...	...	...	...	...	...	...	Large (30).
...	...	...	...	...	...	...	Large (40).
...	...	...	...	...	...	...	Small (8).
...	...	...	...	...	...	...	Small (8–10).
...	...	...	...	...	...	...	Medium (15).
...	...	...	...	...	...	...	Small to medium (15).
...	...	...	...	...	...	...	Small (5).
...	...	...	...	...	...	...	Medium (15).
...	...	...	...	...	...	...	Small (7).
...	...	...	...	...	...	...	Small (5).
...	...	...	...	...	...	...	Small (8–10).
...	...	...	...	...	...	...	Small (10).
...	...	...	...	...	...	...	Small (5).
...	...	...	...	...	...	...	Small (7).
...	...	...	...	...	...	...	Small (6).
...	...	...	...	...	...	...	...
...	...	...	...	...	...	...	Small (6).
...	...	...	...	...	...	...	Medium (15).
Small.	...	...	...	...	...	...	Small.
...	...	...	...	...	...	...	...
...	...	...	...	...	...	...	Small (5).
...	...	...	...	...	...	...	...
...	...	...	...	...	...	...	...
...	...	...	...	...	...	...	Small (10).
...	...	...	...	...	...	...	Small (5).
...	...	...	...	...	...	...	Small (3).
...	...	...	...	...	...	...	Small (8).
...	...	...	...	...	...	...	Small (5).
...	...	...	...	...	...	...	Small (8).
...	...	...	...	...	...	...	Small (6).
...	...	...	...	...	...	...	Small (10).
...	...	...	...	...	...	...	Small (5–7).
...	...	...	...	...	...	...	Small (3).
...	...	...	...	...	...	...	Small (10).
...	...	...	...	...	...	...	Small (5).
Medium.	...	...	...	...	...	...	...
Medium.	...	...	...	...	...	...	Small.
...	...	...	...	...	...	...	Small (5).
Large.	...	...	...	...	...	...	...
Large.	...	...	...	...	...	...	...
Large.	...	...	...	...	...	...	...
Large.	...	Trace.	...	Small.	...	...	...
Large.	Small.	...	...	...	...	Small.	...
V. large.	...	...	...	...	...	...	...
V. large.	...	...	...	...	...	...	...
Large (75).	...	...	...	...	...	...	...
Large.	...	...	...	...	...	...	Small.
Large.	...	...	...	Trace.	...	...	...
Large.	...	...	...	...	...	...	...
Large.	...	...	...	...	...	...	...
Large.	...	...	...	...	...	...	...
Large.	...	...	...	...	...	...	...
V. large.	...	...	...	...	...	...	...
V. large.	...	...	...	...	...	...	Trace possible.
V. large.	...	...	...	...	...	...	...

three compounds. With equilibrium conditions some solid $3CaO.SiO_2$ dissolves in the liquid, and $2CaO.SiO_2$ crystallises out to make up the deficiency in lime at *D*. If this reaction is not allowed to occur at *D*, owing to too rapid cooling, the final product will contain more $3CaO.SiO_2$ than would be present if equilibrium were maintained. The residual liquid then crystallises to form $2CaO.SiO_2$, $3CaO.Al_2O_3$ and $5CaO.3Al_2O_3$, completely solidifying at *E*. The differences in the constitutions of slag *A* at different rates of cooling have been calculated and are given in Table V.

TABLE V.—*The Influence of the Rate of Cooling on the Constitution of a Ternary CaO–SiO$_2$–Al$_2$O$_3$ Slag* (A).

Method of Crystallisation.	Constitution (calculated).				
	$3CaO.SiO_2$. %.	$2CaO.SiO_2$. %.	$3CaO.Al_2O_3$. %.	$5CaO.3Al_2O_3$. %.	Glass. %.
(1) Complete equilibrium . .	79·1	7·7	13·2	...	...
(2) Liquid quenched from *B* . .	84·3	...	...	...	15·7
(3) Liquid quenched from *C* . .	84·9	...	...	...	15·1
(4) Liquid quenched from *D* before separation of $2CaO.SiO_2$.	85·4	...	2·6	...	12·0
(5) Liquid *C* crystallising independently of solid phases present	85·4	3·2	6·8	4·6	...

All slags within the $3CaO.SiO_2$, $2CaO.SiO_2$, $3CaO.Al_2O_3$ field are enriched with $3CaO.SiO_2$ at the expense of $2CaO.SiO_2$ and $3CaO.Al_2O_3$ as a result of rapid cooling. Similar effects will be produced in quaternary slags falling within the field MgO, $3CaO.SiO_2$, $2CaO.SiO_2$ and $3CaO.Al_2O_3$.

The minor constituents which are not detected by the X-ray method either have not crystallised, and are present in the glass, or their amounts are below the limit of visibility.

To obtain an approximate estimate of the relative amounts of phases present in a slag by the X-ray method, a series of 27 standard mixes of known amounts of γ-$2CaO.SiO_2$, $3CaO.SiO_2$, CaF_2 and MgO was prepared (*see* Table VI.) and photographed. With its aid the figures quoted in Table IV. were obtained for various slags. By increasing the number of the standard mixes the relative amounts of phases could be estimated more accurately. It must, however, be noted that the figures represent percentages of the total crystalline material and ignore the glass content and phases below the limit of visibility.

The reducing slags may be classified, depending upon their major constituents, as follows :

(a) *Lime Slags.*—Highly basic white slags containing free lime and $3CaO.SiO_2$ as the major compounds, with varying smaller amounts of MgO, CaF_2 and $3CaO.Al_2O_3$ do not frequently occur. These slags do not "fall" and tend to be viscous in the furnace. They require more fluorspar than the less basic slags for rapid fluxing and suitable fluidity. Such slags do not corrode the dolomite hearth if the CaF_2 content is low. On the contrary, they tend to build up the hearth because a considerable amount of the slag, owing to its viscous nature, may be retained in the furnace after tapping. Consequently, the capacity of a furnace worked with such slags tends to decrease fairly rapidly.

Table VI.—*Standard Mixes of $3CaO.SiO_2$, γ-$2CaO.SiO_2$, CaF_2 and MgO.*

Mix No.	Composition. Weight of—			
	CaF_2. %.	$3CaO.SiO_2$. %.	γ-$2CaO.SiO_2$. %.	MgO. %.
1	10	90	...	...
2	10	70	20	...
3	10	45	45	...
4	10	20	70	...
5	10	...	90	...
6	...	75	25	...
7	...	50	50	...
8	...	25	75	...
9	10	30	60	...
10	10	10	80	...
11	10	...	90	...
12	15	...	85	...
13	15	17·5	67·5	...
14	15	30	55	...
15	15	42·5	42·5	...
16	15	7·5	77·5	...
17	...	10	90	...
18	8·9	80·1	...	11
19	9·0	63	18	10
20	9	...	81	10
21	8·75	17·5	61·25	12·5
22	...	66·75	22·25	11
23	...	44·5	44·5	11
24	...	22·5	67·5	10
25	90	...	...	10
26	...	90	...	10
27	...	...	90	10

These slags also have the disadvantage of severely attacking the launder and ladle refractories. The chief advantage is that increasing basicity of the slag increases the efficiency of sulphur removal.

Fig. 8 shows a photometer record of a reducing slag containing free CaO and CaF_2, together with the corresponding standards. Small amounts of $3CaO.SiO_2$ and β-$2CaO.SiO_2$ are also present.

(b) "*Falling*" *Slags.*—The normal type of reducing slag is known as the falling slag owing to the fact that it disintegrates to a fine powder on cooling. Falling is due to the transformation from the β to the γ form of the large amounts of $2CaO.SiO_2$ present in these slags. These slags also contain smaller amounts of MgO, CaF_2, calcium silicates and aluminates, depending upon the phase group within which the compositions are located (*see* Tables IV. and V.).

(c) *Merwinite Slags.*—It was observed that when there was a delay in tapping the white falling slag altered its properties, becoming brown and non-falling. These slags were shown by the X-ray to contain large amounts of merwinite, and, in most of these cases, it was the only compound detected. The disappearance of free MgO was evidence that this radical forms an integral part of the compound. Chemical analysis of these "pure" merwinite slags showed the CaO/SiO_2 ratio to be lower than that of falling slags, and the $CaO/MgO/SiO_2$ ratios were similar to that for $3CaO.MgO.2SiO_2$. Evidently, silica dripping from the roof during the delay in tapping results in the alteration of the $2CaO.SiO_2$ of the falling slag to $3CaO.MgO.2SiO_2$. The fluorspar contents of these slags are usually low, confirming the fact that CaF_2 is decomposed in the presence of merwinite. Merwinite slags

can, of course, also be formed at an earlier stage in the refining process due to insufficient additions of lime or limestone. For normal steelmelting these slags should be avoided, since they have a greater solubility for dolomite than the white falling slags, thus causing an increased dolomite consumption; furthermore, they are not so efficient in the removal of sulphur.

Typical photometer curves and X-ray photographs of falling slags, merwinitic or non-falling slags, standard mixtures, and the compounds

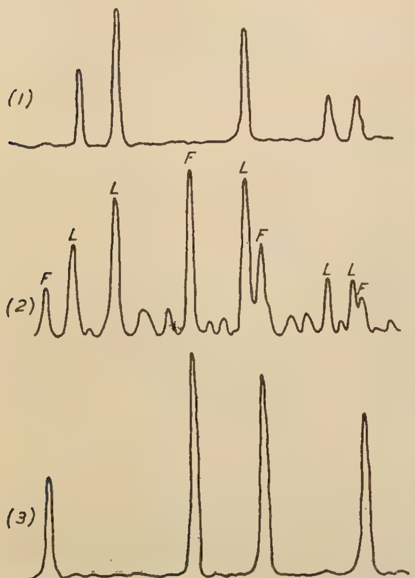


Fig. 8.—Photometer Curves. (1) CaO. (2) Slag No. 1, containing free lime and CaF_2. (3) CaF_2.

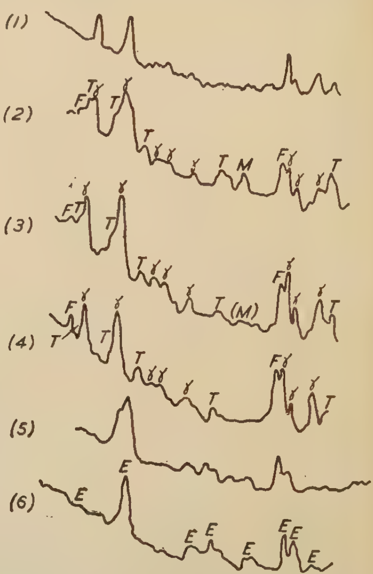


Fig. 9.—Typical Photometer Curves of Basic Electric Reducing Slags. (1) γ-$2CaO.SiO_2$. (2) White falling slag; $3CaO.SiO_2$, γ-$2CaO.SiO_2$, CaF_2 and MgO. (3) White falling slag from same heat as (2) but at a later time. (4) Synthetic mixture, $3CaO.SiO_2$ (30%), γ-$2CaO.SiO_2$ (55%), CaF_2 (15%). (5) Merwinite. (6) Non-falling slag containing merwinite.

γ-$2CaO.SiO_2$ and $3CaO.MgO.2SiO_2$ are reproduced in Figs. 9 and 10, respectively. Certain irregularities were noted in the merwinite pattern in slag No. 39. In actual fact there is insufficient MgO for the amount of merwinite present, and in the calculation this deficiency was made up with MnO, since, as previously reported,[7] there is a manganese analogue of merwinite with the composition $3CaO.MnO.2SiO_2$, having a similar type of pattern to that of merwinite, but with a larger spacing. The irregularity in the merwinite pattern of slag 39 is probably due to the replacement of some of the Mg^{++} ions by Mn^{++} ions in the merwinite lattice.

The constitutions of slags 33 (*a*) and 33 (*b*) exhibit a number of interesting features. These slags were sampled at different periods during the refining of a titanium-bearing stainless steel, the first sample being taken before the addition of ferro-titanium and the second sample just prior to tapping. After taking the first sample and before the addition of titanium, lime was added to the slag. The first slag has a very low CaO/SiO_2 ratio

and, since the MgO content is low, an unusual constitution. Neglecting the small amounts of oxides of iron and manganese, the calculated constitution indicated merwinite as the chief phase, with subsidiary phases monticellite and melilite, which is in agreement with the X-ray result. This type of slag is fluid, thus facilitating the heavy additions of ferro-chromium made during the reducing stage. The ferro-titanium added to the bath contained a considerable amount of aluminium as an impurity. Evidently, the majority of this aluminium has oxidised and passed into the slag as Al_2O_3 along with some of the titanium as TiO_2. The calculated constitution indicated that the TiO_2 occurs as $3CaO.2TiO_2$ in solid solution with the chief phase, $3CaO.MgO.2SiO_2$. In addition, all the alumina occurs as $MgO.Al_2O_3$ (spinel). Confirmation of the essential validity of these calculations was obtained from the X-ray results. The X-ray photographs of these two slags and $CaO.TiO_2$ are reproduced in Fig. 11. The effect of the solid solution of $3CaO.2TiO_2$ on the merwinite pattern and on the occurrence of spinel is illustrated.

(d) *Slags Containing Calcium Carbide.*—A few slags containing calcium carbide (CaC_2) were examined, but the amount of carbide was too low to be detected by the X-ray method. Calcium carbide usually only persists in highly basic slags containing free lime.

An interesting feature was noted in a number of cases where samples of slag were taken at intervals during the refining period. These special samples are indicated by the suffixes (*a*), (*b*), (*c*) &c., to the main slag numbers in Tables I., IV. and V., except in the special cases of slags Nos. 33 and 37. In each case the slag becomes more siliceous owing to the additions of ferro-silicon dust for the reduction of iron oxides on the slag, and, also, probably owing to silica dripping from the roof.

The change in constitution of the finishing slag in passing from the furnace into the ladle is illustrated by samples 37(*a*) and 37 (*b*), the former being sampled just before tapping and the latter from the ladle. Evidently, the slag has reacted with the launder and ladle refractories, thus increasing its content of alumina and silica, while the CaF_2 content has been greatly decreased. The change in constitution shown by calculation is quite striking, and is confirmed by the X-ray result. The constitution and properties of the tapping-slag have an important influence on the life of the ladle linings. The calcium-fluoride content should be kept at the absolute minimum, because CaF_2 reacts directly with the silica of the refractories, and, by lowering the viscosity of the slag, increases its corrosive power.[30] Gehlenite is one of the compounds formed as a result of the reaction between the slag and the ladle lining, and it is believed that some of this reaction product, which adheres to the ladle lining, finds its way into subsequent casts of steel.*

Description of the Phases in Basic Electric Refining Slags.

(1) *Free Oxides.*—Free lime only occurs in a limited number of very basic slags. Free MgO occurs in a large number of slags, but with a decreasing CaO/SiO_2 ratio (below about 1·8) it enters into combination, and, in a few cases, all the MgO is combined. Free MgO is present in the order of 3–10%, the amounts of FeO and MnO being too small to affect appreciably its spacing. The free oxides will receive more detailed attention in a later section.

(2) *Tricalcium Silicate* ($3CaO.SiO_2$).—Tricalcium silicate appears to be unique in its thermal behaviour, since it is stable only over a limited range, *viz.*, from 1250° to 1900° C., and decomposes both above and below this

* J. R. Rait and H. W. Pinder, unpublished work.

range into the same decomposition products—CaO and $2CaO.SiO_2$. It is unstable at ordinary temperatures, but its rate of decomposition is so low that it can exist indefinitely. It does not occur on the liquidus curve of the binary system $CaO–SiO_2$, and it is only stable in contact with liquid in ternary or more complex systems when other oxides are present which lower the liquidus below 1900° C.

Tricalcium silicate has been identified in basic open-hearth slags.[1, 2] In the previous investigations on the constitution of dolomite refractories[5] $3CaO.SiO_2$ was identified by X-ray powder photographs in the presence of $2CaO.SiO_2$ and other compounds; it was also found that it was possible for this to occur in slags.

The structure of $3CaO.SiO_2$ has not yet been established, but the inter-planar spacings have been published elsewhere.[31] Mason[1] reported that it is easy to identify in thin sections of basic open-hearth slags, and has described its appearance and optical properties. There is some evidence that tricalcium silicate can take up small amounts of other oxides in solid solution without appreciably affecting its optical and crystallographic properties.[32]

(3) *Calcium Orthosilicate* ($2CaO.SiO_2$).—Calcium orthosilicate has been identified in basic slags.[1, 2] It melts at 2130° C. and exists in three different crystalline forms, the α form being the stable modification above 1415° C., the β form between 1415° C. and 675° C. and the γ form below 675° C. The inversion at 675° C., on cooling, is accompanied by a 10% increase in volume, which results in the disintegration (popularly known as dusting, or falling) of the slag rich in $2CaO.SiO_2$. As the content of $2CaO.SiO_2$ increases in the finishing slag, the falling tendency increases. Mason[1] was unable to decide which of the two high-temperature forms of $2CaO.SiO_2$, α or β, occurred in basic open-hearth slags, since the differences in optical properties are very slight and are overshadowed by solid-solution effects.

According to Brandenburger[33] the diffraction patterns of α- and β-$2CaO.SiO_2$ are almost identical, showing only minor differences from which it would be extremely difficult to differentiate between the two forms of slags. This subject receives more detailed study, however, in a later section.

The inter-planar spacings of the β and γ forms have been measured by Brownmiller and Bogue[31] and Hansen,[34] respectively.

(4) *Merwinite* ($3CaO.MgO.2SiO_2$).—The history of the discovery of merwinite as a natural mineral and its occurrence in blast-furnace slags was described by Phemister.[12] Parker and Nurse[13] investigated its thermal behaviour, thereby showing that the compound has a primary field of crystallisation in the $CaO–MgO–SiO_2$ system and that it dissociates at 1590° C.

The authors[7] of the present paper reported the discovery of merwinite as a constituent of certain non-falling basic electric finishing slags, and pointed out the similarity of its pattern to that of perovskite, $CaO.TiO_2$. The mineral perovskite is known as the prototype for the crystal structures of a large group of compounds;[35] the unit cell is essentially cubic, but for most members is distorted to various degrees in an orthorhombic, monoclinic, or triclinic manner. Hitherto, the composition ABX_3 was regarded as characteristic of this structure, where A and B are metallic ions of the total valency six and X is usually oxygen. The authors[7] suggested that merwinite and its manganese analogue, $3CaO.MnO.2SiO_2$, also are related to the perovskite type of structure.

The " manganese-merwinite " was identified by X-ray,[7] and is, to the author's knowledge, a new compound. It was obtained unexpectedly by firing 80% of β-$2CaO.SiO_2$ with 20% of $2MnO.SiO_2$ at 1300° C. in hydrogen and cooling rapidly, in an attempt to determine whether a partial solid

solution existed between these two compounds. The X-ray pattern of the fired product was the same as that of merwinite, but of a wider spacing. The large Mn^{++} ion has a radius nearly midway between that of Mg^{++} and Ca^{++} (Ca^{++} = 1·06 Å., Mn^{++} = 0·91 Å., Mg^{++} = 0·78 Å.), so that it may also partly replace calcium, as is known, for example, in the garnets.

The pattern of merwinite corresponds to a pseudo-cubic, but complex and somewhat distorted, unit cell, as is revealed by the splitting of the main cubic lines into groups, together with the appearance of additional reflections. The lattice spacing of the ideal cube of merwinite is 3·76 Å. as against 3·81 Å. for perovskite.

The fact that merwinite shows this type of structure seems important for the following reasons :

(1) The perovskite type of structure is highly accommodating in composition, and permits of a wide variation of relative and absolute ionic radii without breaking down.

(2) No silicate of this type of structure has, to the authors' knowledge, been reported; a direct substitution in $CaO.TiO_2$ of Ti^{++++} by the small Si^{++++} ions (ionic radii, 0·64 and 0·39 Å.) would render the structure unstable (wollastonite, $CaO.SiO_2$, differs entirely), but substitution does become possible if, simultaneously, some Ca^{++} ions are replaced by the smaller Mg^{++} or Mn^{++} ions and certain lattice-positions are vacated.

(3) The structure is not, as usually regarded, tied to the general formula ABX_3. Further evidence for this is afforded by perovskite itself. In the lime-titania binary system * it was observed that $CaO.TiO_2$ and $3CaO.2TiO_2$ are isomorphous except for a slight lattice distortion, a generally permitted feature of the perovskite family. Apparently, a series of solid solutions of CaO in $CaO.TiO_2$ exists, extending beyond 60 mol.-% CaO and modifying, but not destroying, the lattice. The mechanism is again one of omission of ions from the lattice. Thus, contrary to accepted ideas, the perovskite types of compounds may vary over a wide range of compositions from ABO_3 in the direction of, and including, A_2BO_4, also covering merwinite if Ca^{++} and Mg^{++} are classed together.

The perovskite lattice consists essentially of a simple cubic lattice of calcium ions with oxygen ions in the face-centres and a titanium ion in the centre of the cube. The number of atoms per unit cell (N) is 5. For merwinite, assuming the measured spacing to be 3·76 Å., density 3·15 g. per c.c. (value for natural mineral) and the mean atomic weight of $Ca_3Mg.Si_2O_8$ equal to 23·49, calculation gives N of the order of 4·3 atoms, which is in close approach to the value for perovskite, but again indicating a more open lattice (vacant positions). It is necessary to add that the actual merwinite unit cell would consist of a multiple of perovskite cells. Nine such cells stacked in a pseudo-cube containing three molecules (42 atoms) would form the smallest unit able to take an integral number of merwinite molecules and would constitute a probable true unit cell. The detailed structure remains to be analysed. Postulating the basic SiO_4 tetrahedral groups, however, as for all silicates,[36] a line of approach lies in modifying the known perovskite lattice so as to accommodate these groups.

Two examples of finishing slags showing merwinite with a distorted lattice were noted, namely, Nos. 33(*b*) and 39. Calculation indicated 63% of $3CaO.MgO.2SiO_2$ and 9.9% of $3CaO.2TiO_2$, which would yield a perovskite solid solution consisting of 87% of merwinite and 13% of $3CaO.2TiO_2$ in slag 33(*b*). In slag 39 calculation showed 22·0% of $3CaO.MgO.2SiO_2$ and

* J. R. Rait and H. J. Goldschmidt, unpublished work.

7·5% of $3CaO.MnO.2SiO_2$, or a merwinite solid solution consisting of about 75% of $3CaO.MgO.2SiO_2$ and 25% of $3CaO.MnO.2SiO_2$.

Photometer curves and powder photographs of merwinite are shown in Figs. 9 and 10, respectively, in comparison with a non-falling reducing slag.

Table VII. contrasts the interplanar spacings of merwinite and manganese-merwinite. Bannister's * data for merwinite are also given and can be seen to be in reasonably good agreement, in spite of certain differences.

TABLE VII.—*Interplanar Spacings of Merwinite and Manganese-Merwinite.*

Merwinite (Bannister; *see* Phemister [12]). Å.	Merwinite (observed). Å.	Manganese-Merwinite (observed). Å.
...	3·31 W	...
2·94 W	3·05 W	...
2·83 W	2·87 W	2·91 W
...	...	2·85 W
2·73 W	2·74 M	2·74 S
2·65 S	2·65 S	2·65 M
...	2·53 VW	2·54 VW
2·41 VW	2·44 VW	2·44 VW
2·30 W	2·30 M	2·33 W
2·20 W	2·21 M	2·21 MB
2·16 W	2·17 MW	2·18 MW
...	2·06 W	2·07 W
2·03 W	2·04 M	2·03 VW
1·90 M	1·90 S	1·91 S
1·87 MW	1·86 M	1·88 W
...	1·82 W	1·84 W
...	...	1·80 MW
1·75 VW	1·72 W	1·74 W
1·69 VW	...	1·65 VW
1·61 W	$1·60_5$ MW	1·62 W
1·57 W	$1·57_5$ MS	1·58 W
1·53 M	1·53 S	1·56 S
...	1·48 VW	1·51 VW
...	1·45 VW	$1·47_5$ M
1·43 VW	$1·43_8$ MW	...
1·39 VW	1·39 W	1·43 M
1·34 W	1·338 MS	1·400 MW
...	...	1·365 M
1·32 W	1·321 M	1·340 M
...	1·292 W	1·305 W
...	1·277 W	1·282 W
1·23 VW	1·239 M	1·243 W
1·19 VW	1·188 M	1·210 W
...	1·161 W	1·181 W
...	...	1·140 MW
...	1·083 MW	1·090 W
...	1·073 MW	...

VW = very weak.
W = weak.
MW = medium weak.
M = medium.
MS = medium strong.
S = strong.
VS = very strong.

(5) *Calcium Fluoride* (CaF_2).—The pattern of CaF_2 was readily identified in the slags, particularly those with a CaO/SiO_2 ratio greater than 1·4; it belongs to the cubic system. Its structure is well known as the CaF_2 type. There appeared to be no solid solution of any of the other slag

* *See* Phemister.[12]

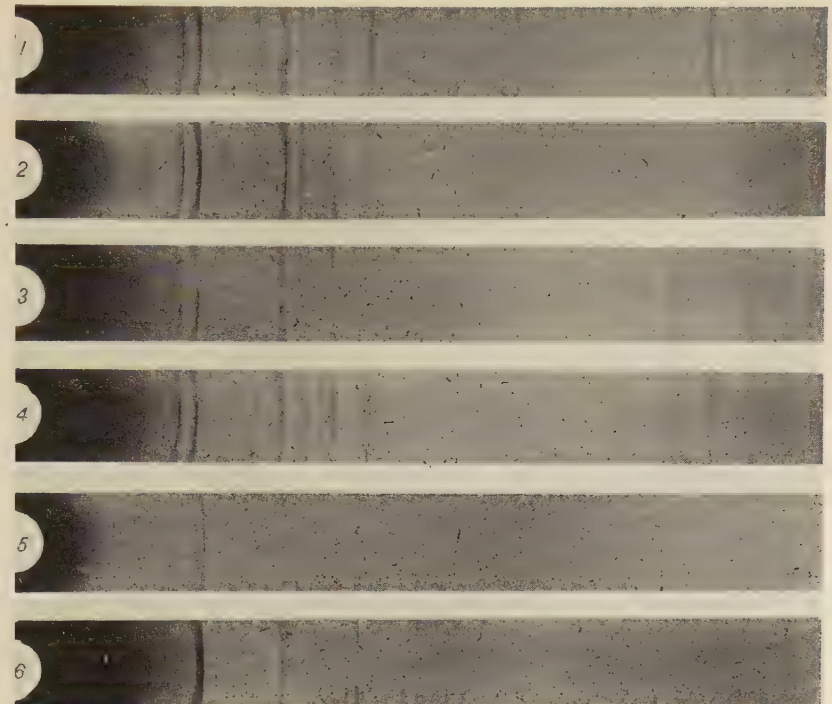


FIG. 10.—Typical X-Ray Photographs of Basic Electric Reducing Slags. (**1**) White falling slag, $3CaO.SiO_2$ (10%), γ-$2CaO.SiO_2$ (60-65%), MgO (15%), CaF_2 (8-10%). (**2**) White falling slag, $3CaO.SiO_2$ (8-10%), γ-$2CaO.SiO_2$ (75%), MgO (5%), CaF_2 (8-10%). (**3**) Synthetic mixture, $3CaO.SiO_2$ (7·5%), γ-$2CaO.SiO_2$ (77·5%), CaF_2 (15%). (**4**) Synthetic mixture, $3CaO.SiO_2$ (17·5%), γ-$2CaO.SiO_2$ (61·25%), CaF_2 (8·75%), MgO (12·5%). (**5, 6**) Non-falling slags containing merwinite.

[*Rait and Goldschmidt.*

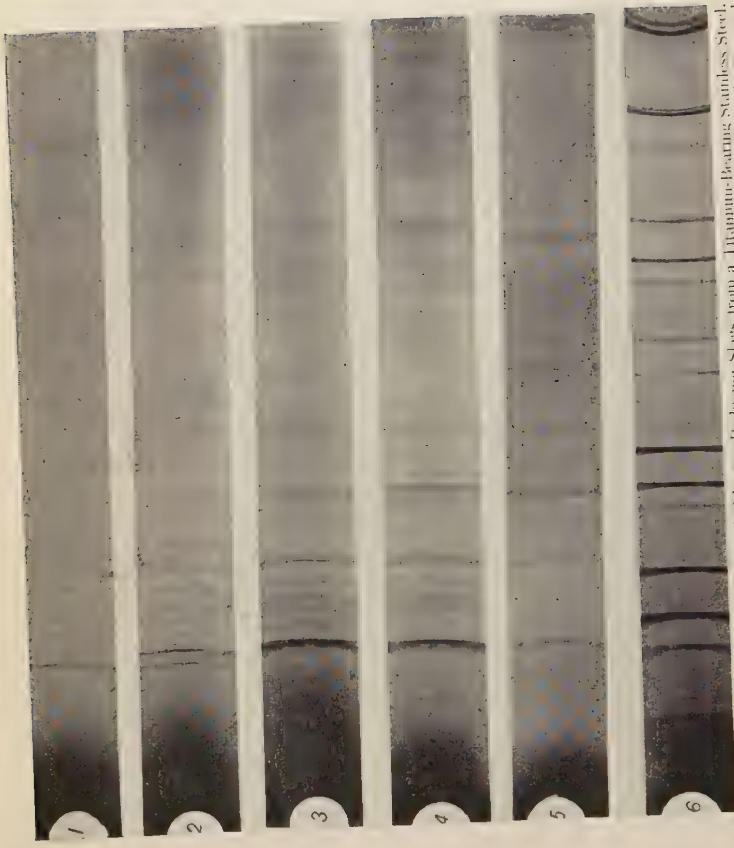


FIG. 11.—X-Ray Photographs of Basic Electric Reducing Slags from a Titanium-Bearing Stainless Steel. **(1)** Gehlenite. **(2)** Reducing slag before addition of Fe-Ti; phases, merwinite, gehlenite, and trace of β-$2CaO.SiO_2$. **(3)** Reducing slag after addition of Fe-Ti, showing merwinite-$3CaO.2TiO_2$ solid solution and $MgO.Al_2O_3$. **(4)** Merwinite. **(5)** $CaO.TiO_2$. **(6)** $MgO.Al_2O_3$.

[*Rait and Goldschmidt.*

PLATE XLVIII.

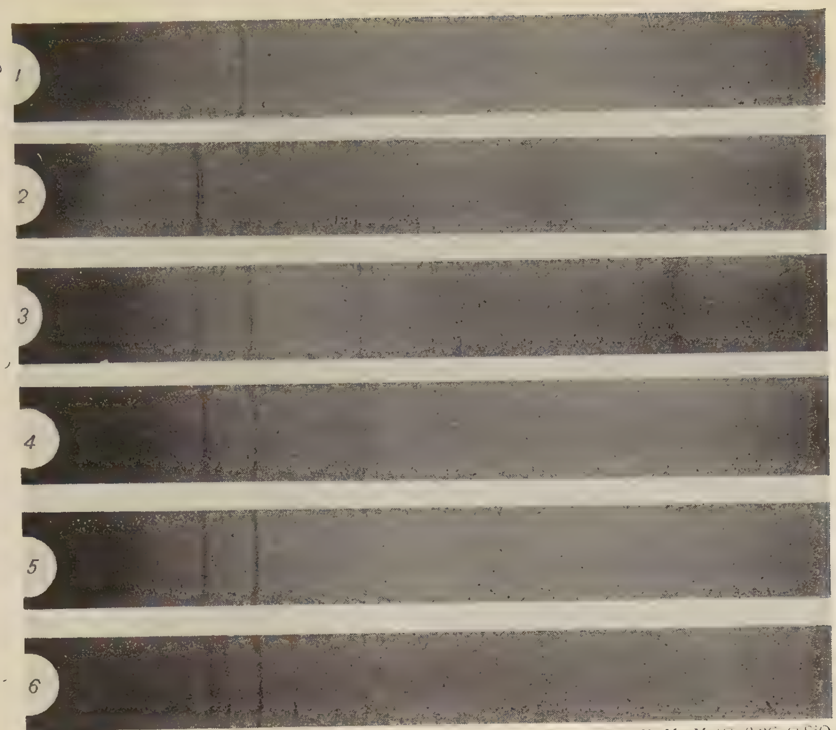


FIG. 12.—Typical X-Ray Photographs of Basic Electric Oxidising Slags. **(1)** (Fe,Mn,Mg)O, β-2CaO.SiO_2 and spinel. **(2)** (Fe,Mn,Mg)O and α'-2CaO.SiO_2. **(3)** (Fe,Mn,Mg)O and β-2CaO.SiO_2. **(4)** (Fe,Mn,Mg)O, β-2CaO.SiO_2 and α'-2CaO.SiO_2. **(5)** (Fe,Mn,Mg)O, β-2CaO.SiO_2, spinel and 2CaO.Fe_2O_3. **(6)** (Fe,Mn,Mg)O, β-2CaO.SiO_2 and spinel.

[*Rait and Goldschmidt.*

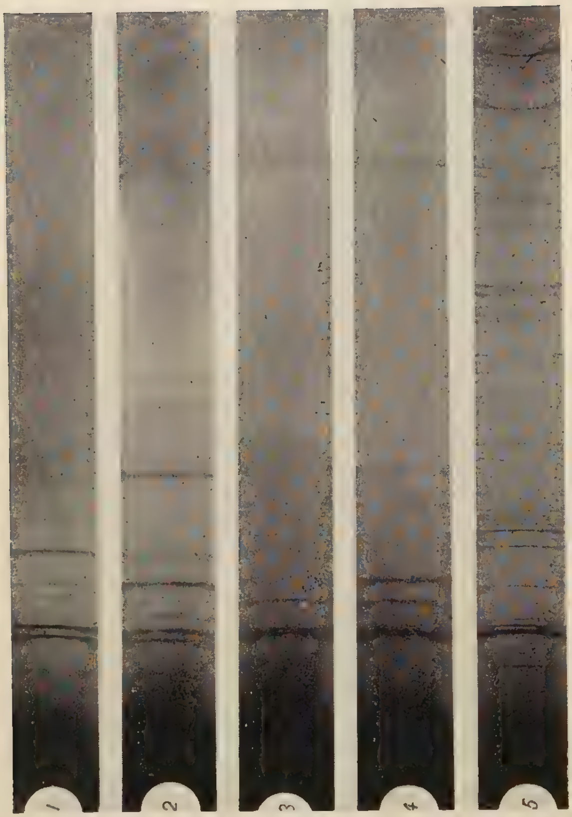


FIG. 13.—Typical X-Ray Photographs of Basic Open-Hearth Slags. **(1)** Nagelschmidtite. **(2)** Slag containing (Fe,Mn,Mg)O, nagelschmidtite and fluorapatite. **(3)** Slag containing (Fe,Mn,Mg)O, (Ca,Mn)O, fluorapatite and nagelschmidtite. **(4)** Slag containing (Fe,Mn,Mg)O, (Ca,Mn)O, fluorapatite and nagelschmidtite. **(5)** Fluorapatite.

[*Rail and Goldschmidt.*

[*To face p.* 417 P.

constituents in CaF_2, since its spacing was constant throughout (a_0 = 5·453 Å.).

Calcium fluoride has a very low melting point (about 1340° C.) and is very fluid at steelmaking temperatures (η (poise) = 0·15 at 1575° C.[27]). Since it does not form solid solutions with the other slag constituents, it must form low-melting eutectics (below 1340° C.), which, together with its remarkable influence in decreasing the viscosity of slag, accounts for its almost universal use in basic electric reducing slags. It increases the rate of slag formation and enables a higher CaO/SiO_2 ratio to be employed, thus facilitating the removal of sulphur. It should be employed with discretion, however, owing to its severe reaction with the furnace and ladle refractories.

(6) *Spinels.*—According to the calculations (Table I.) a spinel phase, chiefly $MgO.Al_2O_3$, should be present in a considerable number of the slags. However, in all cases, with one exception, the indicated amounts were too low for identification by the X-ray method. Calculation indicated 13·4% of $MgO.Al_2O_3$ in slag 33(*b*), and in this case it was identified with certainty by the powder photograph.

$MgO.Al_2O_3$ crystallises in the cubic system and belongs to an isomorphous group of double oxides known as the spinels, with the general formula $R^{\cdot\cdot}O.R_2^{\cdot\cdot\cdot}O_3$. The divalent metal may be $Mg^{\cdot\cdot}$, $Fe^{\cdot\cdot}$ or $Mn^{\cdot\cdot}$, and the trivalent, $Al^{\cdot\cdot\cdot}$, $Fe^{\cdot\cdot\cdot}$, $Mn^{\cdot\cdot\cdot}$ or $Cr^{\cdot\cdot\cdot}$. Recently, White [2] and Mason [1] discussed spinels in relation to basic open-hearth slags.

(7) *Melilite.*—Members of the isomorphous group of minerals with gehlenite and akermanite as end-members are known as melilites. The melilites crystallise in the tetragonal system, their structures having been established by Warren.[37] According to the phase-diagram data melilite can only occur in a reducing slag with a low CaO/SiO_2 ratio and with all the MgO in combination. These conditions are unusual in basic electric practice, so that melilite will only be present on rare occasions.

(8) *Monticellite* ($CaO.MgO.SiO_2$).—Monticellite has a modified olivine structure and a characteristic powder photograph by which it can be readily identified in the presence of other phases. It is a member of an isomorphous group of minerals in which Mg^{++} can be replaced by Fe^{++} or Mn^{++}. Accordingly, its composition can vary considerably in slags.[1] It can only occur in slags with a very low CaO/SiO_2 ratio (less than 1·4), and therefore seldom appears in basic electric finishing slags.

(9) *Other Phases.*—The phases $3CaO.Al_2O_3$ and $5CaO.3Al_2O_3$ are theoretically possible in reducing slags with lime/silica ratios greater than 1·87. These phases were not detected, however. The theoretical amounts are close to the limit of visibility, but the more probable explanation for their absence is that they are among the last phases to crystallise from the melt, and, owing to the rapid cooling, this last liquid is quenched to a glass.

The very small amounts of calcium sulphide and calcium carbide could not be detected by X-rays.

The Constitution of Basic Electric Oxidising Slags.

Table VIII. gives the chemical compositions of a considerable number of basic electric oxidising slags from a variety of steel qualities. The chief components of oxidising slags for ordinary carbon steels are the basic oxides CaO, FeO, MnO and MgO and the acidic oxides SiO_2, Fe_2O_3, Al_3O_2 and P_2O_5. The phase equilibria for this eight-component system have not been worked out, but from a review of the available data White [2] proposed a system of phase assemblages, assuming complete crystallisation for part of the system. As a result of an independent investigation of part of the five-component system $CaO–MgO–SiO_2–Al_2O_3–Fe_2O_3$ * White's system of

* J. R. Rait, unpublished work.

TABLE VIII.—*Chemical Analysis of Basic Electric Oxidising Slags.*

Slag No.*	Chemical Analysis.									
	SiO_2. %.	FeO. %.	Fe_2O_3. %.	Cr_2O_3. %.	Al_2O_3 and TiO_2. %.	MnO. %.	CaO. %.	MgO. %.	P_2O_5. %.	CaF_2. %.
45	12·50	9·50	7·84	3·74	3·12	5·84	49·50	6·40	1·04	...
46	15·10	5·90	6·88	0·56	3·68	8·32	50·5	2·97	0·85	5·00
47 (*b*)	23·10	2·58	1·71	1·01	1·71	3·89	60·1	2·57	0·72	...
48	12·80	18·30	5·44	1·29	3·08	9·12	43·6	5·14	0·87	...
49 (*a*)	12·40	19·90	7·20	1·92	3·43	7·18	44·3	4·26	0·37	...
49 (*b*)	12·80	18·00	11·00	1·82	2·80	7·04	42·1	4·78	0·34	...
49 (*c*)	12·80	20·20	10·70	1·77	3·07	6·70	40·4	4·81	0·32	...
49 (*d*)	12·40	19·40	12·30	1·69	2·35	7·04	40·5	4·97	0·32	...
50 (*a*)	14·60	9·07	6·40	3·57	2·30	4·29	46·3	12·40	0·87	...
50 (*b*)	14·80	10·40	1·76	3·72	2·82	6·12	47·5	11·90	0·68	...
51	11·30	9·36	9·12	5·24	3·39	5·54	46·6	3·80	0·85	4·80
52 (*a*)	16·10	9·79	5·52	1·68	2·77	12·00	45·4	5·94	1·25	...
52 (*b*)	16·00	11·10	4·88	1·68	2·29	12·70	45·3	5·57	1·43	...
60 (*b*)	16·80	8·28	5·60	4·68	2·60	8·69	50·0	3·08	0·82	...
53 (*a*)	19·40	6·91	4·32	3·24	2·84	5·91	52·2	4·42	1·02	...
53 (*b*)	19·30	7·20	2·08	3·34	3·44	9·06	49·2	4·60	1·24	...
54 (*a*)	25·20	2·88	1·44	...	3·04	1·56	57·4	5·47	0·52	...
54 (*b*)	25·00	2·45	1·92	...	3·10	1·53	57·7	4·67	0·56	...
66 (*b*)	19·00	7·20	4·80	2·78	1·92	7·95	49·0	5·79	1·10	...
55 (*a*)	19·80	8·50	2·08	2·73	1·56	9·71	48·6	5·47	1·19	...
55 (*b*)	17·90	6·77	6·24	4·10	1·70	10·40	45·8	5·28	1·24	...
56	19·40	8·20	2·72	3·24	2·22	8·63	48·4	5·21	1·21	...
57	16·20	12·70	2:24	3·34	3·16	8·58	42·4	10·20	0·91	...
58 (*a*)	20·50	8·50	2·24	3·19	2·24	9·49	47·7	5·50	1·10	...
58 (*b*)	19·60	7·92	2·56	3·80	2·00	9·68	47·6	5·50	1·10	...
59	26·80	3·45	1·76	1·11	2·27	4·46	52·0	8·90	0·22	...
60 (*a*)	25·60	5·97	2·48	1·92	2·98	8·38	48·8	4·31	0·38	...
61 (*a*)	23·70	5·83	4·88	2·58	2·32	8·83	46·5	2·21	0·84	...
61 (*b*)	25·60	5·18	2·00	1·41	2·34	6·13	50·8	3·62	0·59	...
62	23·80	6·77	1·60	2·98	2·60	5·68	47·9	8·07	0·80	...
63	23·10	7·56	1·68	2·65	2·18	11·20	44·6	5·47	0·69	...
64	22·00	4·75	10·70	1·75	1·04	7·87	44·5	6·95	0·81	...
65	18·20	7·34	7·68	9·82	1·53	8·86	39·7	4·96	0·81	...
66 (*a*)	18·60	9·51	4·48	4·15	2·28	6·41	48·6	5·57	1·12	...
67 (*a*)	18·20	8·78	3·36	6·81	2·71	9·77	46·4	3·26	1·07	...
67 (*b*)	17·60	8·35	2·88	7·72	2·79	10·50	44·9	3·96	1·34	...
68 (*a*)	15·20	9·94	3·20	9·77	2·44	12·00	42·0	3·55	1·26	...
68 (*b*)	16·20	9·50	5·76	10·30	2·86	10·10	41·7	2·64	1·10	...
69 (*a*)	14·40	12·70	8·64	6·45	2·20	10·50	40·0	3·47	1·22	...
69 (*b*)	12·90	14·80	4·80	10·00	2·84	12·00	37·0	3·47	1·45	...
70 (*b*)	12·70	14·20	6·16	10·40	2·61	7·69	35·6	2·39	0·90	8·00
70 (*a*)	28·10	4·61	1·44	1·44	1·94	6·93	50·9	5·22	0·26	...
71	25·00	4·80	4·24	2·78	4·62	8·62	45·1	4·51	1·02	...
72 (*a*)	24·70	8·49	1·60	7·99	2·68	9·23	41·5	3·74	0·39	...
72 (*b*)	23·40	7·92	1·92	7·43	2·52	9·34	41·9	4·72	0·52	...
72 (*c*)	23·00	9·36	2·72	7·03	2·42	9·14	41·2	5·11	0·53	...
73 (*a*)	22·30	9·65	4·16	4·20	0·92	11·30	40·5	4·42	2·10	...
73 (*b*)	21·30	9·79	7·36	3·97	1·24	11·10	40·1	4·00	1·75	...
74	24·30	6·34	2·40	2·71	2·10	8·00	40·9	12·40	0·29	...
47 (*a*)	28·80	4·32	1·76	1·24	2·24	10·00	43·6	5·50	0·56	...
75 (*a*)	29·40	5·04	1·92	2·02	2·92	9·00	43·1	7·06	0·41	...
75 (*b*)	29·80	5·32	1·20	2·10	3·00	9·31	42·4	7·31	0·17	...
56 (*a*)	26·70	8·20	2·56	6·22	3·32	10·80	35·8	6·22	n.d.	...
76	30·50	5·90	0·80	2·39	4·89	7·95	40·4	6·08	0·16	...

* Slag samples taken (*a*) at "melted" stage, (*b*) before slag-off where two samples recorded or during the boil where more than two samples, and (*c*) just before slag-off.

phase assemblages has been modified and extended as shown in Table IX., column (*a*). In the few cases of phase assemblages containing fluorspar there is usually sufficient of the latter to combine with all the P_2O_5 present as fluorapatite; phase assemblages in which fluorspar is present (Table IX., column (*b*)) are identical with those shown in column (*a*), except that fluorapatite occurs instead of nagelschmidtite. The excess fluorine occurs as CaF_2.

The basic electric oxidising slags from chromium alloy steels have Cr_2O_3 in varying small amounts as an additional component oxide. In the study of the stabilisation of dolomite with Cr_2O_3 and chrome ores Rait and Green [38] concluded that a compound of CaO and Cr_2O_3 containing two molecules of CaO to one of Cr_2O_3 was formed, and that when there was sufficient SiO_2 to form $2CaO.SiO_2$, the Cr_2O_3 combined with MgO to form spinel, $MgO.Cr_2O_3$. Athavale and Jatkar [39] reported a very large number of complex compounds of CaO and Cr_2O_3, the compound with the highest CaO/Cr_2O_3 ratio being $9CaO.4CrO_3.Cr_2O_3$. Ford * has reported the occurrence of the $9CaO.4CrO_3.Cr_2O_3$ instead of $2CaO.Cr_2O_3$, and also that in the presence of MgO, when all the CaO is combined as $9CaO.4CrO_3.Cr_2O_3$, the excess Cr_2O_3 forms $MgO.Cr_2O_3$. Accordingly, the phase groupings proposed by Rait and Green [38] are satisfactory if $9CaO.4CrO_3.Cr_2O_3$ is inserted in place of $2CaO.Cr_2O_3$. The phase assemblages for basic electric oxidising slags containing Cr_2O_3 shown in Table IX., column (*c*), were deduced from these data. The assemblages shown in column (*d*) are similar to those in column (*c*), but modified for the presence of fluorspar. Columns (*c*) and (*d*) must be regarded as tentative, much more information about the complex compounds of CaO and Cr_2O_3 being required.

The constitutions of the various basic electric oxidising slags calculated according to these systems of phase assemblages are reproduced in Table X. These calculations should be compared with the constitutions as determined by the X-ray method (*see* Table XI.).

The calculated and X-ray results are very similar, discrepancies being chiefly due to the fact that the slags, because of their rapid cooling, are not completely crystallised, whereas the calculations are based on complete crystallisation under equilibrium conditions. The lack of thermal equilibrium data for the complex system prevents the deduction of the effects of rapid cooling on the constitution, but this may be studied in the simpler case of the ternary system $CaO–2CaO.SiO_2–4CaO.Al_2O_3.Fe_2O_3$.

Consider the crystallisation path of a quaternary $CaO–SiO_2–Al_2O_3–Fe_2O_3$ mix *A* (Fig. 14) on cooling from the liquid state, equilibrium being main-

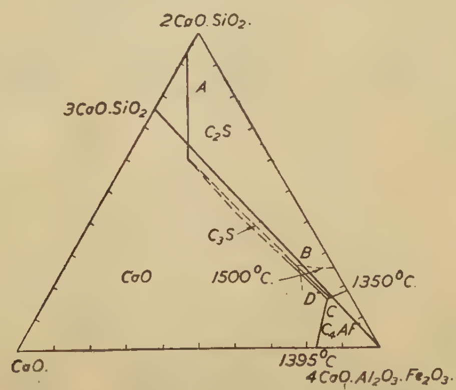


FIG. 14.—Quaternary $CaO–SiO_2–Al_2O_3–Fe_2O_3$ Mix *A*.

tained during cooling. The Al_2O_3/Fe_2O_3 ratio of mix *A* is 0·64, so that all the alumina and iron oxide are combined in the compound $4CaO.Al_2O_3.Fe_2O_3$. The first solid to separate out during cooling will be $2CaO.SiO_2$. At 1500° C.

* Private communication.

TABLE IX.—*Phase Assemblages in Solidified Basic Electric Oxidising Slags.*

$C = CaO$, $A = Al_2O_3$, $S = SiO_2$, $P = P_2O_5$, $F = Fe_2O_3$, $RO = (Fe,Mg,Mn)O$, $R_2O_3 = (Fe,Al,Cr)_2O_3$.

(*a*) Fluorine and Cr_2O_3 Absent.	(*b*) Containing Fluorine, but Cr_2O_3 Absent.	(*c*) Containing Cr_2O_3, but Fluorine Absent.	(*d*) Containing Fluorine and Cr_2O_3.
CaO C_9PS_3 C_3S $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	CaO C_3S Apatite $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	CaO C_9PS_3 C_3S $9CaO.4CrO_3.Cr_2O_3$ $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	CaO C_3S $9CaO.4CrO_3.Cr_2O_3$ Apatite $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$
C_9PS_3 C_3S $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_3S Apatite $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_9PS_3 C_3S $9CaO.4CrO_3.Cr_2O_3$ $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_3S $9CaO.4CrO_3.Cr_2O_3$ Apatite $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$
C_9PS_3 C_3S C_2S $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_3S C_2S Apatite $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_9PS_3 $9CaO.4CrO_3.Cr_2O_3$ C_3S C_2S $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_3S C_2S $9CaO.4CrO_3.Cr_2O_3$ Apatite $\left[\begin{matrix}C_4AF\\ C_2F\end{matrix}\right]$ $\begin{matrix}C_4AF\\ C_3A\end{matrix}$ $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$
C_9PS_3 C_2S C_4AF C_3A C_5A_3 $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	Apatite C_2S C_4AF C_3A C_5A_3 $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	C_9PS_3 $9CaO.4CrO_3.Cr_2O_3$ C_2S C_4AF C_3A C_5A_3 $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$	Apatite $9CaO.4CrO_3.Cr_2O_3$ C_2S C_4AF C_3A C_5A_3 $\left[\begin{matrix}MgO\\ MnO\\ FeO\end{matrix}\right]$

C_9PS_3 C_2S C_4AF C_5A_3 CA [MgO, MnO, FeO]	Apatite C_2S C_4AF C_5A_3 CA [MgO, MnO, FeO]	C_9PS_3 $9CaO.4CrO_3.Cr_2O_3$ C_2S C_4AF C_5A_3 CA [MgO, MnO, FeO]	Apatite $9CaO.4CrO_3.Cr_2O_3$ C_2S C_4AF C_5A_3 CA [MgO, MnO, FeO]
C_9PS_3 C_2S [C_4AF, C_2F] $RO.Fe_2O_3$ — C_4AF CA $RO.Al_2O_3$ [MgO, MnO, FeO]	Apatite C_2S [C_4AF, C_2F] $RO.Fe_2O_3$ — C_4AF CA $RO.Al_2O_3$ [MgO, MnO, FeO]	C_9PS_3 $9CaO.4CrO_3.Cr_2O_3$ C_2S [C_4AF, C_2F] $RO.Fe_2O_3$ — C_4AF CA $RO.Al_2O_3$ [MgO, MnO, FeO]	Apatite $9CaO.4CrO_3.Cr_2O$ C_2S [C_4AF, C_2F] $RO.Fe_2O_3$ — C_4AF CA $RO.Al_2O_3$ [MgO, MnO, FeO]
C_9PS_3 C_2S $C_3.RO.S_2$ $RO.R_2O_3$ RO	Apatite C_2S $C_3.RO.S_2$ $RO.R_2O_3$ RO	C_9PS_3 C_2S $C_3.RO.S_2$ $RO.R_2O_3$ RO	Apatite C_2S $C_3.RO.S_2$ $RO.R_2O_3$ RO
C_9PS_3 $C_3.RO.S_2$ C.RO.S $RO.R_2O_3$ RO	Apatite $C_3.RO.S_2$ C.RO.S $RO.R_2O_3$ RO	C_9PS_3 $C_3.RO.S_2$ C.RO.S $RO.R_2O_3$ RO	Apatite $C_3.RO.S_2$ C.RO.S $RO.R_2O_3$ RO
C_9PS_3 $C_3.RO.S_2$ C.RO.S C_2AS $RO.R_2O_3$	Apatite $C_3.RO.S_2$ C.RO.S C_2AS $RO.R_2O_3$	C_9PS_3 $C_3.RO.S_2$ C.RO.S C_2AS $RO.R_2O_3$	Apatite $C_3.RO.S_3$ C.RO.S C_2AS $RO.R_2O_3$

TABLE X.—*Calculated Constitutions of Basic Electric Oxidising Slags.*

Slag No.	Mineralogical Constitution (calculated).																		
																RO Solid Solution.			
	CaO. %.	$3CaO.SiO_2$. %.	$2CaO.SiO_2$. %.	$3CaO.RO.2SiO_2$. %.	$CaO.RO.SiO_2$. %.	$2CaO.Al_2O_3.SiO_2$. %.	$4CaO.Al_2O_3.Fe_2O_3$. %.	$3CaO.Fe_2O_3$. %.	$9CaO.4CrO.Cr_2O_3$. %.	$3CaO.Al_2O_3$. %.	$CaO.Al_2O_3$. %.	$RO.Cr_2O_3$. %.	$RO.Fe_2O_3$. %.	$RO.Al_2O_3$. %.	$9CaO.P_2O_5.3SiO_2$. %.	$9CaO.CaF_2.3P_2O_5$. %.	FeO. %.	MnO. %.	MgO. %.	CaF_2. %.
45	0·89	42·6	...	...	...	...	15·0	5·52	7·87	...	...	...	...	...	6·04	...	9·50	5·84	6·40	...
46	...	50·9	...	...	...	...	17·5	1·89	...	...	...	...	...	...	...	2·01	5·90	8·32	2·97	4·84
47 (b)	...	47·1	28·1	...	...	...	5·19	...	2·12	1·64	...	...	...	...	4·29	...	2·58	3·89	2·57	...
48	...	41·0	2·64	...	...	...	14·7	1·04	3·10	...	...	...	...	...	4·96	...	18·3	9·12	5·14	...
49 (a)	...	39·5	4·60	...	...	...	16·3	3·09	4·00	...	...	...	...	...	2·13	...	19·9	7·18	4·26	...
49 (b)	...	19·8	20·6	...	...	...	13·2	11·4	3·80	...	...	...	...	...	1·96	...	18·0	7·00	4·80	...
49 (c)	...	13·3	25·6	...	...	...	14·8	9·96	3·67	...	...	...	...	...	1·82	...	22·2	6·70	4·81	...
49 (d)	...	15·9	22·4	...	...	...	11·2	14·7	3·50	...	...	...	...	...	1·82	...	19·4	7·04	4·97	...
59 (a)	...	28·8	17·0		...	...	11·0	4·76	7·57	...	...	...	...	...	5·06	...	9·07	4·29	12·4	...
59 (b)	...	39·5	10·1	...	...	...	5·34	...	7·83	4·50	...	...	...	...	3·95	...	10·4	6·12	11·9	...
51	...	35·1	5·90	...	...	...	16·1	6·47	11·0	...	...	...	...	...	...	2·01	9·36	5·54	3·80	4·64
52 (a)	...	20·7	26·0	...	...	...	13·2	1·95	3·53	...	...	...	...	...	7·27	...	9·79	12·0	5·94	...
52 (b)	...	24·3	22·4	...	...	...	10·9	2·19	3·53	...	...	...	...	...	8·31	...	11·1	12·7	5·57	...
69 (b)	...	23·0	27·9	...	...	...	12·4	2·58	9·85	...	...	...	...	...	4·77	...	8·28	8·69	3·08	...
53 (a)	...	20·3	36·6	...	...	...	13·1	...	6·82	0·24	...	...	...	...	5·93	...	6·91	5·91	4·42	...
53 (b)	...	6·73	45·8	...	...	...	6·31	...	7·02	5·58	...	...	...	...	7·21	...	7·20	9·06	4·60	...
54 (a)	...	17·1	57·6	...	...	...	4·36	...	...	5·63	...	...	...	...	3·02	...	2·88	1·56	5·47	...
54 (b)	...	18·5	55·7	...	...	...	5·84	...	...	5·03	...	...	...	...	3·26	...	2·45	1·53	4·67	...
66 (b)	...	15·0	39·1	...	...	...	9·15	3·06	5·85	...	...	...	...	...	6·40	...	7·20	7·95	5·79	...
55 (a)	...	13·2	43·2	...	...	...	6·31	...	5·74	0·64	...	...	...	...	6·91	...	8·50	9·71	5·47	...
55 (b)	...	0·65	46·3	...	...	...	8·10	6·07	8·63	...	...	...	...	...	7·21	...	6·77	10·4	5·28	...
56 (b)	...	10·4	43·4	...	...	...	8·26	...	6·82	1·30	...	...	...	...	7·03	...	8·20	8·63	5·21	...
57	...	5·74	38·8	...	...	...	6·79	...	7·04	4·61	...	...	...	...	5·29	...	12·7	8·58	10·2	...
58 (a)	...	1·22	53·8	...	...	...	6·91	...	6·74	1·88	...	...	...	...	6·40	...	8·50	9·49	5·50	...
58 (b)	...	5·40	48·1	...	...	...	7·78	...	8·00	0·98	...	...	...	...	6·40	...	7·92	9·68	5·50	...

																	RO.	
59	...	...	76·0	...	...	...	3·68	...	...	...	...	1·40	0·69	2·09	1·28		16·6	...
60 (a)	...	...	72·0	...	...	...	1·31	...	...	...	...	2·43	2·51	3·77	2·20	...	20·0	...
61 (a)	...	...	64·9	...	...	...	2·86	...	...	...	...	2·32	4·93	2·55	4·88	...	14·3	...
61 (b)	...	...	71·2	...	...	...	5·03	...	...	...	...	1·78	0·44	1·28	3·43	...	14·0	...
62	...	...	65·4	...	...	...	4·86	...	...	...	0·70	3·76	...	1·57	4·64	...	19·3	...
63	...	...	63·8	...	...	...	1·43	...	...	...	...	3·35	1·51	2·62	4·00	...	12·5	...
64	...	...	60·2	...	...	...	4·99	0·36	...	...	...	2·21	11·1	...	4·70	...	16·9	...
65	...	...	49·2	...	...	...	7·29	2·75	...	...	...	12·4	4·93	...	4·72	...	17·6	...
66 (a)	...	...	52·6	...	...	...	10·9	1·55	6·76	...	...	1·17	...	...	6·52	...	21·2	...
67 (a)	...	...	48·3	...	...	...	10·2	...	11·7	...	0·88	1·58	...	...	6·22	...	21·5	...
67 (b)	...	...	45·6	...	...	...	8·74	...	10·7	...	1·49	3·33	...	...	7·80	...	22·1	...
68 (a)	...	...	39·0	...	...	...	9·72	...	16·1	...	0·62	2·70	...	...	7·33	...	23·9	...
68 (b)	...	...	42·4	...	...	...	13·6	2·14	5·73	...	...	9·61	...	...	6·40	...	20·2	...
69 (a)	...	...	36·9	...	...	...	10·5	8·82	6·10	...	...	4·40	...	...	7·09	...	25·8	...
69 (b)	...	...	31·7	...	...	...	13·5	0·60	9·02	...	...	7·27	...	...	8·43	...	28·7	...
70 (b)	...	...	36·4	...	...	...	12·4	3·54	6·85	...	...	8·99	...	...	...	2·13	22·4	7·83
70 (a)	...	...	68·0	11·3	...	...	...	...	...	...	...	1·82	1·80	2·71	1·70	...	13·8	...
71	...	...	51·0	16·2	...	...	...	...	...	...	...	3·66	5·30	7·31	5·93	...	10·6	...
72 (a)	...	...	38·1	29·8	...	...	...	...	...	...	...	11·8	2·03	4·57	2·26	...	11·8	...
72 (b)	...	...	50·5	14·1	...	...	...	...	...	...	...	9·38	2·40	3·85	3·02	...	16·5	...
72 (c)	...	...	49·9	13·4	...	...	...	...	...	...	...	8·88	3·40	3·37	3·08	...	18·5	...
73 (a)	...	...	33·3	21·2	...	...	...	...	...	...	...	5·56	5·20	1·57	12·2	...	19·7	...
73 (b)	...	...	43·9	10·4	...	...	...	...	...	...	...	5·14	9·20	2·12	10·2	...	19·7	...
74	...	...	36·1	32·0	...	...	...	...	...	...	...	3·42	3·00	2·93	1·68	...	20·7	...
47 (a)	...	...	13·9	65·4	...	...	...	...	...	...	...	1·83	2·56	3·82	3·27	...	7·35	...
75 (a)	...	...	10·4	71·1	...	...	...	...	...	...	...	2·98	2·78	4·98	0·82	...	7·54	...
75 (b)	...	...	2·29	80·4	...	...	...	...	...	...	...	3·10	1·74	5·11	0·99	...	6·93	...
56 (a)	...	...	...	64·9	10·6	...	...	...	...	...	...	9·16	3·71	5·66	n.d.	...	5·82	...
76	...	...	...	64·7	19·1	5·50	...	...	...	...	...	3·79	1·00	4·86	...	...	...	...

TABLE XI.—*Constitution of Basic Electric Oxidising Slags (X-ray).*

Slag No.	Constitution (X-Ray).								
	$3CaO.SiO_2$.	β-$2CaO.SiO_2$.	α'-$2CaO.SiO_2$.	Merwinite ($3CaO.RO.2SiO_2$).	Monticellite ($CaO.RO.SiO_2$).	Brownmillerite ($4CaO.Al_2O_3.Fe_2O_3.2CaO.Fe_2O_3$).	Spinel ($RO.R_2O_3$).	$(Fe,Mn,Mg)O$.	Additional Phases and Remarks.
45	Large.	...	...	...	...	Small.	...	Medium.	RO phase highly dispersed.
46	Large.	...	...	...	...	Small.	...	Small.	CaF_2 (medium).
47 (*b*)	Large.	Trace.	...	...	...	...	...	Small.	...
48	Large.	...	...	...	...	Small.	...	Large.	...
49 (*a*)	Large.	...	...	...	...	Small/medium.	...	Large.	RO phase highly dispersed.
49 (*b*)	...	Large.	...	...	...	Medium.	...	Large.	
49 (*c*)	...	Large.	...	...	...	Small/medium.	...	Large.	
49 (*d*)	...	Large.	...	...	...	Large.	...	Large.	
50 (*a*)	...	Large.	...	...	...	Medium.	...	Medium.	RO phase highly dispersed.
50 (*b*)	Large.	Small.	...	...	...	...	...	Medium/large.	
51	Large.	...	...	...	...	Small.	...	Medium.	CaF_2 (medium).
52 (*a*)	...	Large.	...	...	...	Small.	Trace (?)	Medium.	...
52 (*b*)	...	Large.	...	...	...	Small.	Trace (?)	Medium.	...
60 (*b*)	...	Medium.	Small.	...	...	Small.	...	Medium.	...
53 (*a*)	...	Large.	...	...	...	Small.	Trace.	Medium.	...
53 (*b*)	...	Large.	...	...	...	Small.	Trace.	Medium.	...

54 (a)	...	Large.	...	...	...	...	...	Small.	γ-$2CaO.SiO_2$(trace).
54 (b)	...	Large.	...	...	...	...	...	Trace.	γ-$2CaO.SiO_2$(trace).
66 (b)	...	Large.	...	...	...	...	Trace.	Medium.	...
55 (a)	...	Large.	...	...	...	...	Trace.	Medium.	...
55 (b)	...	Large.	...	...	...	Small.	Trace.	Medium.	...
56 (b)	...	Large.	Small.	...	...	...	Trace.	Medium.	...
57	...	Large.	...	...	...	...	...	Large.	...
58 (a)	...	Large.	...	...	...	...	Trace.	Medium.	...
58 (b)	...	Large.	...	...	...	...	Trace.	Medium.	...
59	...	...	Large.	...	...	...	Trace.	Small/medium.	...
60 (a)	...	...	Large.	...	...	...	...	Small/medium.	...
61 (a)	...	Large.	...	...	...	Trace.	Trace.	Small/medium.	...
61 (b)	...	Large.	...	...	...	...	Trace.	Small/medium.	...
62	...	Large.	Medium.	...	...	...	Trace.	Medium.	...
63	...	Medium.	...	Small.	...	...	...	Medium.	...
64	...	Large.	...	Small.	...	...	Small.	Medium.	...
65	...	Large.	...	...	...	...	Large.	Medium.	...
66 (a)	...	Large.	...	...	...	...	Trace.	Medium.	...
67 (a)	...	Large.	...	...	...	...	Trace.	Medium.	...
67 (b)	...	Large.	...	...	...	...	Trace.	Medium.	...
68 (a)	...	Large.	...	...	...	Small.	Small.	Medium.	...
68 (b)	...	Large.	...	...	...	Small.	Medium.	Medium.	
69 (a)	...	Medium.	...	...	...	Small.	Medium.	Large.	
69 (b)	...	Medium.	...	...	...	...	Medium.	Large.	...
70 (b)	...	Medium.	...	...	...	...	Medium.	Medium.	CaF_2 (medium).
70 (a)	...	...	Large.	...	...	...	...	Small.	
71	...	Small.	Large.	...	...	...	...	Small/medium.	...
72 (a)	...	...	Large.	...	...	...	Small.	Small.	Probably α'-$2CaO.SiO_2$-merwinite solid solution.
72 (b)	...	...	Large.	...	...	...	Small.	Small.	
72 (c)	...	...	Large.	...	...	...	Small.	Small.	
73 (a)	...	Small.	Large.	...	...	...	Small.	Medium.	...
73 (b)	...	Medium.	Medium.	...	...	...	Small.	Medium.	...
74	...	Small.	...	Large.	...	...	...	Small/medium.	...
47 (a)	...	...	Small.	Large.	...	...	...	Small.	Modified structure of merwinite.
75 (a)	...	...	...	Large.	...	...	...	Small.	
75 (b)	...	...	...	Large.	...	...	...	Small.	
56 (a)	...	...	...	Large.	Small.	...	Small.	Small.	Spacing of $CaO.RO.SiO_2$ phase greater than that of $CaO.MgO.SiO_2$ indicating solid solution of Mn or Fe.
76	...	...	...	Large.	Medium.	...	Small.	...	

the slag will consist of $2CaO.SiO_2$ and $3CaO.SiO_2$ crystals and a liquid of composition *B*, this being the point of intersection of the 1500° C. isotherm and the boundary between the primary-phase fields of these two compounds. On further cooling $3CaO.SiO_2$ and $2CaO.SiO_2$ continue to crystallise, the composition of the liquid moving from *B* to *C*. Since *C* lies on the lime-rich side of the $3CaO.SiO_2$–$2CaO.SiO_2$–$4CaO.Al_2O_3.Fe_2O_3$ triangle, some $2CaO.SiO_2$ re-dissolves in the liquid, combining with the excess CaO to form $3CaO.SiO_2$, which is deposited. Crystallisation is complete at this point, the slag consisting of $3CaO.SiO_2$, $2CaO.SiO_2$ and $4CaO.Al_2O_3.Fe_2O_3$. Evidently, if cooling is too rapid for equilibrium to be maintained there will be a deficiency of $3CaO.SiO_2$ crystals, and free lime will crystallise out, or occur in the glass, final solidification taking place at *D*. The calculated constitutions of this slag for the various modes of cooling are shown in Table XII.

TABLE XII.—*The Influence of the Rate of Cooling on the Constitution of a Quaternary CaO–SiO_2–Al_2O_3–Fe_2O_3 Slag.*

Method of Crystallisation.	Constitution (calculated).				
	Free Lime. %.	$3CaO.SiO_2$. %.	$2CaO.SiO_2$. %.	$4CaO.Al_2O_3.Fe_2O_3$. %.	Glass. %.
(1) Complete equilibrium	...	39·6	50·4	10·0	...
(2) Liquid quenched from *B*	...	33·6	50·5	...	15·9
(3) Liquid quenched from *C*	...	38·1	48·6	...	13·3
(4) Liquid *B* crystallising independently of solid phases present	0·40	39·1	50·5	10·0	...

The majority of the highly basic oxidising slags contain $4CaO.Al_2O_3.Fe_2O_3$. If in the complex system the invariant point corresponding to *C* lies on the lime-rich side, then rapid cooling of the oxidising slags corresponding to *A* would tend to increase $2CaO.SiO_2$ at the expense of $3CaO.SiO_2$, and a certain amount of free lime would occur. Owing to the lack of data, however, it is impossible to deduce the effects of rapid cooling, but the above offers a most probable explanation for certain of the basic oxidising slags being richer in $2CaO.SiO_2$ and poorer in $3CaO.SiO_2$ than is indicated by calculation on the assumption of complete equilibrium. In addition, the evidence indicates that these slags contain small amounts of free CaO, although under equilibrium conditions of cooling all the CaO should be combined.

The minor constituents of the slags which are not detected by the X-ray method have either not crystallised and are present in the glass, or their amounts are below the limit of visibility. The complex compound $9CaO.4Cr_2O_3.Cr_2O_3$ was not detected by the X-ray method in any of the oxidising slags, but the amounts are low, and since it has a very weak pattern this would not be unexpected, even with complete crystallisation. Furthermore, this compound would be among the last to crystallise and, with rapid cooling, some, if not all, will fail to crystallise.

Basic electric oxidising slags vary considerably in constitution following the system of phase assemblages outlined in Table IX., columns (*a*) to (*c*). Nevertheless, the slags summarised in Table XI. may be classified into five

chief types, according to the main phases detected by the X-ray method, as follows :

(*a*) Highly basic slags in which $3CaO.SiO_2$ and RO solid solution are the main phases. Since most of these slags contain considerable amounts of $4CaO.Al_2O_3.Fe_2O_3$, small amounts of free lime probably occur owing to rapid cooling; free lime, of course, can occur with complete crystallisation under equilibrium conditions in the most basic slags, although such slags were not encountered in this survey. The very highly basic slags are viscous when the content of iron oxides is low, and in such cases fluorspar is added to increase the fluidity. Small amounts of the fluorine combine as fluorapatite, the remainder occurring as CaF_2. In only two slags (Nos. 46 and 51) of this type was fluorine detected by analysis, and in both cases CaF_2 was identified by the X-ray; the amounts of fluorapatite are below the limit of visibility. These highly basic slags, provided that they contain sufficient iron oxide, are the most efficient in the removal of phosphorus. They have the lowest solubility for basic oxides such as CaO, MgO and MnO occuring in the various types of slag, other things being equal; accordingly, they are least severe on the dolomite banks.

(*b*) The majority of the slags conform to the type which have β-$2CaO.SiO_2$ and RO solid solution as the outstanding phases. Although these slags contain large amounts of $2CaO.SiO_2$, falling does not occur, owing to the fact that the compound is stabilised in the high-temperature form, probably by the solid solution of the small amounts of P_2O_5. This type of slag is sufficiently fluid without fluorspar, removes phosphorus efficiently and does not attack the dolomite banks severely.

(*c*) This group of slags has a modified form of $2CaO.SiO_2$, which will be referred to as α'-$2CaO.SiO_2$, and RO solid solution as the chief phases. The CaO/SiO_2 ratio is lower than for group (*b*). Small amounts of $3CaO.RO.2SiO_2$, which occur as a co-existing phase, probably dissolve in the $2CaO.SiO_2$, stabilising it in this high-temperature form.

(*d*) Merwinite is the outstanding phase, accompanied by small amounts of RO solid solution and $2CaO.SiO_2$. The combination of RO in merwinite is reflected by the sharp decrease in the amount of free RO solid solution in passing from group (*c*) to group (*d*) with decreasing CaO/SiO_2 ratio. These slags are not so efficient in the removal of phosphorus as the more basic slags, and have a higher solubility for dolomite. The consistent use of such slags increases the consumption of dolomite. Accordingly, they should be avoided for normal basic electric melting.

(*e*) Merwinite is again the outstanding phase, accompanied by monticellite. Negligible amounts of phosphorus are removed by this type of slag. Furthermore, it is very fluid and is unsaturated with respect to dolomite, thus severely corroding the dolomite banks. Such slags should not be employed in the oxidation stage of normal basic electric melting.

Intermediate stages between these groups occur, but the above classification may be useful for practical purposes.

Description of the Phases in Basic Electric Oxidising Slags.

(1) *RO Solid Solution.*—One of the chief phases identified by the X-ray method is the RO solid solution. Since only a single phase of the NaCl type of structure is observed, it follows that the compositions of this phase are within the solid-solution region of the basic-oxide system. The lattice spacings of the RO phase have been measured for most of the slags and are reported in Table XIII. The lattice spacings of the basic oxides of the NaCl type of structure which occur in basic electric oxidising slags are as follows :

CaO . . . . . . 4·80 Å.
MnO . . . . . . 4·43 Å.
FeO (wüstite) . . . 4·27 Å. (at Fe_3O_4 end of solid solution).
MgO . . . . . . 4·20 Å.

The lattice spacings of the RO phase in the oxidising slags taken from different heats and from different stages of the same heat vary considerably, but lie between 4·41 and 4·28 Å., *i.e.*, strictly between the values for MnO and wüstite.

Jay and Andrews [40] recently reported, without publishing experimental data of their X-ray investigation, that the four binary systems FeO–MnO, FeO–MgO, CaO–MnO and MgO–MnO are all single phase. The result on the FeO–MnO system is in agreement with those of Andrew, Maddocks and Howat [41] and McCaughey,[42] but in contradiction to those of Benedicks and Löfquist [43] and Hay, Howat and White.[44] Bowen and Schairer [45] had previously reported the FeO–MgO system as single phase. The ternary system FeO–MnO–MgO is also probably a single phase, but from a consideration of ionic radii it would appear highly probable that although a small amount of CaO can be dissolved in this phase, the introduction of an appreciable amount of CaO would result in the formation of two distinct RO phases. Confirmation of the occurrence of two RO phases was obtained in the examination of basic open-hearth slags discussed in a later section.

In the more basic slags all the FeO, MnO and MgO are in the uncombined state (*see* Table X.). The lattice spacings of the RO solid solutions have been calculated, assuming Vegard's law to be valid (which, it is realised, may not be strictly true), from the FeO, MnO and MgO contents of these slags. These calculated spacings are shown for comparison with the measured spacings in Table XIII. In each case the observed spacing is

TABLE XIII.—*Lattice Spacings of the RO Solid Solutions of Basic Electric Oxidising Slags.*

Slag No.	Lattice Spacing.		Difference between Observed and Calculated Spacings.	Slag No.	Lattice Spacing.		Difference between Observed and Calculated Spacings.
	Observed. Å.	Calculated. Å.			Observed. Å.	Calculated. Å.	
46	4·338	4·326	0·012	63	4·340	4·308	0·032
47 (*b*)	4·312	4·291	0·021	64	4·307	4·271	0·036
48	4·360	4·323	0·037	65	4·322	4·309	0·013
50 (*b*)	4·364	4·330	0·034	66 (*a*)	4·329	4·293	0·036
51	4·320	4·302	0·018	67 (*a*)	4·368	4·328	0·040
52 (*a*)	4·354	4·312	0·042	67 (*b*)	4·355	4·337	0·018
52 (*b*)	4·352	4·319	0·033	68 (*a*)	4·405	4·333	0·072
60 (*b*)	4·384	4·323	0·061	68 (*b*)	4·404	4·347	0·057
53 (*a*)	4·318	4·300	0·018	69 (*a*)	4·371	4·322	0·049
53 (*b*)	4·330	4·312	0·018	69 (*b*)	4·405	4·336	0·069
66 (*b*)	4·328	4·299	0·029	70 (*b*)	4·409	4·329	0·080
56 (*b*)	4·333	4·307	0·026	71	4·349	4·310	0·039
57	4·293	4·282	0·011	72 (*a*)	4·355	4·324	0·031
58 (*a*)	4·320	4·308	0·012	72 (*b*)	4·337	4·306	0·031
58 (*b*)	4·343	4·301	0·042	72 (*c*)	4·332	4·300	0·032
59	4·289	4·257	0·042	73 (*a*)	4·340	4·323	0·017
60 (*a*)	4·372	4·315	0·057	73 (*b*)	4·327	4·325	0·002
61 (*a*)	4·322	4·334	−0·014	75 (*a*)	4·354	4·296	0·058
61 (*b*)	4·303	4·308	−0·005	75 (*b*)	4·385	4·368	0·017
62	4·294	4·274	0·020	56 (*a*)	4·315	4·310	0·005

greater than the calculated value, though in many cases the agreement is very close. The most probable explanation is that small amounts of free

CaO enter the RO solid solution, thus increasing the spacing. It has already been noted that CaO and MnO are mutually soluble, but CaO is insoluble in FeO and MgO. Hence, a limited solubility of CaO in RO solid solution rich in MnO is probable. Support is given to this theory by the fact that the deviation from the calculated spacings is greatest for the RO solid solutions with the highest MnO contents. This is in agreement with the conclusion based on the phase-equilibria data that small amounts of free lime would be present in these slags owing to rapid cooling.

Since, in the less basic slags, each of the RO oxides FeO, MnO and MgO may be combined in two or more co-existing phases such as $RO.R_2O_3$, $3CaO.RO.2SiO_2$, $CaO.RO.SiO_2$ and RO, and since the distribution of each of these oxides is not known, the composition of the phases containing RO cannot be exactly determined. For the purposes of calculation, however, a given distribution may be assumed which will enable the types of phases present in the slags to be indicated, but the exact composition and amount of each phase is not necessarily correct. In other words, the phase $CaO.RO.SiO_2$ may be shown to be present by calculation, but the amounts of FeO, MnO and MgO making up the RO part are not known. However, since MgO is a stronger base than FeO and MnO, it probably takes some preference over FeO and MnO in entering into combination in the silicates and spinel, which would account for the observed lattice spacing being greater than that calculated on a distribution of the oxides in proportion to their molecular concentration (*see* Table XIII.).

It is interesting to note the change in spacing of the RO phase during the oxidising process. The spacing of the RO phase of an oxidising slag is usually greater at slag-off than at the melted stage. In a few cases, the RO spacings were measured in samples taken at melted, boil, and slag-off. The values at boil are the highest, so that it would appear to be a general rule that the spacing increases after melted, reaches a maximum at boil, and decreases towards slagging-off (Fig. 15). Apparently, the variation in spacing is mainly controlled by the MnO/FeO ratio, although a decrease in spacing may be due to an increase in the MgO content resulting from hearth corrosion.

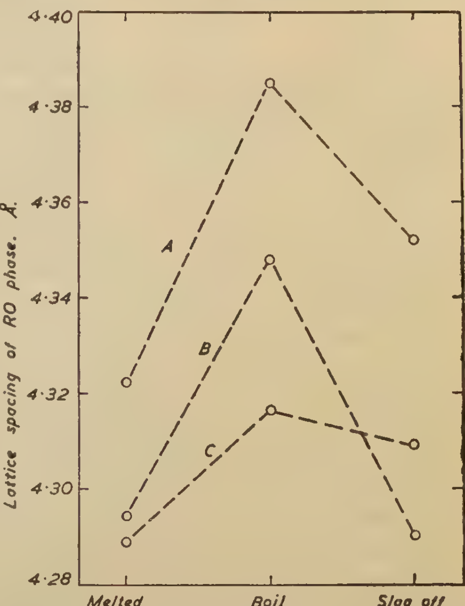


Fig. 15.—Variation in Lattice Parameter of RO Phase during the Oxidising Stage in Three Basic Electric-Furnace Casts.

(2) *Tricalcium Silicate* ($3CaO.SiO_2$).—The properties of $3CaO.SiO_2$, which occurs in a number of the highly basic oxidising slags, have been described in a previous section.

(3) *Calcium Orthosilicate* ($2CaO.SiO_2$).—Calcium orthosilicate occurs in a large number of oxidising slags in the β form. The γ form has not been detected. The stabilisation of the $2CaO.SiO_2$ in the β form is probably due

to the solution of small amounts of P_2O_5, as will be discussed in more detail in a later section.

In a number of the oxidising slags which calculation indicated should contain $2CaO.SiO_2$ and a small amount of merwinite as the silicate phases, a silicate phase "*A*" was observed which differed from those previously encountered. The X-ray pattern resembled that of merwinite, but the differences were sufficiently distinct to show that it was an independent phase. Since the calculations indicated large amounts of $2CaO.SiO_2$, it was considered that phase "*A*" might be α-$2CaO.SiO_2$, known to be stable above 1420° C. Accordingly, an attempt was made to prepare α-$2CaO.SiO_2$ by quenching it from 1790° C. in water. The X-ray pattern of this specimen was identical with that of phase "*A*" as shown by the interplanar spacings in Table XIV. A photometer record of the pattern of high-temperature $2CaO.SiO_2$ is shown in Fig. 16 along with one of a typical slag of this class, the latter also revealing some RO and spinel. The specimen, in the form of a cylinder, rested on magnesite during firing; three samples taken from (*a*) the top, (*b*) the middle and (*c*) the bottom of the pill showed a slight progressive increase in intensities owing to phase "*A*" pattern proceeding from top to bottom. Apparently, reaction with the magnesite and diffusion of the reaction products had stabilised phase "*A*" in favour of the β form. Samples of $2CaO.SiO_2$ contained in rhodium foil, however, fired at 2000° C. and 1800° C., respectively, and quenched, had X-ray patterns identical with that of β-$2CaO.SiO_2$. The indications are, therefore, that the entry of some magnesium, manganese or iron ions into the $2CaO.SiO_2$ lattice are favourable, if not necessary, for the stabilisation of this high-temperature form.

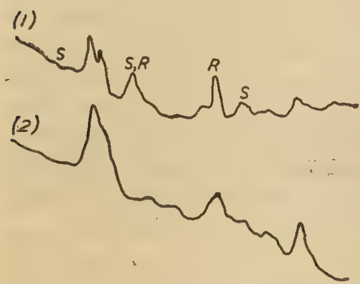


FIG. 16.—Photometer Curve of Basic Electric Oxidising Slag containing α'-$2CaO.SiO_2$ and Spinel. (1) Basic electric oxidising slag. (2) α'-$2CaO.SiO_2$.

The resemblance of the X-ray pattern of manganese-merwinite, reported in an earlier section, with that of phase "*A*" is so great that very probably a continuous series of solid solutions exists between high-temperature $2CaO.SiO_2$ and $3CaO.MnO.2SiO_2$, the similarity between the ionic radii of Ca^{++} and Mn^{++} permitting such a substitution. Since the Mg^{++} ion is much smaller than the Ca^{++} ion, it is doubtful whether such a continuous solid solution extends to $3CaO.MgO.2SiO_2$, although a partial solid solution is probable.

Regarding the identity of phase "*A*" with the known α-$2CaO.SiO_2$, its interplanar spacings were compared with the data given by Brandenburger.[33] The values differ entirely, but they agree better but not completely, with those given by Hanawalt.[46] The values are contrasted in Table XIV. The question thus arises, which of the three reported phases really corresponds to the α-$2CaO.SiO_2$? Brandenburger's pattern is almost identical with that of β-$2CaO.SiO_2$; his compound, however, was prepared by quenching a melt containing 91% of $2CaO.SiO_2$, 7% of $5CaO.3Al_2O_3$ and 2% of Cr_2O_3 from 1650° C., and it would appear that the β form, slightly modified by the substitution of SiO_4 by AlO_4 or CrO_4 groups, was obtained. The history of the American material is unknown, and it is quite evident from the present work that small amounts of impurities can greatly affect the structure. The chief fact for the present purpose, however, is that the pattern of the slag phase "*A*" agrees with that of a high-tempera-

TABLE XIV.—*Interplanar Spacings of Phase "A", α-2CaO.SiO$_2$ and β-2CaO.SiO$_2$.*

α-2CaO.SiO$_2$ (Hanawalt). Å.	α'-2CaO.SiO$_2$, Q., 1790° C. (observed). Å.	Phase "*A*" in Basic Electric Oxidising Slag (observed). Å.	α-2CaO.SiO$_2$ (Brandenburger). Å.	β-2CaO.SiO$_2$ (Brandenburger). Å.
...	...	...	...	3·01 M
5·6 W	...	...	2·81 VS	2·78 VS
4·32 M	...	...	2·61 MS	2·62 MS
4·05 W	...	...	2·42 MSB	2·42 BW
3·80 M	...	...	2·27 MB	2·30 BW
3·38 W	...	...	2·185 S	2·185 S
3·01 MS	...	...	2·02 M	2·03 MW
2·89 W	...	...	1·971 M	1·975 M
2·74 S	2·74 VS	2·74 VS	1·924 VW	...
2·60 W	2·65 M	2·65 M	1·897 M	1·897 MWB
2·51 WM	...	2·50 WM *	1·799 M	1·802 WB
2·44 WM	...	...	1·760 MW	1·760 MS
2·32 WM	2·35 WM	2·34 WM	1·702 MWB	1·696 MB
2·24 W	2·24 W	2·23 M	1·632 M	1·629 S
2·18 WM	2·18 MS	2·16 MS *	1·606 MW	1·608 W
...	2·12 W	2·10 WM	1·583 M	1·573 VW
2·02 WM	2·02 WM	2·02 WM	1·548 MW	1·542 W
1·90 S	1·93 S	1·93 S	1·524 W	1·519 W
1·80 S	1·80 W	1·81 M	1·480 M	1·482 MS
1·75 MS	...	...	1·419 WB	1·417 W
1·68 MS	1·68 W	1·68 VW	...	1·387 W
1·63 MS	...	...	1·365 MWB	1·362 MW
1·53 WM	1·57 S	1·57 M	1·340 W	...
...	...	1·53 S *	1·291 MWB	1·294 MB
1·495 WM	1·49 WM	1·48 WM	1·246 MWB	1·246 M
1·470 M	...	...	1·214 W	1·210 MW
1·440 W	1·44 WM	1·43 WM	1·198 W	1·196 MW
1·405 W	1·41 VW	1·40 WM	...	...
...	1·370 M	1·365 M	...	...
...	1·342 M	1·337 SM	...	...
1·300 W	1·300 M	1·302 M *	...	...
1·268 W	1·270 W	1·262 W	...	...
1·254 M	...	...	...	...
1·230 WM	...	...	...	...
1·180 W	...	...	...	...
1·160 M	...	...	...	...
1·134 M	...	...	...	...
1·092 WM	...	...	...	...

* (Fe,Mn,Mg)O.

ture form of 2CaO.SiO$_2$, which may contain small amounts of RO in place of CaO. The phase will accordingly be described as α'-2CaO.SiO$_2$, so as to leave open the question of its identity with previously reported high-temperature forms.

(4) *Merwinite.*—In the majority of cases the X-ray pattern of the merwinite found in the oxidising slags was identical with that of 3CaO.MgO.2SiO$_2$ and the compound found in reducing slags. In four slags, however, Nos. 75(*a*), 75(*b*), 56(*a*) and 76, a distorted merwinite pattern was observed. In these slags there is insufficient MgO to satisfy the 3CaO.RO.2SiO$_2$ requirements, and a considerable deficiency must be supplied by MnO or FeO, which solid solution probably accounts for the distortion of the merwinite lattice. Since the merwinite pattern in the other slags is normal, it would appear that MgO takes considerable preference over MnO and FeO in the combination with CaO and SiO$_2$ to form the compound.

(5) *Monticellite (CaO.RO.SiO$_2$).*—Monticellite is an unusual constituent in basic electric oxidising slags, since it can only occur in slags with a low

CaO/SiO_2 ratio (approximately 1 : 1). It was only observed in the two slags with the lowest basicity, and in each case the spacing was greater than that of $CaO.MgO.SiO_2$, indicating solid solution of $CaO.FeO.SiO_2$ and $CaO.MnO.SiO_2$. Mason[1] identified $CaO.RO.SiO_2$ in basic open-hearth slags and described its properties. Its composition varies somewhat from slag to slag, the proportions of iron, manganese and magnesium varying, and the CaORO ratio probably also deviates from the ideal 1 : 1.[1] Its optical properties thus vary with composition; the refractive indices are approximately α 1·66, β 1·68, γ 1·69, and the birefringence is about 0·03. It is distinguished in slags from other silicates by its lower refractive indices and higher birefringence, but may be confused with merwinite, since they have similar crystal development, unless merwinite shows its characteristic twinning. Since the X-ray pattern is very characteristic, it is readily identified in the presence of other silicates, and the spacing gives a clue to the proportion of iron, manganese and magnesium present in the RO part.

(6) *Brownmillerite Solid Solutions* ($4CaO.Al_2O_3.Fe_2O_3$–$2CaO.Fe_2O_3$).—The $2CaO.Fe_2O_3$–$4CaO.Al_2O_3.Fe_2O_3$ solid solution has been shown by calculation to occur in all the more basic oxidising slags and only disappears when the basicity is sufficiently low for merwinite to appear. It tends to be most prominent in highly basic slags. In spite of the fact that it is among the last phases to crystallise, rapid cooling probably preventing its complete crystallisation, it has been identified in a considerable number of slags.

The compounds $2CaO.Fe_2O_3$ and $4CaO.Al_2O_3.Fe_2O_3$ form a continuous series of solid solutions.[47] Since their interplanar spacings differ very slightly, it is almost impossible to estimate the extent of solid solution by the X-ray pattern, and, accordingly, the phase is reported as brownmillerite solid solution. The $2CaO.Fe_2O_3$ is orthorhombic, occurring as black crystals which appear yellow-brown in thin section; the refractive indices are α 2·261 and γ 2·274. Brownmillerite appears as prismatic reddish-brown grains, showing pleochroism.

(7) *Spinel* ($RO.R_2O_3$).—It has been shown by calculation and confirmed by the X-ray examination that a small amount of spinel solid solution, $RO.R_2O_3$, occurs in a considerable number of the oxidising slags. Although the composition of the spinel solid solution varies appreciably from slag to slag, it is always single phase, since the spinels are completely isomorphous. The composition of the R_2O_3 part of the spinel, but not the ultimate composition, can be obtained by calculation. Hence, since the proportions of FeO, MnO and MgO in combination in the spinel are not known, the spinels are reported as $RO.Cr_2O_3$, $RO.Fe_2O_3$ and $RO.Al_2O_3$. The lattice spacings of the spinel phase cannot be determined accurately in these slags, since the phase is only subsidiary to others and the high-order lines are ill-defined.

(8) *Other Phases.*—Small amounts of CaF_2 were detected in the two slags to which fluorspar had been added. In these slags the small amounts of P_2O_5 probably occur as fluorapatite ($9CaO.CaF_2.3P_2O_5$). In the other slags, P_2O_5 will be combined as nagelschmidtite ($9CaO.P_2O_5.3SiO_2$), which is soluble in $2CaO.SiO_2$. The phosphates were not detected in any of these oxidising slags because the amounts are below the limit of visibility. A detailed account of the phosphates will be found in the section on basic open-hearth slags.

In the very basic slags, the Cr_2O_3 is combined with CaO, but the exact composition of the compound is not known, although the composition $9CaO.4CrO_3.Cr_2O_3$ has been postulated on the basis of the available data. Since the pattern of this compound is very weak, the amounts shown to be present by calculation would not be detected. Further work is required to establish the exact nature of this compound.

In a number of the slags calcium aluminates and, in one case, gehlenite are indicated by calculation, but the amounts are too small to be detected.

Photometer curves of two basic electric oxidising slags are shown in Fig. 17. The slags were derived from the same cast, one sampled at melted, the other at slagging-off. β-$2CaO.SiO_2$, (Fe,Mn,Mg)O and a trace of spinel are common to both, but the appearance of $2CaO.Fe_2O_3$ in the second sample may be seen. Powder-photographs of various oxidising slags are reproduced in Fig. 12, the constitutions being given underneath.

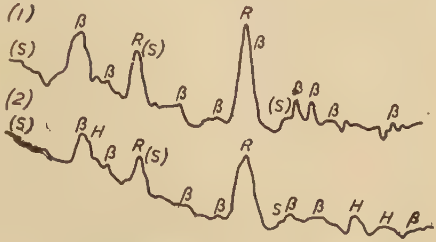


Fig. 17.—Photometer Curves of a Basic Electric Oxidising Slag. (1) At melted. (2) At slag-off.

The Constitution of Basic Open-Hearth Slags.

Basic electric oxidising and basic open-hearth slags have the same component oxides and accordingly belong to the same phase system. As previously noted, the phosphate content of the electric oxidising slags is too low for identification by the X-ray examination; in contrast, P_2O_5 is one of the major acid radicals in basic open-hearth slags and considerable amounts of phosphates occur. Accordingly, to obtain some information about the phosphates in oxidising slags, a number of basic open-hearth slags were investigated by the X-ray method. The preliminary investigation revealed important features about the constitution of basic open-hearth slags not previously published in the literature and, accordingly, the work was extended to include slags from a number of sources.

Numerous investigations have appeared on the constitution of basic open-hearth slags, the more important being those of Hilgenstock,[48] Carnot and Richards,[49] Scott and McArthur[50] and Schneiderhöhn.[51] White[2] has reviewed the literature and attempted to correlate the available data on the constitution of open-hearth slags with his proposed scheme of phase assemblages. As indicated in the preceding section, White's phase assemblages require modification as a result of recent investigations on the five-component system $CaO–MgO–SiO_2–Al_2O_3–Fe_2O_3$ * and the ternary system $CaO–P_2O_5–SiO_2$.

The System $CaO–SiO_2–P_2O_5$.

The existence of the binary compounds $4CaO.P_2O_5$, $3CaO.P_2O_5$, $2CaO.P_2O_5$ and $CaO.P_2O_5$ was established by Körber and Trömel.[52] The two latter compounds do not occur in basic slags. The existence of oxy-apatite $10CaO.3P_2O_5$ has been reported, but the weight of evidence indicates that such a compound does not exist.[2] It is possible, however, that the fluor- and hydroxy-apatites contain oxy-apatite in solution equivalent to a partial omission of F^- or $(OH)^-$ ions from the apatite lattice.

Barrett and McCaughey[9] and Trömel[53] have investigated the CaO–SiO_2–P_2O_5 system since the publication of the review by White.[2] The thermal equilibrium and phase diagrams due to the former authors are reproduced in Fig. 18.

The compound $9CaO.P_2O_5.3SiO_2$ was reported by Körber and Trömel,[54] while Nagelschmidt[8] first isolated, from a basic open-hearth slag, the compound $7CaO.P_2O_5.2SiO_2$, which is isomorphous with the former.

* J. R. Rait, unpublished work.

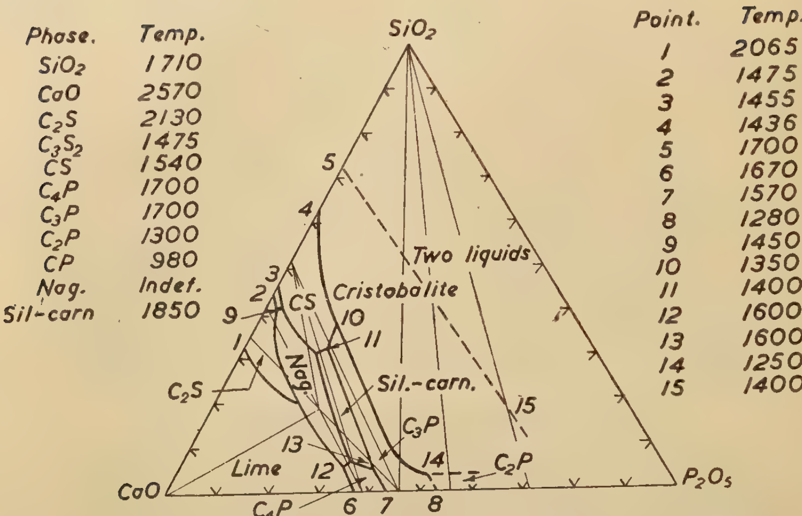


FIG. 18.—Equilibrium Diagram for the System $CaO-SiO_2-P_2O_5$, according to Barret and McCaughey.[9]

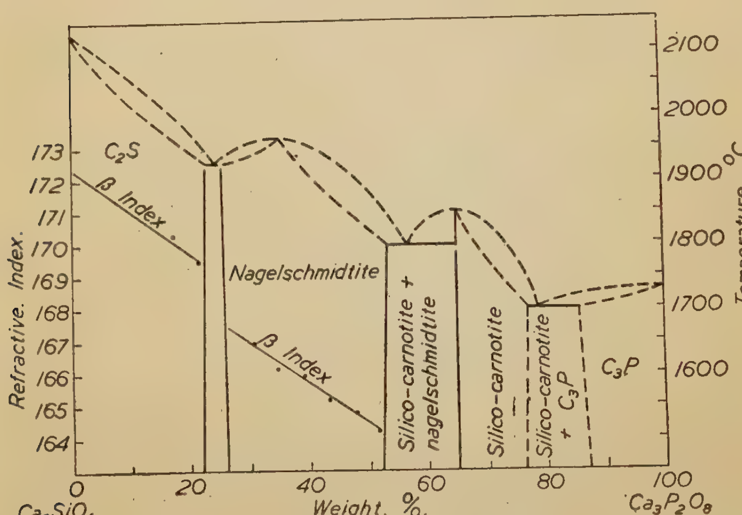


FIG. 19.—Probable Equilibrium Relationships in the System $Ca_2SiO_4-Ca_3P_2O_8$, according to Barrett and McCaughey.[9]

The authors of the present paper confirmed Nagelschmidt's findings by the X-ray examination of two synthetic mixtures which had been fired to about 1700° C. and which had compositions the same as those of the above compounds. These fired mixtures consisted of a single phase, the X-ray patterns being identical and corresponding to nagelschmidtite. These results suggested a range of solubility; this was confirmed by Barrett and McCaughey,[9] who found the nagelschmidtite phase with the P_2O_5 content

varying between 12 and 24%. There is extensive solid solution in the quasi-binary system $2CaO.SiO_2$–$3CaO.P_2O_5$, in which both nagelschmidtite and silico-carnotite ($5CaO.P_2O_5.SiO_2$) are found (Fig. 19); the results obtained by the authors in their examination of slags indicate that the range of solid solubility at the $2CaO.SiO_2$ end of this system is not so extensive as that shown in Fig. 19.

The phase fields in this system as accepted by White [2] differ from those proposed by the American workers as follows :

White :[2]

(*a*) $CaO.3CaO.SiO_2$, $9CaO.P_2O_5.3SiO_2$.
(*b*) $3CaO.SiO_2$, $2CaO.SiO_2$, $9CaO.P_2O_5.3SiO_2$.
(*c*) $CaO.5CaO.P_2O_5.SiO_2$, $9CaO.P_2O_5,3SiO_2$.
(*d*) CaO, $5CaO.P_2O_5.SiO_2$, $4CaO.P_2O_5$.
(*e*) $4CaO.P_2O_5$, $5CaO.P_2O_5.SiO_2$, $3CaO.P_2O_5$.

Barrett and McCaughey :[9]

(*a*) CaO, $2CaO.SiO_2$, $9CaO.P_2O_5.3SiO_2$.
(*b*) CaO, $9CaO.P_2O_5.3SiO_2$, $4CaO.P_2O_5$.
(*c*) $9CaO.P_2O_5.3SiO_2$, $5CaO.P_2O_5.SiO_2$, $4CaO.P_2O_5$.
(*d*) $4CaO.P_2O_5$, $5CaO.P_2O_5.SiO_2$, $3CaO.P_2O_5$.

Thus there are two innovations :

(1) Silico-carnotite does not co-exist with free lime either on the liquidus or in the solid state under equilibrium conditions.

(2) There is no primary field of $3CaO.SiO_2$ in this ternary system and, accordingly, free CaO co-exists with $2CaO.SiO_2$ at liquidus temperatures. According to the phase diagram $3CaO.SiO_2$ does not occur in the solidified mixes, but Barrett and McCaughey do not make it clear whether $3CaO.SiO_2$ is completely unstable in the presence of P_2O_5 or merely not formed as a result of the rapid rates of cooling employed in their investigation. It can, however, be concluded that P_2O_5 does inhibit the formation of $3CaO.SiO_2$ on cooling from the liquid state and that if it has a range of stability in the solid state it is only formed on slow cooling or annealing at temperatures below the liquidus. As will be shown later this has an important influence on the constitution of spoon samples of basic open-hearth slag.

The $2CaO.SiO_2$–$3CaO.P_2O_5$ diagram proposed by Trömel [53] differs from that due to Barrett and McCaughey.[9] Trömel claimed that silico-carnotite does not occur in equilibrium with the melt in Thomas slags, but is only formed by peritectoid reaction below 1300° C. Further, two phases "*K*" and "*R*" are reported. "*R*" is almost certainly nagelschmidtite, and "*K*" occurs between $2CaO.SiO_2$ and nagelschmidtite, each showing a solid-solution range. Phases "*K*" and "*R*" are discussed in more detail in a later section.

The presence of fluorspar affects the equilibrium relationships in the CaO–SiO_2–P_2O_5 slags owing to the formation of the very stable compound fluorapatite ($9CaO.CaF_2.3P_2O_5$), and given sufficient fluorspar all the P_2O_5 will combine as apatite. The hydroxy-apatite ($9CaO.Ca(OH)_2.3P_2O_5$) forms a continuous solid solution with its fluor-analogue. According to Bredig, Franck and Fuldner [55] the compound "voelderite" (or oxy-apatite, $10CaO.3P_2O_5$) has, however, not been confirmed, and Körber and Trömel [54] concluded that at least some moisture is necessary to form the apatite structure.

Phase Assemblages for Basic Open-Hearth Slags.

The phase assemblages for the system CaO–MgO–MnO–FeO–SiO_2–Al_2O_3–Fe_2O_3–P_2O_5, with and without CaF_2, are summarised in Table XV.

TABLE XV.—*Phase Assemblages in Solidified Basic Open-Hearth Slags.*

(a) $\frac{SiO_2}{P_2O_5} > \frac{180}{142}$.	(b) $\frac{SiO_2}{P_2O_5} = \frac{180}{142} - \frac{60}{142}$.	(c) $\frac{SiO_2}{P_2O_5} > \frac{60}{142}$.	(d) Containing Fluorine.
CaO C_9PS_3 C_3S $\begin{bmatrix} C_4AF \\ C_2F \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	CaO C_9PS_3 C_4P $\begin{bmatrix} C_2F \\ C_4AF \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	CaO C_9PS_3 C_4P $\begin{bmatrix} C_2F \\ C_4AF \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	CaO C_3S Apatite $\begin{bmatrix} C_4AF \\ C_2F \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$
C_9PS_3 C_3S $\begin{bmatrix} C_4AF \\ C_2F \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_9PS_3 C_5PS C_4P $\begin{bmatrix} C_2F \\ C_4AF \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_9PS_3 C_5PS C_4P $\begin{bmatrix} C_2F \\ C_4AF \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_3S Apatite $\begin{bmatrix} C_4AF \\ C_2F \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$
C_9PS_3 C_3S C_2S $\begin{bmatrix} C_4AF \\ C_2F \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_9PS_3 C_5PS C_4AF C_3A C_5A_3 $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_5PS C_4P C_3P $\begin{bmatrix} C_2F \\ C_4AF \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_3S C_2S Apatite $\begin{bmatrix} C_4AF \\ C_2F \end{bmatrix}$ C_4AF, C_3A $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$
C_9PS_3 C_2S C_4AF C_3A C_5A_3 $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_9PS_3 C_5PS C_4AF C_5A_3 CA $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	C_5PS C_3P C_4AF C_3A C_5A_3 $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$	Apatite C_2S C_4AF C_3A C_5A_3 $\begin{bmatrix} MgO \\ MnO \\ FeO \end{bmatrix}$

C_9PS_3 C_2S C_4AF C_5A_3 CA [MgO MnO FeO]	C_9PS_3 C_5PS [C_4AF C_2F] $RO.Fe_2O_3$ C_4AF CA $RO.Al_2O_3$ [MgO MnO FeO]	C_5PS C_3P C_4AF C_5A_3 CA [MgO MnO FeO]	Apatite C_2S C_4AF C_5A_3 CA [MgO MnO FeO]
C_9PS_3 C_2S [C_4AF C_2F] $RO.Fe_2O_3$ C_4AF CA $RO.Al_2O_3$ [MgO MnO FeO]		C_5PS C_3P [C_4AF C_2F] $RO.Fe_2O_3$ C_4AF CA $RO.Al_2O_3$ [MgO MnO FeO]	Apatite C_2S [C_4AF C_2F] $RO.Fe_2O_3$ C_4AF CA $RO.Al_2O_3$ [MgO MnO FeO]
C_9PS_3 C_2S $C_3RO.S_2$ $RO.R_2O_3$ RO			Apatite C_2S $C_3.RO.S_2$ $RO.R_2O_3$ RO
C_9PS_3 $C_3.RO.S_2$ $C.RO.S$ $RO.R_2O_3$ RO			Apatite $C_3.RO.S_2$ $C.RO.S$ $RO.R_2O_3$ RO
C_9PS_3 $C_3.RO.S_2$ $C.RO.S$ C_2AS $RO.R_2O_3$			Apatite $C_3.RO.S_2$ $C.RO.S$ C_2AS $RO.R_2O_3$

These phase assemblages are similar to those drawn up by White,[2] but modified according to the more recent equilibrium data. For the slags containing fluorine Table XV. only outlines the case where there is the exact amount of fluorine for combination with all the P_2O_5 as apatite. In actual practice there are, in addition, slags containing excess or insufficient fluorine for combination with the P_2O_5. When there is excess fluorine, the same phase assemblages are found with CaF_2 as an additional phase. In the latter case all the fluorine combines as fluorapatite; the $SiO_2/(P_2O_5)$

TABLE XVI.—*Phase Constitution of Basic Open-Hearth*

Slags arranged in order of

Slag No.	(Fe,Mn,Mg)O Solid Solution.	(Ca,Mn)O Solid Solution.	Apatite.	$4CaO.P_2O_5$.	Nagelschmidtite.
					Batch
OH 1	Large.	...	Large.	...	Small.
OH 2	Large.	...	Small.	...	Large.
OH 3	Large.	Small.	...	...	Large.
OH 4	Large/medium.	Small.	...	...	Large.
OH 5	Large/medium.	Small/medium.	Large/medium.	...	Small.
OH 6	Large/medium.	Medium/small.	Medium.	...	Small.
OH 7	Large/medium.	Medium/small.	Large/medium.	...	Small.
OH 8	Large/medium.	Medium.	Medium.	...	Small/medium.
OH 9	Large/medium.	Medium.	Medium/large.	...	Small/medium.
OH10	Large/medium.	Medium.	Medium/large.	...	Small/medium.
OH11	Large/medium.	Medium.	Large.	...	Small/medium.
OH12	Medium.	Large/medium.	Large/medium.	...	Small.
OH13	Medium.	Large/medium.	Large.	...	Small/medium.
OH14	Medium.	Large/medium.	Large/medium.	...	Small.
OH15	Medium.	Large/medium.	Large/medium.	...	Small.
OH16	Small/medium.	Large/medium.	Large/medium.	...	Small.
OH17	Small.	Large/medium.	Medium.	...	Large/medium.
					Batch
OH18	Large.	...	Small.	Small/doubtful.	Small/probable.
OH19	Large.	...	Small.	...	Small/probable.
OH20	Large.	...	Small.	Small/doubtful.	Small/probable.
OH21	Medium.	Small.	Small.	Medium.	Small/probable.
OH22	Medium.	Small/medium.	Medium.	Medium.	...
OH23	Medium.	Medium.	Small.	...	Small.
OH24	Medium/small.	Medium.	Medium.	...	...
OH25	Medium/small.	Medium/large.	Medium.	...	...
					Batch
OH26	Large.	...	...	...	Large.
OH27	Large.	...	Small.	...	Medium.
OH28	Large.	...	Small.	...	Medium.
OH29	Large.	...	Small.	...	Medium.
OH30	Large.	...	...	...	Small.
OH31	Large.	Small.	...	...	Large.
OH32	Large/medium.	Small/medium.	Medium.	...	Small.
OH33	Large/medium.	Small/medium.	Medium.	...	...
OH34	Large/medium.	Medium.	Large.	Small/medium.	...
OH35	Large/medium.	Medium.	...	...	Large.
OH36	Large/medium.	Large/medium.	Large.	...	...
OH37	Large/medium.	Large/medium.	Medium.	...	Small.
					Batch D.—
OH38	Large.	...	Small.	Small/doubtful.	Small.
OH39	Large.	...	...	...	...
OH40	Large.	Small.	Small.	...	...
OH41	Large.	...	Small.	...	...
BB42	...	...	...	...	Large.
BB42	...	...	...	...	Large.
OH43	Small/medium.	...	Large.	...	...

ratio, where (P_2O_5) is the total P_2O_5 content less the amount combined as fluorapatite, determines the phase-assemblage group.

In the investigation of the CaO–FeO system White[2] concluded that these two oxides do not co-exist; accordingly, in his tables of phase assemblages, free lime does not appear with FeO. The fact that free lime had not been identified in basic open-hearth slags, even in the case of very high lime contents, was quoted as additional evidence for the incompatibility of CaO and FeO. Since the authors have identified free lime and FeO solid solution

Slags as Determined by the X-Ray Powder Method.

increasing CaO/FeO ratio.

β-$2CaO.SiO_2$.	Merwinite.	$2CaO.Fe_2O_3$–$4CaO.Al_2O_3.Fe_2O_3$.	Other Phases.
A.			
Small.	...	Small.	...
...	...	Small.	...
Trace.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	$Ca(OH)_2$ (small).
Small.	...	Small.	...
Small.	...	Small.	$Ca(OH)_2$ (small).
Trace.	...	Small.	...
Small.	...	Small.	...
Trace.	...	Trace.	$Ca(OH)_2$ (large).
B.			
Small.	Medium.	Doubtful.	...
Small.	Medium.	Doubtful.	...
Small.	Medium.	Doubtful.	Trace of metallic iron.
Small.	...	Small.	...
Small.	...	Small.	...
Trace/probable.	...	Small.	...
Small.	...	Small.	...
Small.	...	Small.	...
C.			
Medium.	Small/probable.	...	Trace of metallic iron.
...	Medium.	...	Trace of metallic iron.
...	Medium.	...	Trace of metallic iron.
...	Medium.	...	Trace of metallic iron.
...	Medium.	...	...
...	...	Small.	...
Small.	...	Small.	...
Medium.	...	...	...
...	...	Small.	Silico-carnotite (small/medium).
...	...	Medium.	...
Medium.	...	Small.	...
Small/medium.	...	Small.	...
Miscellaneous Slags.			
...	Medium.	...	Trace of metallic iron.
Medium.	...	Large.	...
Small.	...	Large.	Trace of metallic iron.
Small.	...	Large.	Trace of metallic iron.
...	...	...	$5CaO.P_2O_5.SiO_2$ (large), spinel (trace). Ratio of nagelschmidtite to C_5PS ratio increases on firing, spinel decreases.
...	...	...	$5CaO.P_2O_5.SiO_2$ (large), spinel (trace).
Medium.	...	Small.	...

as co-existing phases in a considerable number of basic open-hearth slags, however, this additional modification to the phase assemblages has been incorporated in Table XV.

Experimental Investigation.

The basic open-hearth slags examined in this investigation were supplied by various steelworks, denoted *A*, *B*, *C*, &c. The phase constitutions as determined by the X-ray powder method are listed in Table XVI. Each batch of slags is arranged, starting with the slags in which free lime is absent, in the order of increasing CaO/FeO ratio as determined by the relative intensities of the X-ray patterns.

TABLE XVII.—*Partial Chemical Analyses of Basic Open-Hearth Slags.*

Slag No.	Chemical Analysis.									
	SiO_2. %.	CaO. %.	MgO. %.	MnO. %.	FeO. %.	Fe_2O_3. %.	Total Iron. %.	Al_2O_3. %.	P_2O_5. %.	Fluorine.
					Batch A.					
OH 1	12·00	45·50	N.d.	N.d.	N.d.	N.d.	11·00	N.d.	7·26	Present.
OH 2	9·30	39·80	6·10	7·24	17·9	5·40	17·70	N.d.	9·34	Present.
OH 3	9·20	38·80	4·74	5·56	22·6	5·28	21·30	N.d.	8·46	Absent.
OH 4	11·60	43·70	N.d.	N.d.	N.d.	N.d.	14·20	N.d.	6·95	Absent.
OH 5	12·10	48·60	5·47	6·90	9·79	3·84	10·30	N.d.	8·65	Present.
OH 6	8·00	41·50	8·90	6·73	20·60	6·08	20·30	N.d.	5·55	Present.
OH 7	12·00	48·10	N.d.	N.d.	N.d.	N.d.	12·60	N.d.	6·96	Present.
OH 8	9·80	47·70	N.d.	N.d.	N.d.	N.d.	12·50	N.d.	8·82	Present.
OH 9	10·00	51·50	N.d.	N.d.	N.d.	N.d.	10·00	N.d.	6·35	Present.
OH10	10·40	50·70	N.d.	N.d.	N.d.	N.d.	11·80	N.d.	7·40	Present.
OH11	9·40	54·60	N.d.	N.d.	N.d.	N.d.	12·70	N.d.	7·25	Present.
OH12	10·00	48·40	N.d.	N.d.	N.d.	N.d.	11·70	N.d.	7·54	Present.
OH13	9·80	54·90	N.d.	N.d.	N.d.	N.d.	10·80	N.d.	8·46	Present.
OH14	10·60	54·60	N.d.	N.d.	N.d.	N.d.	10·10	N.d.	8·75	Present.
OH15	11·50	49·30	1·70	5·14	10·10	3·36	10·20	N.d.	7·13	Present.
OH16	10·30	57·80	3·15	5·31	8·35	3·52	9·01	N.d.	7·97	Present.
OH17	9·40	60·0	N.d.	N.d.	N.d.	N.d.	9·40	N.d.	7·03	Present.
					Batch B.					
OH18	18·90	38·50	N.d.	12·30	11·60	2·80	10·30	4·20	5·60	Present.
OH19	16·80	38·10	N.d.	12·20	11·00	2·10	10·00	4·10	5·20	Present.
OH20	16·60	39·30	N.d.	10·40	13·20	4·20	13·00	3·60	4·60	Present.
OH21	10·90	43·5	N.d.	11·90	12·00	3·10	11·50	3·00	9·80	Present.
OH22	10·00	45·50	N.d.	11·30	11·90	4·30	12·20	2·50	8·20	Present.
OH23	10·60	45·50	N.d.	12·40	13·50	3·50	13·00	2·00	6·50	Present.
OH24	10·30	45·20	N.d.	11·00	13·30	2·70	12·30	1·60	7·80	Present.
OH25	10·00	46·40	N.d.	11·50	9·70	3·90	10·20	2·60	9·00	Present.
					Batch C.					
OH26	15·50	36·60	N.d.	12·30	14·30	3·00	13·20	3·60	5·20	Absent.
OH27	14·90	41·50	N.d.	12·40	9·50	2·60	9·10	4·10	5·70	Present.
OH28	14·80	40·60	N.d.	13·20	10·00	2·90	9·80	3·90	5·90	Present.
OH29	14·60	42·60	N.d.	13·00	10·40	2·80	10·00	4·00	6·00	Present.
OH30	14·70	44·00	N.d.	13·29	10·30	3·92	10·80	N.d.	5·75	Present.
OH31	10·00	34·30	N.d.	11·00	24·70	4·20	22·00	2·50	6·00	Absent.
OH32	13·40	44·00	N.d.	11·00	9·50	3·00	9·50	3·00	7·90	Present.
OH33	12·20	42·50	N.d.	11·50	13·20	4·20	13·00	3·00	4·60	Present.
OH34	8·00	46·60	N.d.	13·60	13·20	5·04	14·40	N.d.	5·98	Present.
OH35	7·60	44·80	N.d.	9·09	14·50	9·04	17·60	N.d.	6·76	Absent.
OH36	11·20	45·20	N.d.	7·90	12·70	4·60	13·10	N.d.	5·66	Present.
OH37	10·70	49·20	N.d.	11·70	11·70	4·60	12·30	N.d.	5·78	Present.
					Batch D.—Miscellaneous Slags.					
OH38	12·80	46·40	5·83	4·46	13·00	4·00	12·90	5·26	7·49	Present.
OH39	10·80	39·60	7·96	4·74	23·30	7·84	23·60	2·64	1·80	Absent.
OH40	10·80	42·80	4·78	5·39	20·00	7·68	20·90	2·31	6·00	Present.
OH41	11·50	41·90	6·08	5·00	20·50	6·08	20·20	3·10	5·70	Present.
BB42	8·30	47·70	2·06	3·69	N.d.	19·80	13·80	1·29	14·30	Absent.
OH43	16·20	44·70	6·85	2·53	N.d.	N.d.	8·93	3·91	12·90	Present.

The partial chemical analyses supplied with the slags are given in Table XVII. Unfortunately, the samples were not retained for complete analysis and the phase constitutions cannot be completely calculated from the partial analyses. The definite correlation between constitution and basicity of the slags may be followed, however; the free CaO phase is most prominent in the most basic slags, the amount decreasing as the basicity falls until a stage is reached when the free lime disappears. On further decrease in basicity the changes in constitution follow the system outlined in the Tables of phase assemblages. Essentially, the theoretical phase assemblages deduced from the phase data hold for basic open-hearth slags. However, since the theoretical phase assemblages are based on complete crystallisation under equilibrium conditions, and the slags in the form of spoon samples cool fairly quickly from the liquid state, it is not surprising that there are a few minor discrepancies; they are as follows:

(1) The presence of $3CaO.SiO_2$ is not observed, free CaO and $2CaO.SiO_2$ occurring in the more basic slags in which $3CaO.SiO_2$ would be expected. In the discussion of the $CaO–P_2O_5–SiO_2$ system, however, it was pointed out that $3CaO.SiO_2$ did not occur on the liquidus, dissociating into CaO and $2CaO.SiO_2$ below liquidus temperatures. Accordingly, it is only formed on slow cooling or on annealing at a temperature just below its upper dissociation temperature in the system. Apparently, the rate of cooling of basic open-hearth slag in spoon samples is sufficiently rapid to prevent its formation. Furthermore, as discussed in the previous section, the presence of $4CaO.Al_2O_3.Fe_2O_3$ tends to increase the $2CaO.SiO_2$ and CaO contents on rapid cooling at the expense of $3CaO.SiO_2$.

(2) In slags *B*18, *B*21 and *B*22 small amounts of $2CaO.SiO_2$ are detected as co-existing with $4CaO.P_2O_5$, although, according to the phase-diagram data, nagelschmidtite and not $2CaO.SiO_2$ should be the co-existing silicate. However, $2CaO.SiO_2$ is a member of the adjoining phase field, so that heterogeneity in the composition of the sample is the most probable explanation of this anomaly.

(3) In slag *C*34 silico-carnotite co-exists with free CaO and $4CaO.P_2O_5$, although, according to the most recent $CaO–SiO_2–P_2O_5$ system,[9] the phase co-existing with free CaO and $4CaO.P_2O_5$ should be nagelschmidtite and not silico-carnotite. The invariant points at which CaO, nagelschmidtite, $4CaO.P_2O_5$ and liquid, and nagelschmidtite, silico-carnotite, $4CaO.P_2O_5$ and liquid are at equilibrium, however, are situated closely together, so that rapid cooling might result in the formation of silico-carnotite. According to White[2] previous investigations of the $CaO–P_2O_5–SiO_2$ system indicated the co-existence of silico-carnotite with CaO and $4CaO.P_2O_5$. It would therefore appear that this part of the ternary system requires further investigation.

Further confirmation of the essential validity of the system of phase assemblages is obtained from the recent investigation of the constitution of basic open-hearth slags by Mason.[1] The constitutions calculated according to the system of phase assemblages are shown in Table XVIII. for comparison with those determined experimentally. The agreement between the observed and calculated results is very striking, but there are two apparent discrepancies as follows:

(*a*) Mason failed to detect independent phosphate phases. In the two slags in which fluorapatite should appear the amounts are approaching the limit of visibility. Mason considers that the phosphate is in solid solution in the silicates, the PO_3''' groups replacing SiO_4'''' groups; solid solution of phosphate in $2CaO.SiO_2$ is indicated by the lowering of the refractive indices of this phase. Although the

TABLE XVIII.—*Phase Construction of Open-Hearth Slags.*

X indicates presence; (*X*), small amounts only. R = (Fe,Mn,Mg).

Slag No.	Calculated Constitution (authors).												
	CaO. %.	$3CaO.SiO_2$. %.	$2CaO.SiO_2$. %.	$3CaO.RO.2SiO_2$. %.	$CaO.RO.SiO_2$. %.	$4CaO.Al_2O_3.Fe_2O_3$. %.	$2CaO.Fe_2O_3$. %.	$9CaO.4CrO_3.Cr_2O_3$. %.	$RO.Cr_2O_3$. %.	$RO.Fe_2O_3$. %.	$RO.Al_2O_3$. %.	$9CaO.P_2O_5.3SiO_2$. %.	$9CaO.CaF_2.3P_2O_5$. %.	RO. %.
M 3	...	...	...	39·8	51·9	...	...	...	...	1·36	3·77	...	...	3·51
M 4	...	...	...	43·7	46·2	...	...	...	...	1·20	4·37	...	...	3·49
M 5	...	...	...	19·8	48·0	...	...	...	1·22	...	4·37	13·9	...	12·9
M 6	...	...	...	12·0	41·0	...	...	...	...	...	4·13	29·0	...	14·0
M 7	...	...	...	21·6	23·3	...	...	...	5·95	...	3·70	29·2	...	16·4
M 8	...	...	4·1	80·6	...	...	...	...	...	1·84	3·51	...	...	9·57
M 9	...	...	8·2	38·2	...	...	...	...	4·81	...	3·10	25·8	...	19·5
M10	...	...	21·2	49·7	...	...	...	...	...	1·88	3·63	...	...	23·3
M11	...	...	50·3	26·6	...	...	...	...	...	1·71	2·85	...	...	18·2
M12	...	...	36·1	...	...	9·86	0·09	1·51	3·17	...	...	23·3	...	26·2
M13	...	41·2	16·0	...	...	6·29	1·75	...	...	...	...	...	10·7	24·2
M14	3·34	66·5	...	...	...	6·14	1·16	...	...	...	...	...	11·2	21·5
M15	7·21	38·1	...	...	...	7·96	4·73	1·45	...	...	...	14·4	...	26·3

Slag No.	Observed Constitution (Mason [1]).												
	Free Lime.	$3CaO.SiO_2$.	$2CaO.SiO_2$.	$3CaO.RO.2SiO_2$.	$CaO.RO.SiO_2$.	$4CaO.Al_2O_3.Fe_2O_3$.	$2CaO.Fe_2O_3$.	$RO.Cr_2O_3$.	$RO.Fe_2O_3$.	$RO.Al_2O_3$.	$9CaO.P_2O_5.3SiO_2$.	$9CaO.CaF_2.3P_2O_5$.	RO.
M 3	...	...	...	(X)	X	...	...	...	(X)	...	...	...	(X)
M 4	...	...	...	(X)	X	...	...	...	(X)	...	...	...	(?)
M 5	...	...	...	...	X	...	...	...	...	...	...	...	X
M 6	...	...	...	...	X	...	...	...	...	...	...	...	X
M 7	...	...	...	(X)	X	...	...	(X)	...	...	...	...	X
M 8	...	...	...	X	...	...	...	...	(?)	...	...	...	(X)
M 9	...	...	(X)	X(?)	...	...	...	(X)	...	...	...	...	X
M10	...	...	(?)	X	...	...	...	...	...	...	...	...	X
M11	...	...	X	(?)	...	...	...	...	...	...	...	...	X
M12	...	...	X	...	...	...	(?)	(X)	...	...	...	...	X
M13	...	X	X	...	...	...	X(?)	...	...	...	...	...	X
M14	...	X	(?)	...	...	...	X(?)	...	...	...	...	...	X
M15	...	X	...	...	...	...	X(?)	...	...	...	...	...	X

authors agree that solid solution in the silicates does occur, the extent is limited, and small amounts of nagelschmidtite should occur as an independent phase in a few of these slags.

(*b*) Small amounts of free lime should occur in slags M 14 and M 15 according to calculation. In M 14 the amount is close to the limit of visibility of the X-ray, but it should be possible to detect free lime in M 15. Mason,(1) however, stated that the Fe_2O_3 value tends to be low, a fact which would account for the calculated free lime being higher than that actually occurring in the slag.

The occurrence of $3CaO.SiO_2$ in the three most basic slags is interesting, because in contrast to the slags examined by the authors, the P_2O_5 contents are low. The rates of cooling of Mason's slags were probably much lower, however, thus allowing the formation of $3CaO.SiO_2$ to take place.

The Phases Identified in Basic Open-Hearth Slags.

(1) *Free RO Phases.*

One of the outstanding features of basic open-hearth slags is that free CaO and FeO solid solution are found to co-exist. The CaO contains some MnO in solid solution, which manifests itself by a decrease in the lattice parameter, while FeO dissolves MnO as well as any free MgO present. The lattice parameter of the (Fe,Mn,Mg)O phase is generally greater than that of wüstite, because the amount of MnO is usually greater than that of MgO.

The occurrence of free CaO as a phase which crystallises independently from the liquid basic open-hearth slag has not been previously reported. As will be shown later this discovery has an important practical application to slag control in the basic open-hearth process. The two types of free-oxide phases in basic slags can only be distinguished with certainty by the X-ray method, which probably accounts for the failure of previous investigators to identify the free-lime phase, even in the most basic slags.

In a number of slags the (Fe,Mn,Mg)O is not a single phase, but consists of a range of compositions which can be of two types, as follows:

(*a*) A continuous range of solid solutions between FeO and MnO.

(*b*) One, two or more discrete compositions, each revealing itself by a separate pattern of varying degree of intensity.

There are intermediate stages between both types, the latter being superimposed upon the former. The resulting appearances of lime groups are represented diagrammatically in Table XIX. At high temperatures the components of the RO phase, chiefly FeO, MnO and MgO, are completely intersoluble, but a miscibility gap probably occurs at lower temperatures, the rate of cooling consequently determining the degree of equilibrium attained. The variation in the rate of cooling from the outside to the inside of the slag sample will result in a corresponding degree of heterogeneity in the RO phase. Rapid rates of cooling across the single-phase/two-phase boundary would tend to preserve one distinct composition, and very slow cooling to establish equilibrium at a lower temperature corresponding to other distinct compositions at the opposite ends of the tie lines; intermediate bands may appear for intermediate rates. When the RO phase is rich in MnO small amounts of CaO may enter the solution, as observed in basic electric oxidising slags; the presence of CaO would tend to widen the miscibility gap at a given temperature, thus causing the separation of several distinct RO compositions. The miscibility gap bound to occur in the quaternary system CaO–MgO–MnO–FeO requires systematic investigation.

It is significant that only slags of batch *A* show this phenomenon of line-split of (Fe,Mn,Mg)O; these slags contain on the average more CaO than the

others, which lends support to the theory that small amounts of CaO enter the (Fe,Mn,Mg)O solution.

In contrast, the free CaO phase (with MnO in solution) does not show any tendency to split into several cubes and is generally of a homogeneous composition, as indicated by the sharp diffraction lines in the various batches. The lattice parameter is never far removed from that of pure CaO, indicating that comparative little MnO has entered the CaO lattice, the bulk of the MnO having dissolved in the (Fe,Mn,Mg)O solution. This is supported by the fact that the parameter of the latter phase is usually considerably above that for wüstite, though in some cases equal or lower.

The measured lattice parameters are shown in Table XIX. both for the (Fe,Mn,Mg)O and (Ca,Mn)O phases. The values for the pure compounds are as follows :

CaO	4·80 Å.
MnO	4·43 Å.
FeO	4·27–4·30 Å. (wüstite solid solution).
MgO	4·20 Å.

The order of magnitude of the observed parameters ranges from 4·71 to 4·77 Å. for (Ca,Mn)O and from 4·26 to 4·37 Å. for (Fe,Mn,Mg)O.

(2) *Nagelschmidtite* ($9CaO.P_2O_5.3SiO_2$–$7CaO.P_2O_5.2SiO_2$).

This compound is a prominent phase in all basic open-hearth slags containing P_2O_5 in which apatite is absent, or present to a lesser extent only. This is convincingly illustrated by batch *A* slags (Table XVI.). It would appear that nagelschmidtite and fluorapatite are competitive in their formation and that the slags can be divided into two broad groups—the nagelschmidtite and apatite types. Photometer records of the two compounds $9CaO.P_2O_5.3SiO_2$ and $7CaO.P_2O_5.2SiO_2$ are given in Fig. 20. The isomorphous character is evident, but there is a certain difference in relative intensities owing to the substitution of PO_4''' by SiO_4'''' groups. The balance of valencies would require some Ca^{++} ions to be simultaneously omitted from the lattice. Barrett and McCaughey [9] point out that the whole series of compounds from $2CaO.SiO_2$ to $3CaO.P_2O_5$ are successive modifications to one fundamental structure. The compositions may be expressed by the type formula :

$$Ca_{3+\frac{x}{2}}Si_xP_{2-x}O_8, \text{ where } 0 \leq x \leq 2.$$

$x = 0$	gives	$3CaO.P_2O_5$.
$x = 0{\cdot}67$	,,	Silico-carnotite.
$x = 1$	,,	$7CaO.P_2O_5.2SiO_2$.
$x = 1{\cdot}2$	,,	$9CaO.P_2O_5.3SiO_2$.
$x = 2$	,,	$2CaO.SiO_2$.

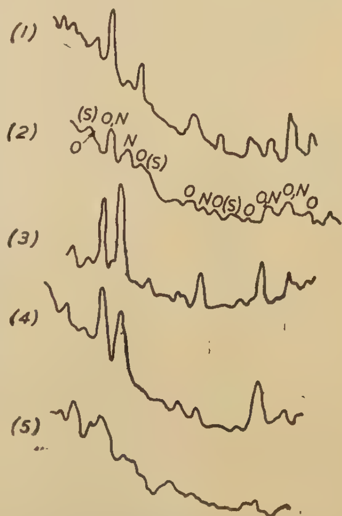
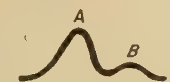
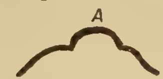
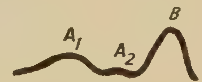
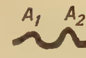
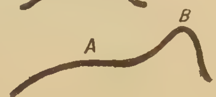


Fig. 20.—Photometer Curves of Phosphates and a Bessemer Slag. (1) Silico-carnotite ($5CaO.P_2O_5.SiO_2$). (2) Basic Bessemer slag. (3) Nagelschmidtite ($9CaO.P_2O_5.3SiO_2$). (4) Nagelschmidtite ($7CaO.P_2O_5.2SiO_2$). (5) $4CaO.P_2O_5$.

The actual structures of these compounds are not known, but the nagelschmidtite pattern, in spite of distinctive differences, shows a marked resemblance to that of merwinite. No difference in the lattice parameter between $7CaO.P_2O_5.2SiO_2$ and $9CaO.P_2O_5.3SiO_2$ is observed; similarly, Nagel-

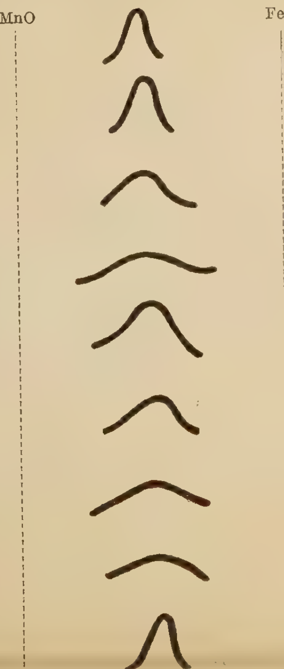
TABLE XIX.—*Basic Open-Hearth Slags.—Lattice Parameters and Character of RO Solid Solution Phases.*

Slag No.	(Fe,Mn)O.			(Ca,Mn)O.	
	Homogeneity of Composition (Description).	Diagrammatic Photometer Curves of RO Line-Groups. ⟶ Increasing $\sin^2\theta$.	Mean Lattice Parameter. Å.*	Homogeneity of Composition.	Mean Lattice Parameter. Å.
		Batch A. MnO ... FeO			
OH 1	Two distinct cubes, *A*, high MnO; *B*, low MnO. *A* stronger than *B*.		$4{\cdot}36_5$	...	...
OH 2	Narrow band of compositions.		$4{\cdot}36_0$	...	...
OH 3	Single cube, *A*, strong and sharp.		4·34 4·27	Single composition.	...
OH 4	Series of distinct compositions: A_1 and A_2 weak, *B* strong, near FeO.		...		...
OH 5	Cube, *A*, strong, band of compositions.		4·36		4·77
OH 6	Cube, *A*, strong, band of compositions fading towards FeO.		4·34		4·75
OH 7	Cube, *A*, strong, band of compositions fading towards FeO; similar to OH 6.		4·34		4·75
OH 8	Two cubes: Diffuse band *A*, high in MnO, terminating in single composition *B*, high in FeO.		4·34 4·27		4·73

OH 9	Cube, A, strong, wide band of compositions. Similar to OH 8, but no B.		4·36		$4{\cdot}75_0$
OH 10	Two cubes: A_1 and A_2 (split of A) closely adjoining, strong, superimposed on wide band.		4·36 4·34		$4{\cdot}75_0$
OH 11	Two cubes: A_1 and A_2, greater separation than in OH 10.		...		$4{\cdot}75_0$
OH 12	Cube, A, narrow band of compositions (cubes merging).		4·34		$4{\cdot}73_3$
OH 13	Single cube, A, sharp.		4·36	Band of compositions (diffuse).	$4{\cdot}73_0$
OH 14	Cube, A, narrow band of compositions.		4·31		$4{\cdot}75_0$
OH 15	Two cubes: A, diffuse band towards MnO from B, which is strong and well-defined (similar to OH 8).		4·33 4·26		$4{\cdot}74_0$
OH 16	Cube, A, medium, diffuse band. Similar to OH 15, but no B.		4·33	Single composition.	$4{\cdot}75_0$
OH 17	Cube, A, small, narrow range.		4·34		$4{\cdot}75_0$

(Continued overleaf.)

TABLE XIX.—*Continued.*

Slag No.	(Fe,Mn)O. Homogeneity of Composition (Description).	(Fe,Mn)O. Diagrammatic Photometer Curves of RO Line-Groups. ⟶ Increasing $\sin^2\theta$.	(Fe,Mn)O. Mean Lattice Parameter. Å.*	(Ca,Mn)O. Homogeneity of Composition.	(Ca,Mn)O. Mean Lattice Parameter. Å.
		Batch B.			
		MnO FeO			
OH 18	Single phase, sharp.		$4{\cdot}34_0$	...	...
OH 19	Single phase, sharp.		$4{\cdot}34_0$	...	...
OH 20	Narrow band of compositions.		$4{\cdot}30_5$	...	...
OH 21	Wider band of compositions, very diffuse.		$4{\cdot}30_5$	Single composition.	...
OH 22	Single phase, sharp.		$4{\cdot}36_5$		$4{\cdot}72_0$
OH 23	Single phase, moderately sharp.		$4{\cdot}31_5$		$4{\cdot}73_0$
OH 24	Narrow band.		$4{\cdot}37_6$		$4{\cdot}74_0$
OH 25	Narrow band.		$4{\cdot}30_0$		$4{\cdot}71_5$
OH 38	Single phase, sharp.		$4{\cdot}31_5$	...	...

Batch C.

OH 26				$4{\cdot}32_5$	...	...
OH 27				$4{\cdot}36_0$	...	...
OH 28	Single composition. (OH 26–OH 29)			$4{\cdot}35_5$	...	...
OH 29				$4{\cdot}35_5$	...	...
OH 30				$4{\cdot}38_5$	...	...
OH 31	Wide band of compositions. (OH 30–OH 32)			$4{\cdot}37_5$	Narrow band of compositions.	...
OH 32				$4{\cdot}37_5$		...
OH 33	Wide band of compositions terminating in distinct cube near FeO (similar to OH8).			$4{\cdot}27_5$		...
OH 34	Wide band of compositions.			Immeasurable	Single composition. (OH 32–OH 36)	...
OH 35	Narrow band of compositions.			$4{\cdot}38_5$		$4{\cdot}74_0$
OH 36	Wide band of compositions terminating in distinct cube near FeO (similar to OH 8 and OH 33).			$4{\cdot}28_5$		$4{\cdot}71_5$

* Left-hand figures refer to cube *A* (approaching MnO) and the right-hand figures to cube *B* (approaching FeO).

schmidt[8] found the inter-planar spacings of his $7CaO.P_2O_5.2SiO_2$ compound to agree with those of a compound supplied by Trömel of the ascribed composition $8CaO.P_2O_5.2.5SiO_2$. The interplanar spacings of the compound as determined by Nagelschmidt are as follows:

3·93 W	2·22 VW	1·668 W	1·346 M
3·51 W	1·960 S	1·591 W	1·260 VW
2·85 S	1·862 VW	1·562 W	1·220 VW
2·70 S	1·761 M	1·485 M	1·185 VW
2·33 VW			

In order to examine the possibility of solid solution between $2CaO.SiO_2$, merwinite and nagelschmidtite, sample mixes of (*a*) 50% nagelschmidtite and 50% $2CaO.SiO_2$, and (*b*) 50% nagelschmidtite and 50% $3CaO.MgO.2SiO_2$ were thoroughly mixed and fired in air at 1600° C. The results of the X-ray examination of these samples are set out in Table XX.

TABLE XX.—*Constitution of Mixes of Nagelschmidtite with $2CaO.SiO_2$ and $3CaO.MgO.2SiO_2$, as Revealed by X-Ray Examination.*

Sample.	Rate of Cooling.	Constitution (X-ray).
(*a*) 50% nagelschmidtite and 50% $2CaO.SiO_2$.	Quenched from 1600° C.	α'-$2CaO.SiO_2$ (large), nagelschmidtite (small).
	Slowly cooled in furnace.	α'-$2CaO.SiO_2$ (medium), nagelschmidtite (medium). The $2CaO.SiO_2$ appears to be transforming from α' to β.
(*b*) 50% nagelschmidtite and 50% $3CaO.MgO.2SiO_2$.	Quenched from 1600° C.	α'-$2CaO.SiO_2$ (large), nagelschmidtite (small). Identical with sample (*a*), quenched, except that spacing of α'-$2CaO.SiO_2$ is slightly smaller.
	Slowly cooled in furnace.	Nagelschmidtite (large), but pattern modified to approach that of merwinite.

These tests, although only explorative, indicate the following:

(*a*) Nagelschmidtite is able to dissolve in α'-$2CaO.SiO_2$, which can be preserved metastably by quenching, but which tends to decompose into β-$2CaO.SiO_2$ and nagelschmidtite on cooling.

As mentioned previously, Trömel[53] reported a compound "*K*" between nagelschmidtite (Trömel's compound "*R*") and $2CaO.SiO_2$, having a potassium-sulphate type of structure. Comparison with actual K_2SO_4 by the authors confirms a fair agreement between the K_2SO_4 and the α'-$2CaO.SiO_2$ patterns, except for a considerably smaller unit cell of the α'-$2CaO.SiO_2$ lattice. Thus Trömel's "*K*" phase appears to be not an independent phase, but a solid solution of nagelschmidtite in α'-$2CaO.SiO_2$.

The basic open-hearth slags investigated by the authors do not, however, contain the α'-$2CaO.SiO_2$ solid solution, cooling in these cases having been slow enough to allow decomposition.

(*b*) Intermediate stages between the nagelschmidtite and the merwinite structures exist; the compounds are not, of course, isomorphous, but successive distortions of similar unit cells are liable to occur while SiO_4 tetrahedral groups replace PO_4, and some Mg^{++} substitutes Ca^{++}.

In several basic open-hearth slags irregularities in the nagelschmidtite patterns indicate such modifications, but to a much lesser extent.

The equilibria between $2CaO.SiO_2$, $3CaO.MgO.2SiO_2$, $3CaO.MnO.2SiO_2$

and nagelschmidtite require detailed investigation, but from the point of view of structure these compounds appear to be intimately related. Writing formally the compositions of these compounds as A_2BO_4, where A = Ca, Mn, Fe, Mg, and B = Si, P, it is evident that the structures of the compounds can differ slightly from each other but belong to the same fundamental crystal system, and considerable mutual solubility is possible, provided that " vacant lattice sites " are assumed.

A powder photograph of nagelschmidtite, together with a nagelschmidtite-bearing slag (OH 2) is shown in Fig. 13.

(3) *Dicalcium Silicate* ($2CaO.SiO_2$).

Dicalcium silicate was found in the majority of these basic open-hearth slags, but, unlike the case of basic electric oxidising slags, it is usually a minor phase, owing to the fact that the presence of P_2O_5 results in the formation of nagelschmidtite at the expense of $2CaO.SiO_2$. In each case only the β form is observed. The occurrence of $2CaO.SiO_2$ in basic open-hearth slags has been reported by previous investigators.[2] The refractive indices of $2CaO.SiO_2$ vary noticeably from slag to slag, being lowered by the substitution of silicon by phosphorus.[1] The γ form of $2CaO.SiO_2$ does not occur, because the solution of nagelschmidtite apparently suppresses the β-to-γ transformation.

(4) *Merwinite* ($3CaO.RO.2SiO_2$).

Merwinite only occurs in a few of the less basic slags as predicated by the system of phase assemblages; the co-existing phases are also in agreement with the phase data. In basic open-hearth slags with low P_2O_5 contents and appropriate CaO/SiO_2 ratios, merwinite can be a major phase, as in some basic electric oxidising slags. The exact composition of the RO part of this compound occurring in slags is not known, but is probably largely MgO with a little substitution of manganese and iron.

The P_2O_5 present in normal basic open-hearth slags combines with CaO and SiO_2 to form calcium silico-phosphates at the expense of merwinite. Hence, in normal basic open-hearth slags either the CaO/SiO_2 ratio or the P_2O_5 content is too high for merwinite to be a major phase. Accordingly, it is not surprising that earlier investigators[2] failed to detect it in open-hearth slags, probably confusing it with $2CaO.SiO_2$ and $CaO.RO.SiO_2$. Merwinite also undergoes a peritectic dissociation into $2CaO.SiO_2$ and liquid at 1590° C., and in complex slags at even lower temperatures; consequently, the formation of merwinite may be suppressed by very rapid cooling from the liquid state. Mason,[1] however, identified merwinite in some basic slags with low P_2O_5 contents in which merwinite was the major phase.

(5) *Brownmillerite Solid Solution* ($4CaO.Al_2O_3.Fe_2O_3$–$2CaO.Fe_2O_3$).

Since there is usually a preponderance of $2CaO.Fe_2O_3$ in this solid solution of slags, this phase is referred to as dicalcium ferrite. It is generally a subsidiary phase in the more basic slags, disappearing with the occurrence of merwinite. Owing to the coincidence of the stronger lines of its pattern with those of other phases, some doubt is attached to its identification in certain cases, particularly when the amount is small. Large amounts of $2CaO.Fe_2O_3$ were only detected in three slags (Nos. 39, 40 and 41). Mason[1] detected a yellow-brown birefringent phase in a few slags which answered to the description of $2CaO.Fe_2O_3$, but positive proof of its identification was lacking. The compound was previously isolated from basic slags and its frequent occurrence, particularly towards the end of the process, is generally accepted.[2]

(6) *Hilgenstockite* ($4CaO.P_2O_5$).

Hilgenstockite produces a weak powder pattern very rich in lines, and, consequently, small amounts are difficult to identify in basic slags by the X-ray method. It was observed in a few basic slags with the lowest SiO_2/P_2O_5 ratios. It is not detected in open-hearth slags in which the SiO_2/P_2O_5 ratio is greater than 180 : 142. The major co-existing phases detected in these slags are CaO, apatite and nagelschmidtite, which agrees with the system of phase assemblages. Small amounts of $2CaO.SiO_2$, however, also occur, which is at variance with the system. Hilgenstockite does not frequently occur in basic open-hearth slags because the SiO/P_2O_5 ratios are usually too high, but separation and analysis of this compound from Bessemer slags has been carried out several times.[2]

Trömel[53] has remarked upon the ready tendency to hydration shown by this compound; the normal moisture content of the atmosphere is sufficient to form the hydroxy-apatite at as high a temperature as 1100° C., according to the equation :

$$3(4CaO.P_2O_5) + H_2O \rightarrow 9CaO.Ca(OH)_2.3P_2O_5 + 2CaO.$$

Schneiderhohn[51] found, in some of his slags, CaO in a fine state of dispersion, which had been formed by the action of moisture on $4CaO.P_2O_5$ during cooling.

(7) *Apatite.*

Fluorapatite ($9CaO.CaF_2.3P_2O_5$) occurs in the majority of the slags (*see* Table XVI.). Fig. 21 shows a photometer record of fluorapatite, together with a basic open-hearth slag (OH 24) in which this compound is prominent. The X-ray pattern of fluorapatite and slags (OH 9 and OH 16) containing it is reproduced in Fig. 13. Apatite, which has a very low citric-acid solubility, is characteristic of all slags to which fluorspar has been added. If sufficient CaF_2 is present, all the P_2O_5 of the slag occurs as apatite; if not, apatite is found to co-exist with normal phosphates as outlined in the system of phase assemblages.

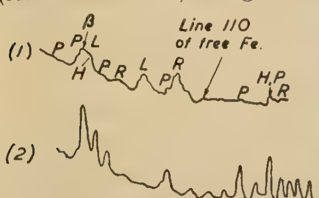


Fig. 21.—Photometer Curves. (1) Basic open-hearth slag, containing fluorapatite, (Ca,Mn)O, (Fe,Mn,Mg)O, β-$2CaO.SiO_2$ and $2CaO.Fe_2O_3$. (2) Fluorapatite.

Bainbridge,[56] who investigated the influence of fluorspar on the citric-acid solubility of phosphate slags, proved that the following reactions occur :

$$(a)\ 3(4CaO.P_2O_5) + CaF_2 \rightarrow 9CaO.CaF_2.3P_2O_5 + 3CaO.$$
$$(b)\ 3(5CaO.P_2O_5.SiO_2) + CaF_2 \rightarrow 9CaO.CaF_2.3P_2O_5 + 3(2CaO.SiO_2).$$

With increasing addition of CaF_2 to these compounds, the citric-acid solubility decreased progressively, reaching a minimum and, thereafter, constant value at a $CaF_2/3CaO.P_2O_5$ ratio of approximately 8 : 100, this ratio in pure fluorapatite being 94 : 1042. Bainbridge found similar relationships when CaF_2 was added to a citric-acid-soluble slag of known P_2O_5 content.

Evidently, the following reactions also occur :

$$(c)\ 3(9CaO.P_2O_5.3SiO_2) + CaF_2 \rightarrow 9CaO.CaF_2.3P_2O_5 + 9(2CaO.SiO_2).$$
$$(d)\ 3(7CaO.P_2O_5.2SiO_2) + CaF_2 \rightarrow 9CaO.CaF_2.3P_2O_5 + 6(2CaO.SiO_2).$$

The tentative phase diagram for the system CaO–$3CaO.P_2O_5$–$2CaO.SiO_2$–CaF_2 proposed by White[2] and based on the observations of Bainbridge[56]

has been modified in Fig. 22 according to the recent investigation of the CaO–SiO_2–P_2O_5 system.[9]

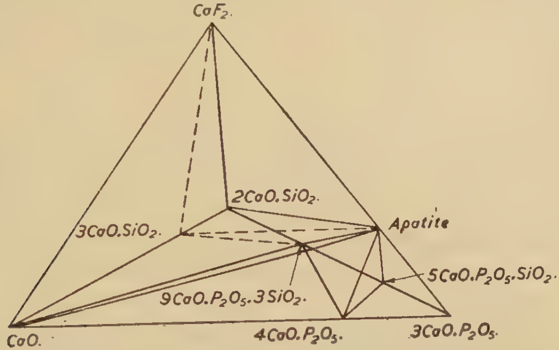


FIG. 22.—Modified Tentative Phase Diagram for the System CaO–$3CaO.P_2O_5$–$2CaO.SiO_2$–CaF_2.

According to Mehmel[57] the crystal structure of fluorapatite is hexagonal with $a = 9{\cdot}36$ Å. and $c = 6{\cdot}88$ Å., the PO_4''' groups being tetrahedral like SiO_4''''. Körber and Trömel found the patterns of the fluor- and hydroxy-apatites to be virtually identical, which is not unexpected in view of the practically identical sizes of the radii of $(OH)^-$ and F^- ions (1·33 Å.). Hence, the identification of apatite in basic slags is not a criterion for the presence of fluorine, since the apatite may have been formed by the hydration of hilgenstockite. However, since fluorine was detected chemically in the great majority of the open-hearth slags examined by the authors, the bulk of the apatite phase is undoubtedly the fluorapatite, with only small, if any, substitution by $(OH)^-$ groups.

(8) *Silico-carnotite* ($5CaO.P_2O_5.SiO_2$).

This compound, which has been identified in basic slags by many previous workers,[2] only occurred in one of the basic open-hearth slags (OH 34) which had a low SiO_2/P_2O_5 ratio. Silico-carnotite does not occur in the usual type of basic open-hearth slag because the SiO_2/P_2O_5 ratio is too high, but is one of the major phases in highly phosphoric slags, *e.g.*, Bessemer slags. The phases co-existing with silico-carnotite in slag OH 34 are apatite, $4CaO.P_2O_5$, (Fe,Mn,Mg)O, and free CaO. The co-existence of free CaO is at variance with the recent findings of Barrett and McCaughey,[9] but in agreement with the phase data accepted by White.[2] As discussed in an earlier section, however, the free CaO may have been formed by the partial hydration of $4CaO.P_2O_5$, or departure from equilibrium has been caused by too-rapid cooling.

A large amount of silico-carnotite was identified in a basic Bessemer slag which, in addition, contained a large amount of nagelschmidtite and a trace of spinel. The photometer records in Fig. 20 illustrate this constitution. Free CaO and (Fe,Mn,Mg)O were not detected. The occurrence of silico-carnotite and nagelschmidtite as the chief phases in this slag is in line with the SiO_2/P_2O_5 ratio, but the complete absence of (Fe,Mn,Mg)O is difficult to reconcile with its composition. To test whether any free iron or very highly dispersed FeO may have escaped detection, the slag was examined by the X-ray method after firing at 1200° C. for 2 hr. Metallic iron and FeO were not visible, however. The bulk of the RO is probably

combined in the spinel, but some RO may also have entered the silico-phosphate lattices. The spacing of the silico-carnotite in the slag is slightly greater than that of the synthetic compound, thus indicating solid solution.

A photometer record of silico-carnotite and of the basic Bessemer slag (BB 42) is shown in Fig. 20.

(9) *Metallic Iron.*

Traces of metallic α-iron (less than 1%) were detected in a few slags. The presence of iron introduces an error in the chemical analysis, the FeO being estimated too high and the Fe_2O_3 too low.

(10) *Calcium Hydroxide.*

Small amounts of $Ca(OH)_2$ were observed in a few slags containing free CaO as a major phase. Hydration of the free CaO occurred during storage.

(11) *Other Phases.*

A number of other phases reported by earlier workers [2] but not detected by the authors are worthy of consideration.

3CaO.Fe₂O₃.—Stead and Ridsdale [58] and Ferguson [59] claimed to have isolated crystals from basic slag corresponding to the composition $3CaO.Fe_2O_3$, but actually containing, in addition to CaO and Fe_2O_3, appreciable amounts of FeO, MnO, MgO and Al_2O_3. Nagai and Asaoka [60] claim to have established the existence of $3CaO.Fe_2O_3$ but they state that it dissociates at high temperatures into CaO and $2CaO.Fe_2O_3$. This compound has not been found however in the $CaO–Fe_2O_3$ system,[61, 62] cement [29] or dolomite clinkers,[5]. Further, in a recent examination of basic-oxide systems containing Fe_2O_3* evidence for the occurrence of this compound was not obtained. It must be concluded that the weight of evidence is against the occurrence of $3CaO.Fe_2O_3$.

Spinels.—White [2] stated that there is some doubt as to whether true spinels will normally occur in basic slags, as they do not, in spite of their characteristic crystalline form, appear to have been identified with certainty. Mason,[1] however, identified spinels in a few basic open-hearth slags, and it has been noted in a previous section that spinels occur in basic electric slags. Nevertheless, spinels were not detected in the basic open-hearth slags (*see* Table XVII.), because they are too basic, and, as shown in the Table of phase assemblages, spinels only appear at the lower basicities.

Phosphates. — Steadite ($3CaO.P_2O_5$), octo-basic phosphate ($6CaO.P_2O_5.SiO_2.2FeO.SiO_2$) and substituted silico-phosphates, the occurrence of which in certain highly phosphoric basic slags has been reported,[2] were not detected. These phases do not occur in normal basic open-hearth slags.

Slag Control.

The constitution and properties of basic slag and its influence on the steelmaking process has received much attention in the past. It is accepted that for efficient production the constitution of the slag should be controlled within fairly close limits. When the CaO content of the slag is too low phosphorus and sulphur removal from the steel is inadequate, and the slag, being unsaturated with basic oxide, tends to dissolve CaO and MgO from the basic hearth, thus resulting in excessive corrosion and consequent loss of output due to increased fettling. When the slag contains too much lime it tends to be "thick," necessitating the addition of fluorspar to increase its fluidity, thus destroying its value as a fertiliser.

* J. R. Rait, unpublished work.

The constitutions of normal basic open-hearth slags summarised in Table XVII. show a fairly wide scatter, but a method of control is suggested by these results. At the lower lime contents all the lime is in combination as silicates and silico-phosphates, but, with increasing basicity, a stage is reached when free CaO appears, the amount of free CaO increasing steadily with further increasing basicity; the ferrous oxide is uncombined and occurs as an independent phase with MgO and MnO in solution. Thus, a satisfactory normal basic open-hearth slag should show free CaO and (Fe,Mn,Mg)O in addition to the silicates and silico-phosphates. These two free-oxide phases are readily identified by the X-ray powder method and, further, an approximate estimate of these amounts is possible. Thus, by determining the free CaO and RO phases in a number of slags by the X-ray powder method and correlating these results with their steelmaking properties, the ideal range of free CaO and RO contents as obtained by the X-ray method can be established, and, by X-ray examination of the slag for free CaO and FeO during the process, additions of lime and ore may be made as required.

The co-existence of (Ca,Mn)O and (Fe,Mn,Mg)O in basic open-hearth slags is illustrated in Fig. 23, which shows photometer records of a sequence

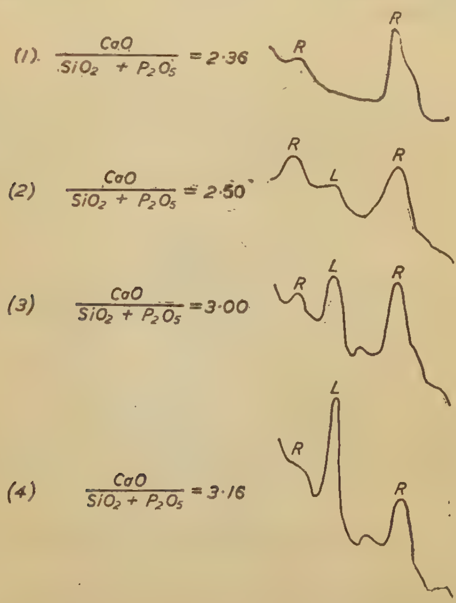


FIG. 23.—Photometer Curves of Basic Open-Hearth Slags, illustrating the increase of the (Ca,Mn)O phase with increasing basicity.

of slags of no free lime, and of trace, medium and high lime contents. Applied with due caution, the X-ray powder method would form a perfectly sound method of control, provided that the technique could be devised to obtain results rapidly—within, say, 10 min.

The exposure time in ordinary powder photography can be reduced by special means such as the rotating-anticathode tube, or, for a given X-ray set, by the use of the smallest-diameter Debye-Scherrer camera consistent

with sufficient resolution, omission of β-filter, use of intensifying screens and special rapid films and developers.

Instead of photography an ionisation-chamber method similar to that used in Bragg's single-crystal spectrometer offers the possibility of almost instantaneously detecting the presence of phases such as CaO and FeO in basic open-hearth slags, but its successful application depends on whether, by sufficient amplification, the diffraction pattern of a given phase can be recorded without ambiguity.

In basic open-hearth slags, if the answer could be gained within 30 min. after sampling, X-ray control could be usefully applied, whereas for basic electric slags the answer would be required more rapidly. In this case slag-control can also be applied to the presence of either $2CaO.SiO_2$ or merwinite in reducing slags, for instance, except that the weak nature of the patterns militates against speed. The authors have, however, carried out a series of short-time-exposure experiments on these slags, and the result can be stated to be promising, interpretation being, under certain conditions, already possible after 5 min. exposure. These tests, which are being continued, apply in their conclusions, of course, generally to any two-, or more, component mixture, and in the case of FeO and CaO in basic open-hearth slags, the position is even more favourable because of the high symmetry of their structures.

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[This paper was discussed jointly with the following two by Y. K. Zea on "The Phosphorus Reaction in Basic Open-Hearth Practice" and by R. H. Jay on "A Study of the Basic Open-Hearth Process, with Particular Reference to Slag Constitution."]

THE PHOSPHORUS REACTION IN BASIC OPEN-HEARTH PRACTICE.*

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Synopsis.

The practical applications of the equations for calculating the phosphorus content of the metal on the basis of the slag composition and the bath temperature, established by four investigators, are examined in relation to new temperature data and slag and metal analyses obtained from fifteen casts of basic open-hearth steel. It is found that Schenck and Riess's method gives the best results.

The reversion of phosphorus from slag to metal during casting was studied on twenty-five casts of basic open-hearth steel, firstly by graphical examination of the slag and metal analyses together with the bath temperature, and secondly by the application of Schenck's equilibrium diagrams. It is found that for a given set of working conditions, the rephosphorisation likely to occur in the ladle may be estimated quantitatively by the use of Schenck's equilibrium diagrams to establish a ratio of free lime to free iron oxide concentration in the final furnace slag from its composition and the bath temperature.

The results of this investigation show that rephosphorisation during casting is due to a change in the composition of the slag by enrichment of the silica content resulting from its reaction with the fireclay lining of the ladle. It is found that rephosphorisation does not occur when the ladle is lined with basic material; and that the degree of rephosphorisation, when a fireclay-brick-lined ladle is used, may be controlled to a large extent by keeping the temperature of the slag and metal in the ladle as low as possible, and secondly, by adjusting the slag composition so that, in spite of the reaction with the ladle lining, the basicity of the slag will not fall below a certain minimum necessary for holding phosphorus. The latter aim can be achieved, when circumstances permit, by increasing the lime content of the slag.

Introduction.

The complex nature of steelmaking slags and the lack of information concerning the relative stability of their components at steelmaking temperatures constitute the main difficulties in the study of slag/metal reactions. This is particularly true of the basic open-hearth process, in which reactions involving the removal of phosphorus and carbon and deoxidation overlap one another. In the case of the phosphorus reaction, it is generally assumed that equilibrium is almost reached at the end of refining period; and this assumption is essential to any attempt to estimate the distribution of phosphorus between the slag and metal in the bath. Moreover, the importance of accurate temperature measurement can hardly be over-emphasised, since lack of a precise temperature record leads to misinterpretation and miscalculation. The difficulty of obtaining accurate measurements, however, has been largely overcome by the recent development of the quick-immersion pyrometer.

Our present knowledge of the chemical constitution of slags has been largely obtained by means of the petrographic and X-ray examination of solidified samples, along with a study of the thermal equilibrium diagrams of the slag-forming oxides. It is generally agreed, however, that in attempting to explain the reactions involved in steelmaking processes on

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these bases, certain limitations must be taken into account; these are due primarily to the fact that whilst equilibrium diagrams provide useful information on the relationship between the melting points of oxide mixtures and their compositions, they give no certain indication of possible molecular combinations above the liquidus. The information obtained by petrographic and X-ray examination of solidified slags is equally open to criticism on the grounds of the possible dissociation of certain compounds at steelmaking temperatures.

It is true that the open-hearth process is governed by the physico-chemical behaviour of the slag components in contact with metal at steel-making temperatures; and because of the many factors involved, the simplest and most practical way of tackling this problem is to assume some theoretical basis. A comprehensive survey of this subject has been made by Schenck and Riess,[1] who by means of an ingenious correlation of practical data based on a sound physico-chemical knowledge, have worked out empirically a general guide for the study of basic open-hearth slag components. For convenience, their work may be adopted as a preliminary guide to the interpretation of various slag reactions on theoretical lines.

Schenck and Riess's work can be briefly described as an extensive study of basic open-hearth melts combined with the laboratory data of Körber and Oelsen[2] and the acid open-hearth plant data of Schenck and Brüggemann,[3] together with the best available information on heats of formation of the various compounds. From this study they attempted to define the modes of combination of lime, silica and iron oxides, and, further, determined the equilibria of the phosphorus reaction. From their results and plant data they compiled a series of equilibrium diagrams by means of which the free concentrations of lime and iron oxide may be obtained for a basic slag at a given temperature. They concluded that the logarithm of the constant for the phosphorus reaction is given by the equation:

$$\text{Log } K_p^{\text{IV}} = \frac{\log [\Sigma\text{P}] . (\text{FeO})^5 . (\text{CaO})^4}{(\Sigma\text{P}_2\text{O}_5)} + 0{\cdot}06\,(\Sigma\text{P}_2\text{O}_5)$$
$$= -\frac{51{,}800}{T} + 35{\cdot}05,$$

where T = absolute temperature.

Hence:

$$\text{Log } [\Sigma\text{P}] = \log K_p^{\text{IV}} + \log (\Sigma\text{P}_2\text{O}_5) - 5 \log (\text{FeO}) - 4 \log (\text{CaO}) - 0{\cdot}06(\Sigma\text{P}_2\text{O}_5).$$

From these equations the phosphorus content of the metal can thus be calculated for a given slag composition at temperatures between 1527° and 1627° C.

In spite of certain analogies and assumptions made in evaluating the free concentrations of lime and iron oxide in basic slags, this work appears to offer a means of approach which might also be extended to cover the phenomenon of ladle rephosphorisation.

The distribution of phosphorus between slag and metal in the basic open-hearth steelmaking process has been studied by many investigators. A review of the literature on this subject[4 to 17] showed that these authors dealt mainly with the removal of phosphorus from the metal in the furnace; and that very little had been written on the reversion of phosphorus from the slag to the metal, either in the furnace or in the ladle. References to rephosphorisation of the metal in the literature were seldom accompanied by practical data to support the theory that was put forward.

Rephosphorisation of the metal to an appreciable extent is not a frequent occurrence in present-day steelmaking, but a small degree of phosphorus reversion to the metal often takes place during deoxidation in the furnace. More serious rephosphorisation takes place in the teeming ladle; and it

usually occurs during the teeming of the last few ingots. In some instances, the reversion of phosphorus from the slag in the ladle commences early in teeming; so that more than half the cast may be thrown out of the phosphorus specification limits.

The present work consists of two parts. In the first, use is made of new data regarding bath temperatures and slag and metal analyses in order to examine the relative merits of the theories of dephosphorisation of steel in the furnace, as advanced by various investigators. The second part is an attempt to examine the conditions under which rephosphorisation takes place in the basic open-hearth steelmaking process (*a*) during the finishing stages in the furnace, and (*b*) in the teeming ladle.

Practical Application of the Phosphorus-Reaction Equations.

In the absence of a precise knowledge of the constitution of basic slags in open-hearth furnaces at steelmaking temperatures, it is difficult to assess the relative merits of the dephosphorisation theories put forward by various investigators. However, the papers of Whiteley,[4] Herty,[11] Maurer and Bischof,[14] and Schenck and Riess[1] each provide a means of calculating the phosphorus content of the metal that is in equilibrium with a slag of given composition at a given temperature. It is interesting, therefore, to work out the theoretical phosphorus content of the metal in the present series of open-hearth casts by each of the four methods suggested by the above-mentioned authors, and to compare the theoretical results with the phosphorus contents of the bath as determined by analysis. The soundness of each of the dephosphorisation theories may be judged, with some degree of fairness, by its accuracy in practical application.

Slag and metal samples and temperatures taken by the quick-immersion pyrometer of fifteen casts of basic open-hearth steel towards the end of the refining period and at tapping are used in the following calculations. One example of each of the four methods employed for phosphorus calculations is shown below for samples taken during the refining period on Cast 34/2238:

Slag Analysis.	SiO_2.	Fe.	CaO.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.
Wt.-% . . .	8·46	13·5	51·7	5·04	5·74	6·49	12·8	5·1

Metal Analysis.	C.	Si.	S.	P.	Mn.
Wt.-% . . .	0·13	0·001	0·045	0·020	0·17

Bath temperature : 1623° C.

Method (A), *Based on Whiteley's Paper.*[4]

		Analysis.	Molecular Wt.	Molecular Ratio	Sum.
Bases :	FeO . . .	12·8%	71·84	0·17817	1·34194
	MnO . . .	5·74%	70·93	0·08092	
	CaO . . .	51·7%	56·08	0·92189	
	MgO . . .	6·49%	40·32	0·16096	
Acids :	SiO_2 . . .	8·46%	60·06	0·14085	
	P_2O_5 . . .	5·04%	142·04	0·03548	

$$\text{Basicity Index} = \frac{\text{molecular ratio of available bases}}{\text{molecular ratio of } SiO_2}$$

$$= \frac{1{\cdot}34194 - (3 \times 0{\cdot}03548)}{0{\cdot}14085}$$

$$= 8{\cdot}772$$

The distribution factor $\frac{(P)_{slag}}{[P]_{metal}}$ corresponding to the basicity index of 8·772 may be read off from one of Whiteley's graphs. On examining this graph, however, it was found that the scale used for the distribution factor was rather small for accurate extrapolation; also that the position

TABLE I.—*Deviation in Phosphorus Content of Metal as Determined by Chemical Analysis and as Calculated from the Relation between the Basicity Index and the Distribution Factor Given by Whiteley.*

Cast No.	Basicity Index.	Calculated Distribution Factor.	P in Slag. %.	P in Metal. %. Calc.	P in Metal. %. Analysed.	P Discrepancy. %.
Sampled towards end of refining:						
34/2238	8·772	358·39	2·2025	0·006	0·020	0·014
30/4600	6·210	218·39	3·7014	0·017	0·025	0·008
31/501	5·574	183·64	2·3990	0·013	0·018	0·005
31/508	5·950	204·18	1·4990	0·007	0·012	0·005
35/7148	5·941	203·69	3·3037	0·016	0·024	0·008
35/7156	5·997	206·75	2·4035	0·012	0·030	0·018
31/599	7·210	273·04	2·4820	0·009	0·023	0·014
31/572	6·266	221·45	2·5302	0·011	0·024	0·013
30/4591	4·814	142·11	4·0030	0·028	0·031	0·003
25/1518	8·044	318·61	3·1027	0·010	0·023	0·013
25/1521	7·328	279·49	2·3642	0·008	0·016	0·008
23/3736	7·324	279·27	3·7407	0·013	0·021	0·008
25/1527	5·824	197·30	3·7320	0·019	0·029	0·010
25/1528	4·455	122·49	3·0153	0·024	0·018	0·006
22/2309	7·495	288·61	2·9497	0·010	0·022	0·012
Sampled on tapping:						
34/2238	8·690	353·91	2·0010	0·006	0·018	0·012
30/4600	6·805	250·90	3·3560	0·013	0·020	0·007
31/501	5·680	189·43	2·4166	0·013	0·021	0·008
31/508	5·900	201·45	1·5295	0·008	0·016	0·008
35/7148	6·558	237·41	2·9017	0·012	0·027	0·015
35/7156	7·350	280·69	2·0010	0·007	0·016	0·009
31/599	7·385	282·60	2·3250	0·008	0·011	0·003
31/572	6·947	258·69	2·3816	0·009	0·022	0·013
25/1518	8·921	366·53	3·0153	0·008	0·028	0·020
25/1521	7·092	266·59	2·3554	0·009	0·018	0·009
23/3736	7·114	267·79	3·6839	0·014	0·022	0·008
25/1527	5·734	192·38	3·7014	0·019	0·039	0·020
25/1528	4·582	129·43	3·0328	0·023	0·018	0·005
22/2309	7·390	282·87	2·8317	0·010	0·027	0·017

Discrepancy: Maximum . . . . . 0·020% P.
Minimum . . . . . 0·003% P.
Mean . . . . . 0·0103% P.
Number of Samples 29
Standard Deviation . . . . . 0·0047

of the line drawn through the experimental points was fixed empirically. In order to do justice to Whiteley's work in relation to the work of other investigators, the author has made use of Whiteley's experimental data shown in his Table II. to work out an equation of the best line (in the mathematical sense) that can be drawn through the experimental points. The method of least squares was used and the following equation was obtained:

$$\text{Distribution Factor} = \frac{\text{Basicity Index} - 2{\cdot}2134}{0{\cdot}0183}$$

Substituting the value 8·772, obtained above for the basicity index, in this equation:

$$\text{Distribution Factor} = \frac{8{\cdot}772 - 2{\cdot}2134}{0{\cdot}0183} = 358{\cdot}39.$$

The P_2O_5 content of the slag is converted to phosphorus in the slag by the factor 0·437, *i.e.*, 5·04% P_2O_5 ≡ 2·2025% P in the slag. The phosphorus content of the metal equals the phosphorus in the slag divided by the distribution factor,

$$i.e., [P]_{metal} = \frac{2{\cdot}2025}{358{\cdot}39} = 0{\cdot}0061\%.$$

Method (B), *based on Herty's Paper.*[11]

Herty states that the amount of phosphorus in the metal depends on the total phosphorus in the charge, the iron oxide content of the slag, the basicity of the slag, the slag volume and the temperature; and that the quantitative relationship between these variables is :

$$K_p = \frac{(P_2O_5)_{slag}}{[P]^2_{metal}\,(FeO)^5_{slag}\,(CaO)^3_{slag}}$$

and
$$\log (K_p \times 10^{-5}) = \left[\left(\frac{1}{T} \times 10^5\right) - 28{\cdot}52\right]\left[\frac{1}{1{\cdot}322}\right],$$

when slag concentrations are expressed as mols per 100 weight units, metal concentrations as weights % and the temperature T as ° F. absolute.

In this example (Cast 34/2238—refining period), $t = 1623°$ C.; *i.e.*, $T = 3413{\cdot}4°$ F. absolute.

Thus
$$\begin{aligned}
\log (1/T \times 10^5) &= \log 1 - \log T + 5 \log 10,\\
&= 0 - 3{\cdot}5332 + 5(1),\\
&= 1{\cdot}4668\\
i.e., \quad (1/T \times 10^5) &= 29{\cdot}29\\
\therefore \quad [(1/T \times 10^5) - 28{\cdot}52] &= 0{\cdot}77\\
\log [(1/T \times 10^5) - 28{\cdot}52][1/1{\cdot}322] &= \log 0{\cdot}77 + \log 1 - \log 1{\cdot}322\\
&= \bar{1}{\cdot}8865 + 0 - 0{\cdot}1212\\
&= \bar{1}{\cdot}7653\\
i.e., \quad [(1/T \times 10^5) - 28{\cdot}52][1/1{\cdot}322] &= 0{\cdot}5825 = \log (K_p \times 10^{-5})\\
i.e., \quad 0{\cdot}5825 &= \log K_p - 5 \log 10\\
\therefore \quad \log K_p &= 5{\cdot}5825\\
\therefore \quad K_p &= 382{,}300
\end{aligned}$$

Now,
$$\begin{aligned}
(P_2O_5)\text{ weight-}\% &= 5{\cdot}04,\\
i.e., \quad (P_2O_5)\text{ mols. per 100 weight units of slag} &= \frac{(P_2O_5)\text{ wt.-}\%}{\text{Molecular wt. of } P_2O_5}\\
&= 5{\cdot}04/142{\cdot}04\\
&= 0{\cdot}03548\\
\therefore \quad \log (P_2O_5) &= \bar{2}{\cdot}5500\\
(CaO)\text{ weight-}\% &= 51{\cdot}7,\\
i.e., \quad (CaO)\text{ mols. per 100 weight units of slag} &= \frac{(CaO)\text{ wt.-}\%}{\text{Molecular wt. of CaO}}\\
&= 51{\cdot}7/56{\cdot}08\\
&= 0{\cdot}92189\\
\therefore \quad 3 \log (CaO) &= \bar{1}{\cdot}8941.\\
(FeO)\text{ weight-}\% &= 12{\cdot}8 \text{ and } (Fe_2O_3)\text{ weight-}\% = 5{\cdot}1.
\end{aligned}$$

In his calculations, Herty assumed that the Fe_2O_3 found in slags is reduced by iron to FeO according to the equation :

$$Fe_2O_3 + Fe = 3FeO.$$

Thus Herty would assume the slag under consideration to contain 12·8 + (5·1 × 215·4/159·7) = 19·68% FeO by weight.

$$\begin{aligned}
\therefore \quad (FeO)\text{ mols. per 100 weight units of slag} &= \frac{(FeO)\text{ wt.-}\%}{\text{Molecular wt. of FeO}}\\
&= \frac{19{\cdot}68}{71{\cdot}84} = 0{\cdot}27394\\
i.e., \quad 5 \log (FeO) &= \bar{3}{\cdot}1885.
\end{aligned}$$

Substituting the above values in Herty's equation:

$$K_p = \frac{(P_2O_5)}{[P]^2 . (FeO)^5 . (CaO)^3}$$

or, $\log [P] = \frac{1}{2}[\log (P_2O_5) - \log K_p - 5 \log (FeO) - 3 \log (CaO)]$

thus, $\log [P] = \frac{1}{2}[2{\cdot}5500 - 5{\cdot}5825 - \bar{3}{\cdot}1885 - \bar{1}{\cdot}8941]$

$= \frac{1}{2}[\bar{5}{\cdot}8849] = \bar{3}{\cdot}94245$

$= \log 0{\cdot}00876$

$\therefore \quad [P]_{metal} = 0{\cdot}009\%.$

Herty stated that the equation:

$$\log K_{Mn}'' = [(1/T \times 10^5) - 28{\cdot}25][1/5],$$

where $K_{Mn}'' = K_{Mn}' \times (\text{available base})^{0{\cdot}65}$

and $K_{Mn}' = \frac{(MnO)_{slag}}{[Mn]_{metal} \times (FeO)_{slag}}$,

which he used in calculating the bath temperatures incorporated in the equation representing the phosphorus equilibrium, would hold only for slags containing not more than 5% of phosphoric acid. However, the

TABLE II.—*Deviation in Phosphorus Content of Metal as Determined by Chemical Analysis and as Calculated from Herty's Work.*

Cast No.	Temp. °F.	$K_p \times 10^{-5}$.	P_2O_5 in Slag. %.	P in Metal. %. Calc.	P in Metal. %. Analysed.	P Discrepancy. %.
Sampled towards end of refining:						
34/2238	2953	3·823	5·04	0·009	0·020	0·011
30/4600	2822	29·850	8·47	0·005	0·025	0·020
31/501	2935	5·054	5·49	0·012	0·018	0·006
31/508	2946	4·319	3·43	0·009	0·012	0·003
35/7148	2887	10·690	7·56	0·008	0·024	0·016
35/7156	2858	16·830	5·50	0·003	0·030	0·027
31/599	2865	15·140	5·68	0·004	0·023	0·019
31/572	2924	6·013	5·79	0·012	0·024	0·012
30/4591	2915	6·913	9·16	0·012	0·031	0·019
25/1518	2948	4·172	7·10	0·013	0·023	0·010
25/1521	2914	7·034	5·41	0·010	0·016	0·006
23/3736	2908	7·676	8·56	0·013	0·021	0·008
25/1527	2930	5·418	8·54	0·024	0·029	0·005
25/1528	2833	25·120	6·90	0·009	0·018	0·009
22/2309	2906	7·945	6·75	0·010	0·022	0·012
Sampled on tapping:						
34/2238	2948	4·172	4·58	0·007	0·018	0·011
30/4600	2867	14·620	7·68	0·009	0·020	0·011
31/501	2944	4·473	5·53	0·011	0·021	0·010
31/508	2930	5·418	3·50	0·008	0·016	0·008
35/7148	2910	7·544	6·64	0·009	0·027	0·018
35/7156	2878	12·270	4·58	0·003	0·016	0·013
31/599	2885	11·070	5·32	0·004	0·011	0·007
31/572	2910	7·544	5·45	0·010	0·022	0·012
25/1518	2937	4·966	6·90	0·012	0·028	0·016
25/1521	2903	8·375	5·39	0·008	0·018	0·010
23/3736	2891	9·296	8·43	0·012	0·022	0·010
25/1527	2919	6·449	8·47	0·023	0·039	0·016
25/1528	2822	29·850	6·94	0·010	0·018	0·008
22/2309	2896	9·296	6·48	0·010	0·027	0·017

Discrepancy: Maximum	0·027% P.
Minimum	0·003% P.
Mean	0·0121% P.
Number of Samples	29
Standard Deviation	0·0052

results given in Table II. showed that the discrepancy between the calculated and the analysed figures for phosphorus in metal corresponding to slags containing more than 5% of phosphoric acid is of the same order as those containing less than that amount.

Method (C), *Based on Maurer and Bischof's Paper.*(14)

Figs. 1(*a*) and 1(*b*) are reproduced from Maurer and Bischof's paper. The phosphorus content of the steel corresponding to a given combination of phosphoric acid, lime and silica in the slag and of manganese in the steel is obtained by the combined use of these two diagrams. In this example (Cast 34/2238—refining period), the slag contained 51·7% of lime and 8·46% of silica by weight. To calculate the phosphorus content the procedure is

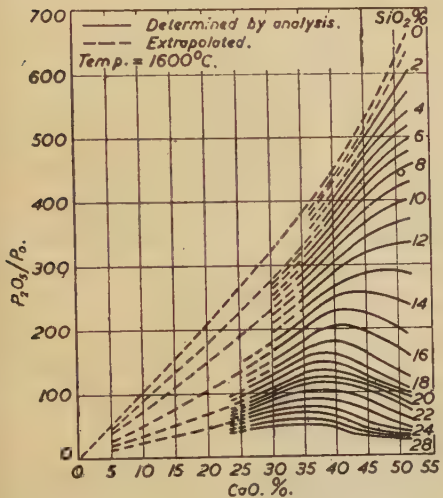


Fig. 1(*a*).—Distribution of Phosphorus between Steel and Slag in Basic Process (after manganese content of steel allowed for by means of Fig. 1(*b*)).

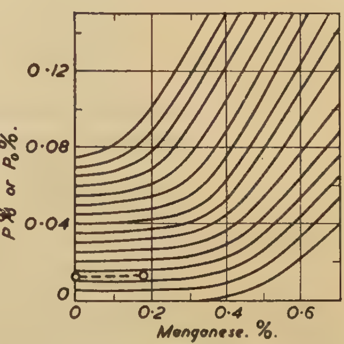


Fig. 1(*b*).—Diagram for Determining Theoretical Phosphorus Content, P_0, corresponding to Mn = 0% from phosphorus and manganese analyses of steel.

Fig. 1.—Maurer and Bischof's Diagrams.(11)

as follows: A point is plotted in Fig. 1(*a*) (as marked by a circle) to correspond with this composition; and the value for the ratio $(P_2O_5)/P_0$ is read on the vertical scale (P_0 represents the phosphorus content of the metal when there is no manganese present). This value will be found to be 445. As the slag also contained 5·04% of phosphoric acid, the value for P_0 is 5·04/445, *i.e.*, $P_0 = 0{\cdot}0114$. The steel contained 0·17% of manganese; so the ordinate 0·0114 is sought in Fig. 1(*b*) and followed along the curve as far as the point corresponding to 0·17% of manganese; this gives the phosphorus content of the metal as 0·0115%. In the present series of casts, the phosphorus content of the metal has been calculated by this method only in the case of those samples which were taken during the refining period. Since ferro-manganese was added to the bath shortly before tapping, it was considered that equilibrium between the slag and the metal would not be attained, as far as manganese was concerned,

before the tapping samples were taken; for this reason calculations on the latter were omitted.

TABLE III.—*Deviation in Phosphorus Content of Metal as Determined by Chemical Analysis and as Obtained by Interpolation on Maurer and Bischof's Diagrams.*

Cast No.	CaO. Wt.-%.	SiO_2. Wt.-%.	$\frac{P_2O_5}{P_0}$.	P_2O_5. Wt.-%.	P_0.	Mn. Wt.-%.	P in Metal. %. Calc.	P in Metal. %. Analysed.	P Discrepancy. %.
Refining samples:									
34/2238	51·7	8·46	445	5·04	0·0114	0·17	0·0115	0·020	0·0085
30/4600	51·24	10·2	395	8·47	0·0214	0·12	0·0215	0·025	0·0035
31/501	52·4	12·3	315	5·49	0·0174	0·17	0·018	0·018	nil
31/508	50·65	12·56	305	3·43	0·0112	0·17	0·011	0·012	0·001
35/7148	51·5	10·8	375	7·56	0·0202	0·19	0·021	0·024	0·003
35/7156	51·0	11·8	340	5·50	0·0162	0·20	0·017	0·030	0·013
31/599	49·6	9·56	402	5·68	0·0141	0·13	0·014	0·023	0·009
31/572	53·2	11·2	364	5·79	0·0159	0·16	0·017	0·024	0·007
30/4591	50·0	12·8	296	9·16	0·031	0·27	0·035	0·031	0·004
25/1518	51·3	8·44	440	7·10	0·0161	0·21	0·017	0·023	0·006
25/1521	51·7	10·04	400	5·41	0·0135	0·17	0·014	0·016	0·002
23/3736	54·7	9·64	425	8·56	0·0201	0·19	0·021	0·021	nil
25/1527	51·5	11·32	357	8·54	0·0239	0·26	0·026	0·029	0·003
25/1528	50·0	14·96	198	6·90	0·0348	0·21	0·037	0·018	0·019
22/2309	52·6	9·58	419	6·75	0·0161	0·21	0·017	0·022	0·005

Discrepancy: Maximum . . . . . 0·019% P.
Minimum . . . . nil
Mean . . . . . 0·0056% P.
Number of Samples 15
Standard Deviation . . . . . 0·005

Method (D), *Based on Schenck and Riess's Paper.*[1]

In describing Schenck and Riess's method it is necessary to explain fully the conventional signs used by them and the details of the procedure adopted in their calculations. These are as follows:

(1) Schenck's conventional signs:

Square brackets [ ] represent concentration (per cent.) of substance named contained in the metal.

Round brackets ( ) represent concentration (per cent.) of substance named contained in the slag.

Σ represents the sum total percentage of the substance named in brackets, however it may be combined.

Any chemical symbol in brackets which is not preceded by Σ represents the concentration (per cent.) of that substance existing exactly and only in the form shown. Thus (FeO) represents percentage free FeO contained in the slag. This portion is not combined with anything else and is actually in solution.

(FeO) must be distinguished from (ΣFeO). The latter represents the total percentage of FeO in the slag, however it may occur, free or combined.

(2) The total available lime in the slag, denoted by (ΣCaO)′, is obtained by deducting the amount of lime in combination with the phosphoric acid as $4CaO.P_2O_5$ from the total lime in the slag, *i.e.*, $(\Sigma CaO)' = (\Sigma CaO) - 1{\cdot}57(\Sigma P_2O_5)$.

(3) The equilibrium diagram of suitable temperature, (ΣFe) and (ΣMnO) is selected, and the values of (FeO) and (CaO) corresponding to the slag composition and temperature given are read off from it. Interpolations are made from one or more diagrams if necessary.

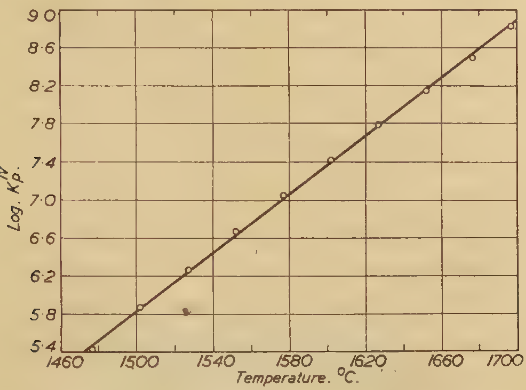


FIG. 2.—Relation between Log K_p^{IV} and Temperature. (From data calculated by Schenck.)

(4) The phosphorus in steel is then calculated from the equation :

$$\log [P] = \log K_p^{IV} + \log (\Sigma P_2O_5) - 5 \log (FeO) - 4 \log (CaO) - 0{\cdot}06(\Sigma P_2O_5).$$

The factors in the right-hand side of the equation are evaluated in the following manner :

Schenck gives the following values for $\log K_p^{IV}$ at various temperatures :[18]

Temperature. ° C.	1552.	1577.	1602.	1627.	1652.
$\log K_p^{IV}$ . . .	6·67	7·05	7·42	7·79	8·14

Since there is a linear relationship between $\log K_p^{IV}$ and temperature, use may be made of these values in plotting Fig. 2, which will enable an interpolation of $\log K_p^{IV}$ to be made at any intermediate temperature.

The slag in this example (Cast 34/2238—refining period) has the following analysis :

SiO_2.	Fe.	CaO.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.
8·46%	13·5%	51·7%	5·04%	5·74%	6·49%	12·8%	5·1%

The bath temperature at the time of sampling the slag was 1623° C.

Therefore, according to Schenck :

$$(\Sigma CaO) = 51{\cdot}7$$
$$(\Sigma P_2O_5) = 5{\cdot}04$$
and $$(\Sigma P_2O_5) \times 1{\cdot}57 = 7{\cdot}91$$
i.e., $$(\Sigma CaO)' = 51{\cdot}7 - 7{\cdot}91 = 43{\cdot}8$$
Also $$(\Sigma SiO_2) = 8{\cdot}46$$
$$T = 1623° \text{ C.}$$
$$(\Sigma Fe) = 13{\cdot}5$$
and $$(\Sigma MnO) = 5{\cdot}74$$

It now remains to calculate the two factors (FeO) and (CaO) from the above data, and this is done by means of Schenck's equilibrium diagrams.[18]

A specimen diagram is shown in Fig. 3, in which curves are drawn for (FeO) and (CaO) with (ΣSiO_2) as ordinate and (ΣCaO)′ as abscissa. This

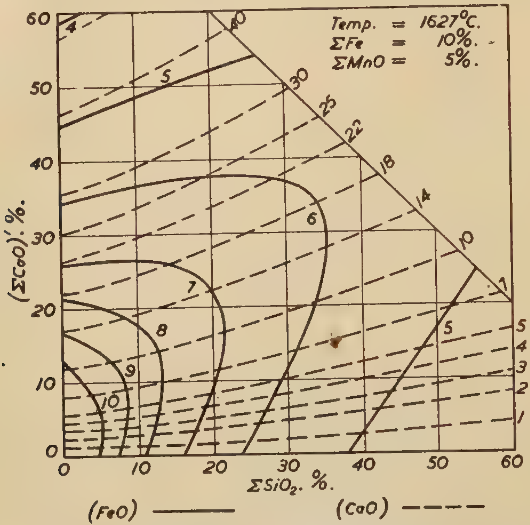


FIG. 3.—One of a Set of Sixty of Schenck's Equilibrium Diagrams, showing only the free lime and the free iron oxide curves.

diagram has been drawn up for the following conditions: $T = 1627°$ C., (ΣFe) = 10 and (ΣMnO) = 5. For each of the temperatures chosen by Schenck, two sets of diagrams have been compiled, the first with (ΣMnO) constant and (ΣFe) varying, the second with (ΣFe) constant and (ΣMnO) varying. In examining these diagrams it was found that (CaO) and (FeO) are controlled by the variables temperature, (ΣFe) and (ΣMnO) in descending order of magnitude. Thus in evaluating (CaO) and (FeO) under consideration, corrections for the above variables are made in the reverse

TABLE IV.—*(CaO) and (FeO) from Schenck's Equilibrium Diagrams.*

Diagram.	(CaO).	(FeO).	Diagram.	(CaO).	(FeO).
(1) $T = 1627°$ C. (ΣFe) = 10 (ΣMnO) = 5	34·8	5·36	(5) $T = 1577°$ C. (ΣFe) = 10 (ΣMnO) = 5	33·5	4·09
(2) $T = 1627°$ C. (ΣFe) = 10 (ΣMnO) = 10	35·4	5·35	(6) $T = 1577°$ C. (ΣFe) = 10 (ΣMnO) = 10	33·9	4·08
(3) $T = 1627°$ C. (ΣFe) = 15 (ΣMnO) = 5	31·0	6·56	(7) $T = 1577°$ C. (ΣFe) = 15 (ΣMnO) = 5	29·0	5·12
(4) $T = 1627°$ C. (ΣFe) = 15 (ΣMnO) = 10	31·4	6·54	(8) $T = 1577°$ C. (ΣFe) = 15 (ΣMnO) = 10	30·0	5·00

order. The first step in obtaining these values is to choose the four following pairs of diagrams of suitable composition and temperature, and to obtain the free concentrations of lime and iron oxide corresponding to $(\Sigma CaO)' = 43{\cdot}8$ and $(\Sigma SiO_2) = 8{\cdot}46$.

(ΣMnO) for the slag under consideration is 5·74. Values of (ΣMnO) for Diagrams (1) and (2) in Table IV. are 5 and 10 respectively. By simple proportion, (CaO) and (FeO) are determined for the (ΣMnO) of 5·74 with $(\Sigma Fe) = 10$ and $T = 1627°$ C. The result is shown in (I.) Table V. By similar use of diagrams (3) and (4) in Table IV., (CaO) and (FeO) are obtained for the (ΣMnO) of 5·74 with $(\Sigma Fe) = 15$ and $T = 1627°$ C. These values are shown in (II.) Table V. The same procedure is carried out for a temperature of 1577° C., and the values of (CaO) and (FeO) are given in (III.) and (IV.) Table V.

TABLE V.—*(CaO) and (FeO) by Simple Proportion.*

Temperature and Slag Composition.	(CaO).	(FeO).	Temperature and Slag Composition.	(CaO).	(FeO).
(I.) $T = 1627°$ C. $(\Sigma Fe) = 10$ $(\Sigma MnO) = 5{\cdot}74$	34·89	5·358	(III.) $T = 1577°$ C. $(\Sigma Fe) = 10$ $(\Sigma MnO) = 5{\cdot}74$	33·56	4·088
(II.) $T = 1627°$ C. $(\Sigma Fe) = 15$ $(\Sigma MnO) = 5{\cdot}74$	31·06	6·557	(IV.) $T = 1577°$ C. $(\Sigma Fe) = 15$ $(\Sigma MnO) = 5{\cdot}74$	29·14	5·10

In a like manner, by using (I.) and (II.) in Table V., (CaO) and (FeO) are obtained for a (ΣFe) of 13·5. This is repeated for the lower temperature by means of (III.) and (IV.) in Table V.; and (CaO) and (FeO) are thus obtained for the given slag composition at two temperatures, 1627° and 1577° C. The results are shown below:

$$\left.\begin{aligned} T &= 1627°\text{ C.} \\ (\Sigma Fe) &= 13{\cdot}5 \\ (\Sigma MnO) &= 5{\cdot}74 \\ (\Sigma CaO)' &= 43{\cdot}8 \\ (\Sigma SiO_2) &= 8{\cdot}46 \end{aligned}\right\} \begin{aligned} (CaO) &= 32{\cdot}23 \\ (FeO) &= 6{\cdot}20 \end{aligned} \qquad \left.\begin{aligned} T &= 1577°\text{ C.} \\ (\Sigma Fe) &= 13{\cdot}5 \\ (\Sigma MnO) &= 5{\cdot}74 \\ (\Sigma CaO)' &= 43{\cdot}8 \\ (\Sigma SiO_2) &= 8{\cdot}46 \end{aligned}\right\} \begin{aligned} (CaO) &= 30{\cdot}47 \\ (FeO) &= 4{\cdot}80 \end{aligned}$$

The third variable, namely the temperature, can now be dealt with, and the values of (CaO) and (FeO) obtained by simple proportion for a temperature of 1623° C. Thus, $(CaO) = 32{\cdot}10$ and $(FeO) = 6{\cdot}07$. In this way the values of (CaO) and (FeO) for the slag under consideration are obtained.

$$\begin{aligned} \text{Now,}\quad (\Sigma P_2O_5) &= 5{\cdot}04 & \therefore\ \log(\Sigma P_2O_5) &= 0{\cdot}7024, \\ (FeO) &= 6{\cdot}07 & \therefore\ 5\log(FeO) &= 3{\cdot}9160, \\ (CaO) &= 32{\cdot}10 & \therefore\ 4\log(CaO) &= 6{\cdot}0260 \\ \text{and}\quad 0{\cdot}06\,(\Sigma P_2O_5) &= 0{\cdot}3024. \end{aligned}$$

Also $\log K_p^{IV}$ at 1623° C. = 7·73, which is obtained from Fig. 2.

Schenck and Riess's equation may now be applied:

$$\begin{aligned} \log[P] &= \log K_p^{IV} + \log(\Sigma P_2O_5) - 5\log(FeO) - 4\log(CaO) - 0{\cdot}06\,(\Sigma P_2O_5) \\ \therefore\ \log[P] &= 7{\cdot}73 + 0{\cdot}7024 - 3{\cdot}9160 - 6{\cdot}0260 - 0{\cdot}3024 \\ &= 8{\cdot}4324 - 10{\cdot}2444 \\ &= \bar{2}{\cdot}1880 \\ &= \log 0{\cdot}0155 \\ \therefore\ [P] &= 0{\cdot}0155. \end{aligned}$$

TABLE VI.—*Deviation in Phosphorus Content of Metal as Determined by Chemical Analysis and as Calculated from Schenck and Riess's Work.*

Cast No.	Temp. °C.	log K_p^{IV}.	(FeO).	(CaO).	P in Metal. %. Calc.	P in Metal. %. Analysed.	P Discrepancy. %.
Sampled during refining:							
34/2238	1623	7·73	6·07	32·10	0·0155	0·020	0·0045
30/4600	1550	6·61	4·58	24·45	0·015	0·025	0·010
31/501	1613	7·58	5·50	31·26	0·0205	0·018	0·0025
31/508	1619	7·66	5·65	32·50	0·015	0·012	0·003
35/7148	1586	7·16	5·18	27·20	0·019	0·024	0·005
35/7156	1570	6·92	5·20	26·55	0·012	0·030	0·018
31/599	1574	6·98	5·27	26·65	0·012	0·023	0·011
31/572	1607	7·48	5·10	32·72	0·020	0·024	0·004
30/4591	1602	7·42	5·90	24·19	0·028	0·031	0·003
25/1518	1620	7·68	5·91	30·42	0·021	0·023	0·002
25/1521	1601	7·39	5·205	31·30	0·017	0·016	0·001
23/3736	1598	7·34	5·05	30·50	0·020	0·021	0·001
25/1527	1610	7·53	5·19	28·43	0·036	0·029	0·007
25/1528	1556	6·70	4·125	26·10	0·024	0·018	0·006
22/2309	1597	7·33	5·26	30·65	0·016	0·022	0·006
Sampled on tapping:							
34/2238	1620	7·68	6·40	29·91	0·014	0·018	0·004
30/4600	1575	7·00	4·91	26·80	0·018	0·020	0·002
31/501	1618	7·65	5·62	31·87	0·020	0·021	0·001
31/508	1610	7·53	5·507	31·77	0·015	0·016	0·001
35/7148	1599	7·36	5·62	27·51	0·019	0·027	0·008
35/7156	1581	7·08	5·44	28·21	0·010	0·016	0·006
31/599	1585	7·15	6·22	23·93	0·012	0·011	0·001
31/572	1599	7·36	4·93	32·65	0·018	0·022	0·004
25/1518	1614	7·59	5·71	30·39	0·020	0·028	0·008
25/1521	1595	7·30	4·98	31·31	0·017	0·018	0·001
23/3736	1592	7·25	4·80	30·81	0·020	0·022	0·002
25/1527	1604	7·44	5·03	27·71	0·038	0·039	0·001
25/1528	1550	6·61	3·91	26·17	0·025	0·018	0·007
22/2309	1591	7·24	4·86	31·10	0·018	0·027	0·009

Discrepancy: Maximum . . . . . 0·018% P.
Minimum . . . . . 0·001% P.
Mean . . . . . 0·0048% P.
Number of Samples 29
Standard Deviation . . . . . 0·0038

In Table VII., the phosphorus results calculated on the bases of the work of Whiteley, Herty, Maurer and Bischof, and Schenck and Riess are compared.

TABLE VII.—*Comparison of Calculated Phosphorus Results.*

Basis of Work.	Discrepancy between Calculated and Analysed Phosphorus in Metal. % P. Maximum.	Minimum.	Average.	Standard Deviation.	No. of Samples Involved.
Whiteley [4]	0·020	0·003	$0{\cdot}010_3$	$0{\cdot}004_7$	29
Herty [11]	0·027	0·003	$0{\cdot}012_1$	$0{\cdot}005_2$	29
Maurer and Bischof [14]	0·019	nil	$0{\cdot}005_8$	0·005	15
Schenck and Riess [1]	0·018	0·001	$0{\cdot}004_8$	$0{\cdot}003_8$	29

It will be seen that the results worked out on the basis of Schenck and Riess's work give the least average discrepancy between the calculated and the analysed phosphorus contents. The standard deviation of the discrepancies is also the smallest by this method of calculation; which means that more consistent results are obtained for a given degree of accuracy.

It can therefore be concluded that the use of Schenck and Riess's work as a basis for estimating the phosphorus content of metal from the composition of the slag and the bath temperature for the casts in the present investigation, gives results which approach nearer to the analysed figures than any of the other methods.

Correlation of Bath Temperature and Slag Analysis with Furnace and Ladle Rephosphorisation.

From the analytical data given in Appendix A, it was noted that the reversion of phosphorus from slag to metal in the ladle was more pronounced on the higher-carbon casts (Rotherham plant, 0·42–0·47 and 0·58–0·62% carbon steel specifications) than on the lower-carbon casts (Templeborough plant, 0·24–0·28 and 0·26–0·30% carbon steel specifications). It was also noted that the apparent rate of phosphorus reversion varied from cast to cast, and, further, that it varied at different periods during the casting process. In order, therefore, to enable a true comparison to be made of the relative degree of rephosphorisation on the observed casts, it was necessary to express in a graphic form the phosphorus contents of the successive ladle-metal samples in each cast. It was also found necessary to treat the Rotherham and the Templeborough casts separately, owing to the difference in the carbon contents of the steels. The diagrams described below are based on the data obtained from the first series of high-carbon casts (Rotherham plant).

Description of the Diagrams.

The weight of metal in the ladle, calculated from the actual weight of ingots cast at the time when a ladle sample was taken, was plotted against the analysed phosphorus content of the corresponding metal sample (Fig. 4(*a*)). The number in brackets shown beside a plotted point corresponds to the identification number of the cast to which the point relates. These curves may be taken to represent the apparent rates of phosphorus pick-up in the metal from the beginning to the end of the casting process. The phosphorus content of the tapping samples, which were taken in the furnace as the metal started to run out from the tap-hole, was used to represent the phosphorus content of the metal in the ladle at the commencement of casting. This is not a true representation, since there were in each case 10·5–13·5 cwt. of ferro-manganese and ferro-silicon added to the metal as ladle finishings, which would introduce something like 0·002% of phosphorus into the metal. It may be further complicated by the amount rephosphorised, if any, whilst the metal was held in the ladle before commencing to cast.

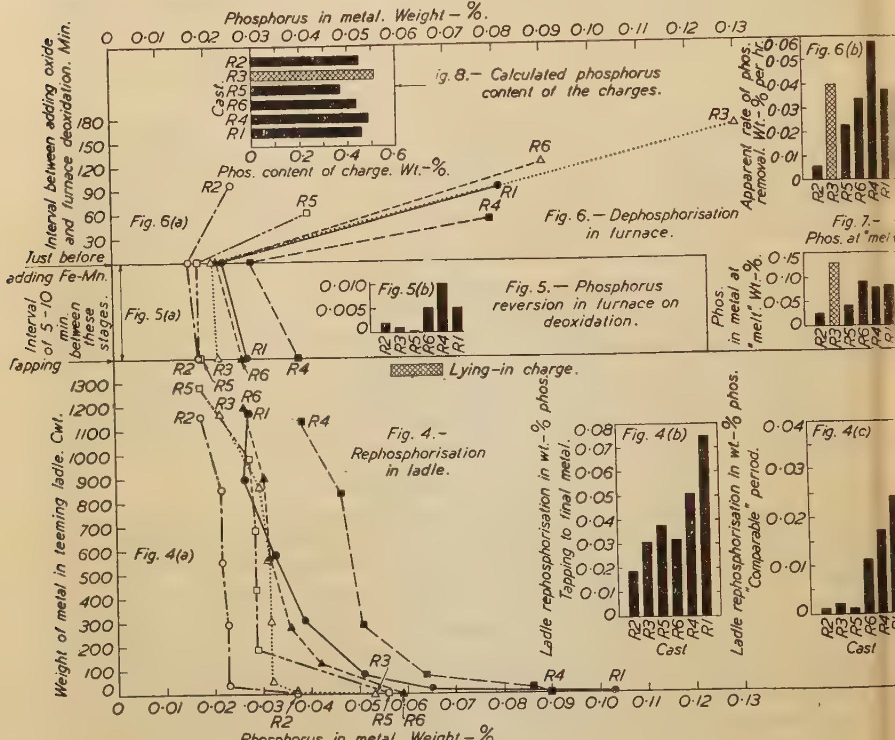
Fig. 4(*b*) is a " block diagram " showing the relative amounts of phosphorus pick-up in the metal between tapping and the end of casting, for the various casts.

Fig. 4(*c*) shows the phosphorus pick-up from the end of teeming the first plate of ingots, that is when the weight of metal in the ladle had fallen to 45 tons to the moment when it contained 5 tons. In effect this represents phosphorus pick-up during the casting of the last 40 tons of metal from the ladle, excluding the 5 tons mentioned.

In studying the rephosphorisation in the ladle along with other factors,

Fig. 4(c) was taken as the basis, as it eliminates any possible error due to a variation in the small amount of metal that may be trapped in the bottom of the ladle. It also eliminates the possible error resulting from any initial ladle rephosphorisation as discussed above.

Fig. 5(a) shows the apparent rate of phosphorus reversion in the furnace after an addition of 5 cwt. of ferro-manganese to the bath. The phosphorus increase due to the amount of phosphorus contained in the ferro-manganese added would only amount to 0·001%. The interval between taking the samples during refining and at tapping varied only from 5 to



FIGS. 4 to 8.—Phosphorus Removal and Reversion Data.

10 min. on these casts; it may therefore be assumed that the time variable, in this instance, would not affect the results.

The block diagram, Fig. 5(b), shows the relative degree of rephosphorisation in the furnace on deoxidation with ferro-manganese, for the various casts.

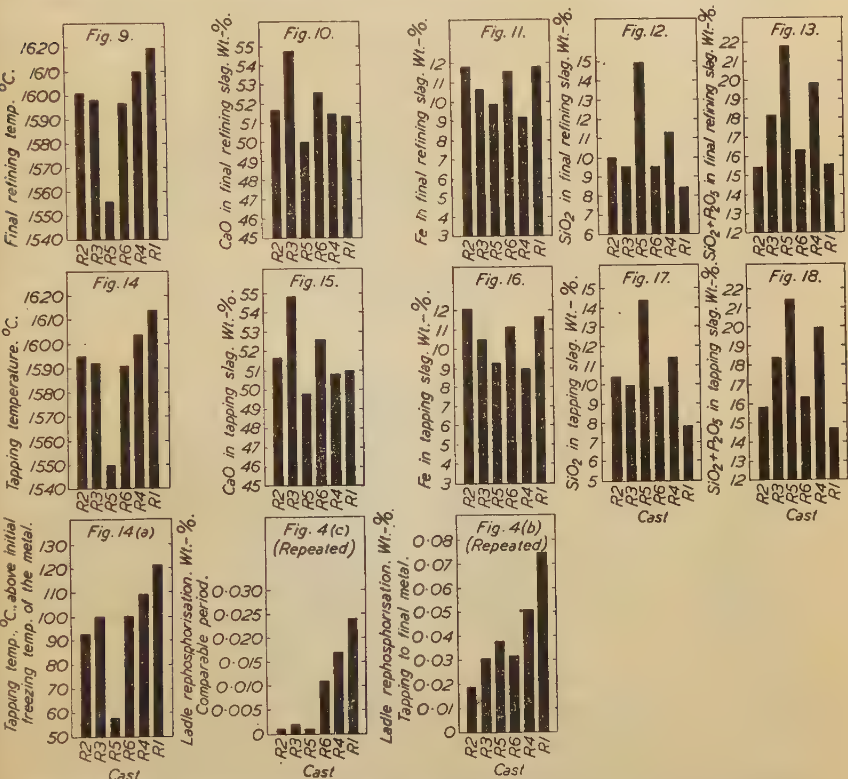
Fig. 6(a) shows the drop in phosphorus content of the bath from the beginning to the end of the refining period. Since the total refining time for each of these casts varied from 50 to 165 min., it was found necessary to take into account the actual refining interval; this has been introduced as the ordinate in Fig. 6(a). In this way the slope of each curve may be taken to represent the apparent rate of phosphorus removal. This is graphically shown by the block diagram, Fig. 6(b).

The block diagrams, Figs. 7 and 8, show the phosphorus content of the metal in the bath when melted, and that contained in the charged material, respectively.

The block diagrams, Figs. 9 to 18, show the relative immersion temperatures of the baths, the relative amounts of the total lime, the total iron, the total silica, and the sum of the total acid contents in the furnace slags. The sum of the percentages of silica and phosphoric acid is taken as representing the total acid content.

Observational Results.

(1) A similarity between Figs. 4(*c*) and 5(*b*) is at once evident, indicating a tendency for rephosphorisation in the furnace on deoxidation to be associated with pronounced rephosphorisation in the ladle.



FIGS. 9 to 18.—Bath Temperature and Slag Composition of Six Casts of Similar Steel. Figs. 4(*b*) and 4(*c*).—Extent of ladle rephosphorisation in these casts.

(2) On comparing Figs. 6(*b*) and 7 with Figs. 5(*b*) or 4(*c*), it will be seen that the rephosphorised casts had a higher rate of phosphorus removal during the process of refining, and that the phosphorus content of the bath when melted was also higher in the case of these casts. There was a slight variation in the phosphorus content of the charge, but it does not appear to be responsible for the high melt-out phosphorus of these casts.

(3) Fig. 4(*c*), taken in conjunction with Figs. 14 and 14(*a*), shows that when superheat, *i.e.*, temperature above the calculated freezing point,[19] is considered, there is a definite indication that high superheat increases the tendency to rephosphorisation. Fig. 14 shows this only to a limited extent, presumably owing to the fact that it does not take into consideration the carbon content of the metal.

(4) Figs. 4(*c*) and 15, taken in conjunction with Fig. 14 and 14(*a*), suggest that when a high temperature is associated with a high lime content in the slag, the effect of the latter counteracts the rephosphorisation tendency of the former.

(5) A comparison of Fig. 4(*c*) and the remaining block diagrams (Figs. 16–18), representing the relative amounts of the total iron, the total silica and the total acid contents in the slags respectively, shows that none of these factors was directly associated with rephosphorisation of the metal.

Possible Conclusions.

(I.) As far as the casts examined can be taken as typical, the effect of superheat on rephosphorisation appears to be predominant, and it therefore follows that when working with one type of steel in a given carbon range the actual bath temperature may be substituted for the superheat.

(II.) The mechanism of ladle rephosphorisation may be on similar lines to that of phosphorus reversion in the furnace during deoxidation with ferro-manganese; but the rephosphorisation in the furnace represents such a small proportion of that which occurs in the ladle that it may be unwise to attach much importance to this conclusion.

(III.) The phenomenon of rephosphorisation is possibly associated with such slag conditions as would result in a high melt-out phosphorus of the bath for a given charge.

Application of Schenck's Equilibrium Diagrams to the Study of Ladle Rephosphorisation.

Knowing the ultimate analysis of a slag and its corresponding bath temperature, it is possible to read from Schenck's equilibrium diagrams, either directly or by interpolation when necessary, the free concentrations of lime and iron oxide in that slag. Details of the rather complicated method for obtaining accurate values of the free concentrations and the conventional signs adopted for use with the equilibrium diagrams have already been fully described (*see* p. 8).

The free lime and the free iron oxide concentrations of the final refining and the tapping slags for the six high-carbon casts are shown, alongside the corresponding amount of rephosphorisation observed in each cast, in Table VIII.

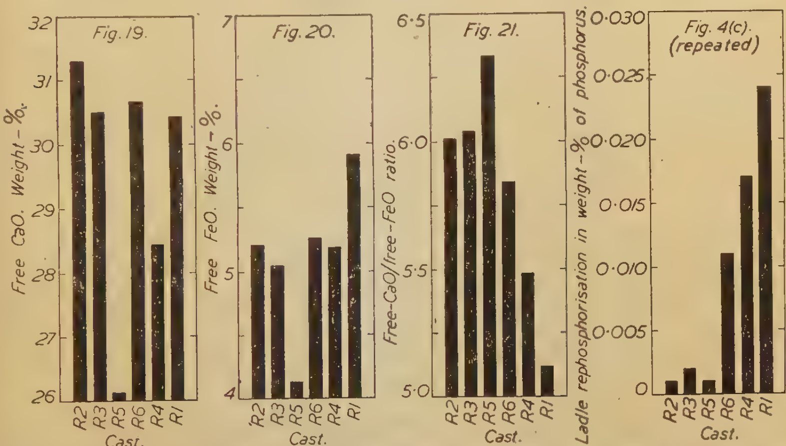
Owing to the slag foaming out through the furnace doors at the tapping of these high-carbon casts, possibly due in part to the physical reaction caused by the finishing additions of ferro-manganese which were at a very much lower temperature than the bath, and in part to the reduction of the gas supply to the furnace at tapping, it was not possible to take immersion temperatures at that stage. Temperatures calculated on the basis of the final refining temperature, with a correction of 1·3° C.[20] temperature drop per cwt. of ferro-manganese added to the bath, were used in conjunction with the ultimate slag analysis in obtaining the free concentrations of lime and iron oxide for the tapping slags. In view of the lack of actual tapping temperatures for these casts, it is proposed that figures corresponding to samples taken immediately before furnace deoxidation be used in the following work. Incidentally, it may be an advantage to base the study on the final refining figures; since after the addition of ferro-manganese to

TABLE VIII.—*Correlation of (CaO) and (FeO) Values with Rephosphorisation.*

Cast Identification No.	Tapping Slag.			Final Refining Slag.			Amount Rephosphorised (1 point = 0·001% P).	
	(CaO).	(FeO).	Ratio $\frac{(CaO)}{(FeO)}$.	(CaO).	(FeO).	Ratio $\frac{(CaO)}{(FeO)}$.	During Casting Last 40 tons of Metal from Ladle, Excluding the Final 5 tons.	From Immediately Before Furnace Deoxidation to End of Casting.
R.1 .	30·39	5·71	5·32	30·42	5·91	5·13	24 points	80 points
R.2 .	31·31	4·98	6·29	31·30	$5{\cdot}20_5$	6·01	1 ,,	21 ,,
R.3 .	30·81	4·80	6·42	30·50	5·05	6·04	2 ,,	32 ,,
R.4 .	27·71	5·03	5·51	28·43	5·19	5·48	17 ,,	61 ,,
R.5 .	26·17	3·91	6·69	26·10	$4{\cdot}12_5$	6·33	1 ,,	38 ,,
R.6 .	31·10	4·86	6·40	30·65	5·26	5·83	11 ,,	32 ,,

the bath, which on these casts was within 10 min. of tapping, further control of the process is impracticable owing to the excessive loss of manganese from the metal.

The values of the free lime and the free iron oxide concentrations given in Table VIII. are shown graphically in Figs. 19 and 20, respectively.



FIGS. 19, 20 and 21.—Relative Amounts of Free Lime, Free Iron Oxide and their Ratio for slags belonging to six casts of 0·58–0·62% carbon steel. *Cf.* Fig. 4(*c*), corresponding ladle rephosphorisation on each cast.

On comparing these block diagrams with Fig. 4(*c*), which represents graphically the respective amounts of ladle rephosphorisation which took place in these casts, it was found that there was no correlation between the free lime or the free iron oxide concentrations and the rephosphorisation in the ladle.

Fig. 21 was then drawn, based on the values of the ratio (CaO)/(FeO). When this diagram is compared with Fig. 4(*c*), it is at once evident that there is a definite correlation between the ratio (CaO)/(FeO) and pronounced rephosphorisation in the ladle.

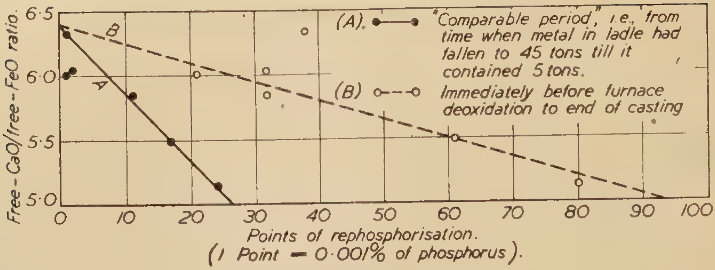


FIG. 22.—Correlation between Free-CaO/Free-FeO Ratio and Extent of Rephosphorisation at various stages during casting.

In Fig. 22 the value of the ratio (CaO)/(FeO) is plotted against the actual amount of rephosphorisation which occurred during the following periods :

(*A*) "Comparable casting period," *i.e.*, the interval of the casting period during which the metal in the ladle decreased from 45 to 5 tons.

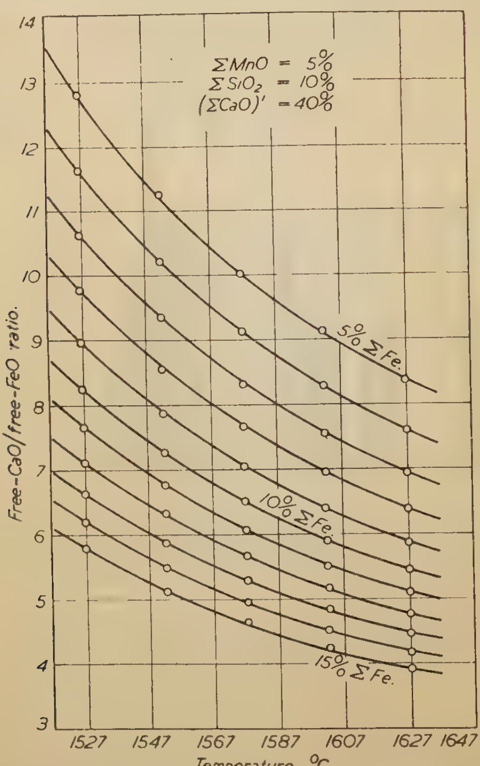


FIG. 23.—Correlation between Free-CaO/Free-FeO Ratio and Temperature at various (ΣFe) contents.

For simplicity the term " comparable casting period " will be used in future to designate precisely this particular casting period, as defined above.

(*B*) Immediately before furnace deoxidation to the end of casting.

It will be seen that a straight-line relationship may be drawn for each set of points thus plotted. On producing the lines to the left, it appears that, with a ratio of free lime to free iron oxide concentration corresponding to a value of not less than 6·4, no ladle rephosphorisation should take place; or at least no appreciable amount under the conditions obtaining in the manufacture of the casts under observation.

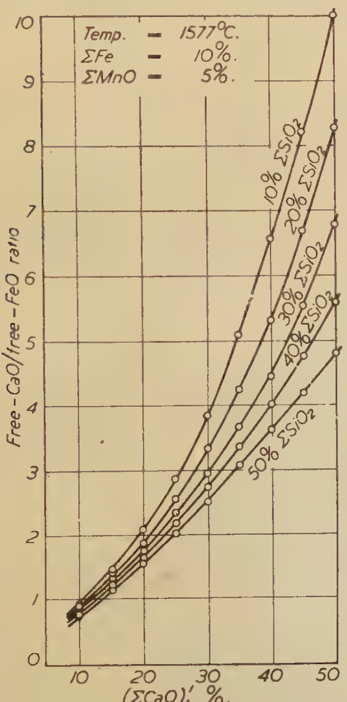


Fig. 24.—Correlation between Free-CaO/Free-FeO Ratio and Total Available CaO at various (ΣSiO_2) contents.

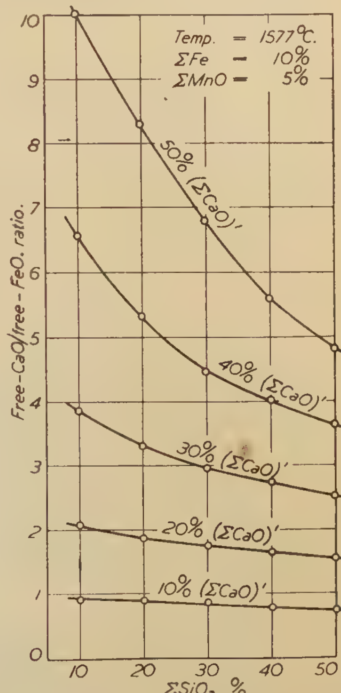


Fig. 25.—Correlation between Free-CaO/Free-FeO Ratio and Total Silica at Various $(\Sigma CaO)'$ Contents.

It is impossible to study the effect of one constituent alone in a series of experiments where all the factors are interdependent; thus an alteration in silica content will cause a change in the concentration of all other constituents in the slag. Therefore it would seem that no single slag constituent can be cited which would have general application in indicating or measuring the tendency to rephosphorisation.

This may be further illustrated by considering Figs. 23–27, which were prepared from values directly interpolated from Schenck's equilibrium diagrams. These graphs show how the temperature and each of the main slag constituents affect the ratio (CaO)/(FeO). The values of 10 and 40

for the total silica and the total available lime concentrations, respectively, are chosen as bases for these graphs simply because they represent roughly the actual figures for the slags dealt with in the present casts. The values $(\Sigma Fe) = 10$ and $(\Sigma MnO) = 5$ are chosen for the same reason. Incidently, at 1577° C. the ratio $(CaO)/(FeO)$ for a slag having $(\Sigma SiO_2) = 10$, $(\Sigma CaO)' = 40$, $(\Sigma Fe) = 10$ and $(\Sigma MnO) = 5$ is approximately 6·4, which corresponds to the estimated value of this ratio for no rephosphorisation, as indicated by Fig. 22.

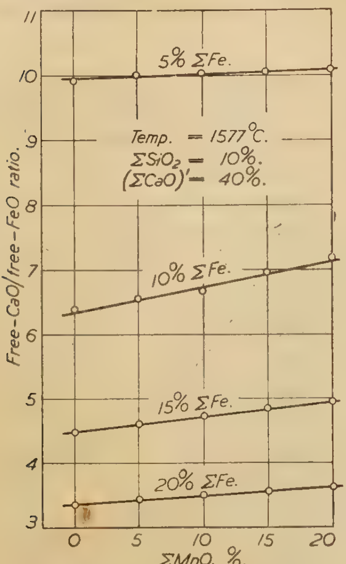


FIG. 26.—Correlation between Free-CaO/Free-FeO Ratio and (ΣMnO) at various (ΣFe) contents.

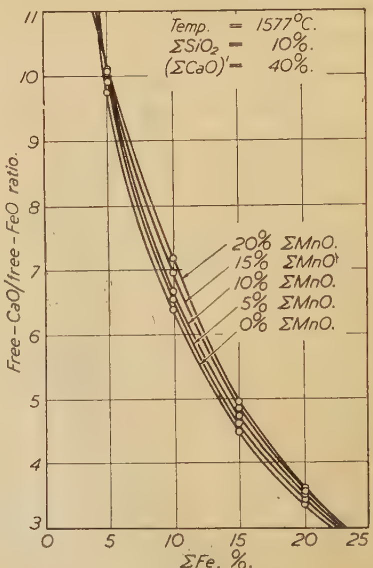


FIG. 27.—Correlation between Free-CaO/Free-FeO Ratio and (ΣFe) at various (ΣMnO) contents.

From the foregoing it will be seen how the ratio $(CaO)/(FeO)$ is governed by each of the main constituents of the slag and by the temperature. It may be said, therefore, that the effect of each constituent and of the temperature is integrated by the ratio of free lime to free iron oxide, and therefore, if one accepts the results given in Fig. 22, a straight-line relationship may be said to have been established between the ratio $(CaO)/(FeO)$ and the extent of rephosphorisation in the ladle.

A summary of the results obtained from a study of the curves shown in Figs. 23–27 is given in Table IX.

The factors influencing the ratio $(CaO)/(FeO)$ given in this table are largely interdependent. Decreasing temperature will affect the total iron content of the slag; increased $(\Sigma CaO)'$ will, by dilution, reduce the total silica and also directly increase the total iron content by the formation and dissociation of calcium ferrites; increased total manganous oxide content depends upon the manganese content of the charge; with increased total manganous oxide content of the slag, the free iron oxide will tend to be lower and the lime-carrying power of the slag will also be affected. Hence the two primary factors, given equal manganese contents of the

TABLE IX.—*Summary of Figs.* 23–27.

Figure No.	Requirements to Increase the Ratio (CaO)/(FeO).	Remarks.
23	Decreasing temperature.	Influence slightly more pronounced at lower (ΣFe) than at higher (ΣFe).
24	Increasing total available lime.	Influence more pronounced at lower (ΣSiO_2) than at higher (ΣSiO_2). Influence also increased with increasing (ΣCaO)′.
25	Decreasing total silica.	Influence increased with increasing (ΣCaO)′ and increased slightly with decreasing (ΣSiO_2).
26	Increasing total manganous oxide.	A straight-line relationship. Influence most pronounced at about 10% total iron. Above or below this value the influence gradually became negligible.
27	Decreasing total iron.	Influence more pronounced at lower (ΣFe) than at higher (ΣFe). Influence slightly more pronounced when (ΣMnO) is high.

charges, are temperature and lime. In order of importance, therefore, the factors appear to be :

(1) Temperature,
(2) Total available lime,
(3) Total iron,
(4) Total silica,
(5) Total manganous oxide.

In open-hearth practice increased total lime and total iron (at equilibrium) will be interrelated, and hence any deleterious influence of increased iron will be offset by the beneficial influence of the increased total lime. It must also be mentioned that in actual practice there are limitations to the governing factors listed in Table IX.; for example, the lower limit of the temperature at tapping will be fixed by the minimum safe temperature for casting the steel, and the upper limit of the total available lime or the total lime for a given phosphoric acid content in the slag, will be governed by the slag becoming inactive in the bath. A lower limit will also have to be imposed on the total iron content of the slag, which will be dependent upon the phosphorus content of the charge and that desired in the final steel.

In Figs. 23–27 are given the relationships between the ratio (CaO)/(FeO) and each of the main slag constituents; thus for the given conditions it is possible to read off from each curve the change in the factor, such as temperature in Fig. 23 or (ΣCaO)′ in Fig. 24, &c., which corresponds to a drop in the ratio (CaO)/(FeO) from 6·5 to 5·5. From Fig. 22, curve *A*, this drop in the ratio corresponds to 19 points of rephosphorisation. It is thus possible to obtain a measure of the necessary change in the value of each factor, *i.e.*, temperature, (ΣCaO)′, &c., which would give 19 points of rephosphorisation; and then by simple proportion one can obtain a figure for the degree of rephosphorisation for a unit change of each of the factors temperature, (ΣCaO)′, (ΣSiO_2), (ΣMnO) and (ΣFe). This has been done in Table X.

It must be emphasised that the figures shown in Table X. will not have general application, but are limited to similar steelmaking conditions to those relating to the casts under consideration.

In Table XI. are listed the actual temperatures and the concentrations of the four factors that influence the ratio (CaO)/(FeO) and the deviation of these values from the standard conditions for the casts studied.

TABLE X.—*Comparison of Factors Influencing Rephosphorisation.*

Basis.	Effect Required from Each of the Influencing Factors Alone to Lower the Ratio (CaO)/(FeO) by 1 Unit on the Graphs, Corresponding to 19 Points of Ladle Rephosphorisation during the "Comparable Casting Period."
Fig. 23	46° C. increase in temperature, *i.e.*, 10° C. temperature increase gives 4·14 points rephosphorisation.
Fig. 24	3½% decrease in $(\Sigma CaO)'$, *i.e.*, 1% $(\Sigma CaO)'$ decrease gives 5·41 points rephosphorisation.
Fig. 25	7·8% increase in (ΣSiO_2), *i.e.*, 1% (ΣSiO_2) increase gives 2·44 points rephosphorisation.
Fig. 26	25·0% decrease in (ΣMnO), *i.e.*, 1% (ΣMnO) decrease gives 0·76 points rephosphorisation.
Fig. 27	2·5% increase in (ΣFe), *i.e.*, 1% (ΣFe) increase gives 7·60 points rephosphorisation.

TABLE XI.—*Recorded Temperatures and Slag Analyses and Their Deviation from Standard Conditions.*

Cast Index No.	Temperature. °C.		$(\Sigma CaO)'$. %.		(ΣSiO_2). %.		(ΣMnO). %.		(ΣFe). %.	
	Recorded.	Deviation from 1577.	Present.	Deviation from 40.	Present.	Deviation from 10.	Present.	Deviation from 5.	Present.	Deviation from 10.
R.1	1620	+43	40·1	+0·1	8·44	−1·56	6·71	+1·71	11·8	+1·8
R.2	1601	+24	43·2	+3·2	10·04	+0·04	6·66	+1·66	11·8	+1·8
R.3	1596	+21	41·3	+1·3	9·64	−0·36	5·53	+0·53	10·7	+0·7
R.4	1610	+33	38·1	−1·9	11·32	+1·32	5·98	+0·98	9·2	−0·8
R.5	1556	−21	39·2	−0·8	14·96	+4·96	6·20	+1·20	9·9	−0·1
R.6	1597	+20	42·0	+2·0	9·58	−0·42	6·10	+1·10	11·6	+1·6

The deviations of each of the controlling factors shown in Table XI. may be converted into rephosphorisation values by means of Table X. This is done in Tables XII. and XII*a*., from which it can be seen which slag would be expected to give high rephosphorisation.

The results shown in Tables XII. and XII*a*. may be considered reasonably good, particularly those relating to the "comparable casting period." Incidentally this period corresponds approximately to the interval between taking the first and the last pit samples in general practice at the

TABLE XII.—*Calculated Extent of Rephosphorisation for the "Comparable Casting Period."*

1 Unit Drop in the Ratio (CaO)/(FeO), Interpolated from Fig. 22, Curve *A*, Results in 19 Points of Rephosphorisation for the "Comparable Casting Period."

Influence:	Temperature.	$(\Sigma CaO)'$.	(ΣSiO_2).	(ΣMnO).	(ΣFe).	Points of Rephosphorisation.	
Factor of Influence:	0·414.	5·41.	2·44.	0·76.	7·60.	Calc.	Actual.
Cast Identification No.						1 point ≡ 0·001% P.	
R.1 .	+17·80	− 0·54	− 3·81	−1·30	+13·68	25·83	24
R.2 .	+ 9·94	−17·31	+ 0·10	−1·26	+13·68	5·12	1
R.3 .	+ 8·69	− 7·08	− 0·88	−0·40	+ 5·32	5·70	2
R.4 .	+13·66	+10·28	+ 3·22	−0·75	− 6·08	20·33	17
R.5 .	− 8·69	+ 4·33	+12·10	−0·91	− 0·76	6·23	1
R.6 .	+ 8·28	−10·82	− 1·02	−0·64	+12·16	7·76	11

TABLE XII*a*.—*Calculated Extent of Rephosphorisation for the Period from Immediately before Furnace Deoxidation to the End of Casting.*

1 Unit Drop in the Ratio (CaO)/(FeO), Interpolated from Fig. 22, Curve *B*, Results in 67 Points of Rephosphorisation during Furnace Deoxidation and to the End of Casting.

Influence :	Temperature.	$(\Sigma CaO)'$.	(ΣSiO_2).	(ΣMnO).	(ΣFe).	Points of Rephosphorisation.	
Factor of Influence :	1·456.	19·143.	8·59.	2·68.	26·80.	Calc.	Actual.
Cast Identification No.						1 point ≡ 0·001% P.	
R.1 .	+62·60	$-1{\cdot}91_4$	−13·40	$-4{\cdot}58_3$	+48·24	90·94	80
R.2 .	+34·94	−61·26	+ 0·34	−4·45	+48·24	17·81	21
R.3 .	+30·57	−24·89	− 3·09	−1·42	+18·76	19·93	32
R.4 .	+48·05	+36·37	+11·33	−2·63	−21·44	71·68	61
R.5 .	−30·57	+15·31	+42·60	−3·22	− 2·41	21·71	38
R.6 .	+29·12	−38·28	− 3·60	−2·95	+42·88	27·17	32

Rotherham plant, where the casts studied were made. The maximum difference between the calculated and the analysed values of rephosphorisation on these casts is 0·005% of phosphorus (*see* Table XII., cast R.5). The maximum difference between the calculated and the analysed values of rephosphorisation corresponding to the total casting period, is somewhat greater. This is seen in Table XII*a*., cast R.5, the difference being 0·016% of phosphorus. It may be mentioned that the rate of phosphorus return from slag to metal during the final stage of casting was rapid on all casts observed. This feature, coupled with the difficulties encountered in obtaining a representative metal sample as soon as slag appeared in the teeming stream, possibly accounts for this increased deviation of the calculated values from the analytical values. Fig. 22, showing the extent of rephosphorisation against the ratio (CaO)/(FeO), also shows a greater scattering for the total casting period.

As the results given in Fig. 22 and Tables XII. and XII*a*. were based on a small number of casts, it was thought desirable to have a check on other casts of similar steel, and observe whether these results still applied. For this purpose, four more casts were sampled. The slag and metal analysis data are given in Appendix A and are identified by the prefix "R.C." before the cast number. "R.C." stands for Rotherham plant checking cast.

Fig. 28 shows the analysed phosphorus content of the metal at various

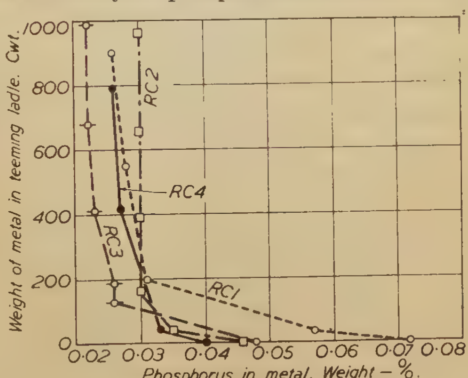


FIG. 28.—Phosphorus Pick-up in Ladle; 0·58–0·62% carbon R.C. casts.

stages of teeming the four new casts. The concentrations of free lime and free iron oxide in the slags were obtained from Schenck's equilibrium diagrams, as in the previous casts, and are given in Table XIII., together with the degrees of rephosphorisation actually found.

TABLE XIII.—*Checking Casts: (CaO), (FeO) and Degree of Rephosphorisation.*

Cast Identification Number.	(CaO).	(FeO).	Ratio $\frac{(CaO)}{(FeO)}$.	Amount Rephosphorised during "Comparable Casting Period." 1 point ≡ 0·001% P.
R.C.1 . .	28·07	5·64	4·98	31 points
R.C.2 . .	29·40	4·80	6·11	5 ,,
R.C.3 . .	29·31	4·58	6·40	4 ,,
R.C.4 . .	31·48	5·45	5·76	7 ,,

When the results given in Table XIII. were plotted on Fig. 22, a certain amount of scatter about curve *A* was observed, but the deviation was only small. The maximum horizontal deviation from the curve was less than 0·005% of phosphorus.

The calculated rephosphorisation based on the results given in Table X. is shown in Table XIV.; it will be seen that a close agreement is again obtained between the calculated rephosphorisation and that actually found.

TABLE XIV.—*Calculated and Actual Rephosphorisation Compared.*

Influence:	Temperature.	(ΣCaO)′.	(ΣSiO_2).	(ΣMnO).	(ΣFe).	Points of Rephosphorisation (1 point ≡ 0·001% P). Calc.	Actual.
Factor of Influence (from Table X.):	0·414.	5·41.	2·44.	0·76.	7·60.		
Cast Identification No.							
R.C.1 . .	+9·52	+ 7·90	−2·54	−1·33	+13·68	27·23	31
R.C.2 . .	+2·90	+ 1·20	−3·46	−0·91	+ 7·60	7·33	5
R.C.3 . .	−2·90	−10·38	−0·93	−0·72	+17·48	2·55	4
R.C.4 . .	+9·94	−14·98	+1·80	−0·14	+ 6·08	2·70	7

The relationship established between rephosphorisation and the ratio (CaO)/(FeO), based on the results of the first series of high-carbon (0·58–0·62% carbon) casts is therefore confirmed by the present series of results.

In order to examine the general application of this relationship to steels of other carbon ranges, a similar study was made on nine casts of lower-carbon steel (0·24–0·30% carbon). The procedure was as follows: The analysed phosphorus content of the metal at various stages of teeming is shown graphically in Fig. 29. The complete slag and metal analyses are shown in Appendix A, under the casts marked "T," for the Templeborough plant where these casts were made. It will be noted that Cast 31/572, denoted in Fig. 29 by T.8, was abnormal, the ferro-alloy additions to the ladle being made later than usual. It was therefore suspected that a portion of the ferro-manganese was trapped in the slag. This may be seen from the metal analysis, which shows the manganese content to be increasing in all the subsequent samples. The action of this trapped ferro-alloy as a strong reducing agent, may be responsible for the extensive

rephosphorisation found in this cast. In consequence it is thought advisable not to include cast T.8 in the following study.

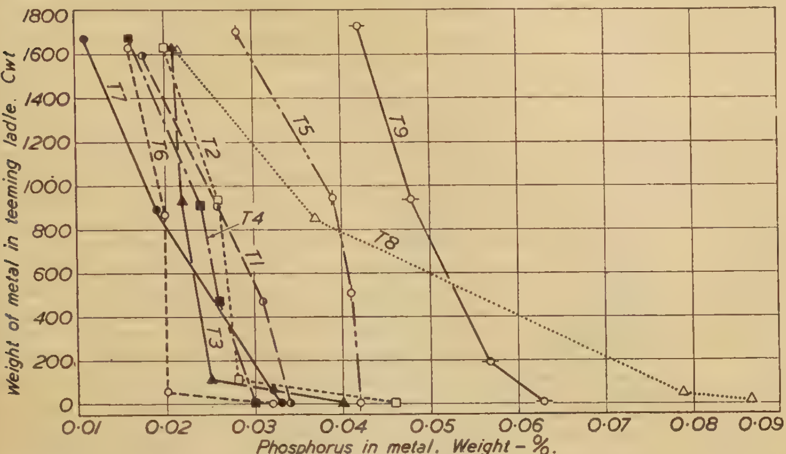


FIG. 29.—Phosphorus Pick-up in Ladle; 0·24–0·30% carbon casts.

In Table XV. are shown the free lime and the free iron oxide concentrations of the tapping slag and the amount of phosphorus reversion actually found for the "comparable casting period" of each cast. It may be mentioned that these free concentrations were based on the slag composition and temperature taken at tapping, instead of those of the final refining period which were used in the study of the high-carbon casts. This was done because the refining slags from the present casts were sampled some 7–65 min. before tapping and were therefore not suitable for use as a basis for the study of rephosphorisation. Since the foaming of the slag on tapping these low-carbon casts was not severe, it was possible to take immersion temperatures of the bath at tapping. The substitution of the tapping slag for the final refining slag, adopted in the present procedure, will not affect the final results in correlating rephosphorisation with the ratio (CaO)/(FeO), as may be seen in Appendix A, where the ultimate analyses of the final refining and the tapping slags on each of the high-carbon casts showed no appreciable differences.

The results given in Table XV. are shown graphically in Fig. 30.

TABLE XV.—*(CaO), (FeO) and Degree of Rephosphorisation on Low-Carbon Casts.*

Cast Identification No.	(CaO).	(FeO).	Ratio $\frac{(CaO)}{(FeO)}$.	Rephosphorisation during "Comparable Casting Period" (1 point ≡ 0·001% P).
T.1 . . .	29·91	6·40	$4{\cdot}67_5$	5 points
T.2 . . .	26·80	4·91	5·46	2 ,,
T.3 . . .	31·87	5·62	5·67	2 ,,
T.4 . . .	31·77	$5{\cdot}50_7$	5·77	2 ,,
T.5 . . .	27·51	5·62	$4{\cdot}89_5$	2 ,,
T.6 . . .	28·21	5·44	$5{\cdot}18_5$	Nil
T.7 . . .	23·93	6·22	$3{\cdot}84_5$	13 ,,
T.9 . . .	24·75	5·99	4·13	9 ,,

It may be seen from this diagram that a linear relationship between ephosphorisation and the ratio of the free lime to the free iron oxide concentration in the slag also exists on the low-carbon casts. The value of the ratio (CaO)/(FeO) corresponding to no rephosphorisation, as indicated in Fig. 30 for the 0·24–0·30% carbon range, is 5·15, which is considerably less than the corresponding value of 6·4 obtained previously for the higher-carbon casts. The higher temperature of the bath and the higher percentage of total iron in the slag normally carried in these low-carbon heats at tapping may account for this difference in value of the ratio (CaO)/(FeO) corresponding to no rephosphorisation. From Fig. 30 it may also be seen that for each unit drop in the ratio (CaO)/(FeO) the corresponding value of phosphorus reversion to the metal during the "comparable casting period" is 9·5 points, as compared with 19 points for the same period with the high-carbon heats. This agrees with the observation made in practice that a low-carbon steel is less susceptible to rephosphorisation than a high-carbon steel.

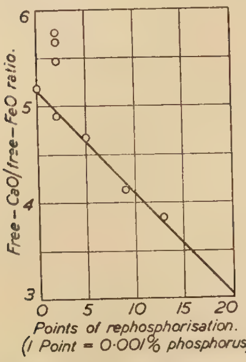


Fig. 30.—Correlation between Rephosphorisation and Free-CaO/Free-FeO Ratio for 0·24–0·30% carbon casts.

In the opinion of the author, the reason for this may lie in the fact that at equilibrium the iron oxide content of the metal was less when the steel contained 0·6% of carbon than when it contained 0·3%. Therefore, for a given amount of ferro-silicon and ferro-manganese added to the metal, the 0·6% carbon steel would have the greater amount of silicon and manganese available for the reduction of the phosphoric acid in the slag. It may be further stated that in the absence of sufficient quantities of the stronger reducing elements silicon and manganese, the action of carbon itself may play a direct rôle in reducing phosphorus from the slag. This state, however, is very rarely approached in open-hearth steelmaking practice. It will be seen in Appendix A. that the analysis of the pit samples showed only negligible change in the carbon content of the metal during the casting period.

The rephosphorisation values for each of the slag constituents and for temperature, such as those given in Tables XII., XII*a*., and XIV., were not calculated for this low-carbon series, since the figures shown in Table X. giving the degree of rephosphorisation per unit change in each of the factors temperature, $(\Sigma CaO)'$, (ΣSiO_2), (ΣMnO) and (ΣFe) are not applicable to the slag compositions and temperatures of the present casts. These do not come within the range covered by the compositions and temperatures from which Table X. has been drawn. A new set of diagrams similar to Figs. 23–27 would be required to be compiled on the following basis:

$$\begin{aligned} \text{Temperature} &= 1602^\circ \text{ C.} \\ (\Sigma CaO)' &= 40, \\ (\Sigma SiO_2) &= 10, \\ (\Sigma MnO) &= 5, \\ \text{and} \quad (\Sigma Fe) &= 15. \end{aligned}$$

The slag composition and temperature based on the above would then represent roughly the average of each of the respective factors found in these low-carbon casts; and when interpolated from Schenck's equilibrium diagrams this slag composition and temperature would give $(CaO) = 24{\cdot}85$ and $(FeO) = 4{\cdot}85_5$, resulting in a ratio $(CaO)/(FeO) = 5{\cdot}12$, which corresponds closely to the desired value of 5·15 for no rephosphorisation,

as indicated in Fig. 30. Curves prepared in this way would enable one to read off directly the influence on each of the factors temperature, $(\Sigma CaO)'$, &c., by a drop in the ratio (CaO)/(FeO) from 5 to 4, and would correspond to a rephosphorisation value of 9·5 points. Hence, by simple proportion, the conversion factors such as those shown in Table X. may be obtained for an approximate calculation of the expected rephosphorisation during the "comparable casting period" of the low-carbon casts.

Summary of Results.

(1) As far as the nineteen casts studied are concerned, covering two ranges of carbon from 0·24 to 0·62%, there is a linear relationship between the rephosphorisation of the metal and the ratio of the concentrations of free lime to free iron oxide in the slag.

(2) The tendency for the metal to rephosphorise increases with increasing carbon content.

(3) To prevent rephosphorisation on a 0·58–0·62% carbon cast, the final slag composition and temperature of the bath should be such that the value of the ratio (CaO)/(FeO) is not less than 6·4; whereas for a 0·24–0·30% carbon cast a value of 5·2 for this ratio should suffice.

The Physical Chemistry of Rephosphorisation.

It has been shown that for a given set of conditions the ratio (CaO)/(FeO) governs the tendency to rephosphorisation, and as this ratio is itself influenced by each of the slag constituents and by temperature, it may be accepted as a factor which integrates the influence of each of the slag constituents and temperature on the tendency to rephosphorisation. It may follow, therefore, that an explanation of why rephosphorisation should be governed by the ratio (CaO)/(FeO) will serve as a useful introduction to the study of the physical chemistry of rephosphorisation.

In attempting to elucidate the mechanism by which the ratio of free lime to free iron oxide governs rephosphorisation, two possible hypotheses may be considered.

Hypothesis A : *Based on the Physical Condition of the Slag.*

The formula :

$$\log [P] = \log K_p^{IV} + \log (\Sigma P_2O_5) - 5 \log (FeO) - 4 \log (CaO) - 0{\cdot}06(\Sigma P_2O_5)$$

given by Schenck and Riess [1] for phosphorus equilibrium, indicates that rephosphorisation will be promoted by the following factors :

(1) Raising the temperature of both the steel and the slag, owing to the fact that $\log K_p^{IV}$ increases with temperature at a more rapid rate than 5 log (FeO) and 4 log (CaO).

(2) Lowering (CaO).

(3) Lowering (FeO).

The results obtained in the present investigation, however, are not in full agreement with this interpretation of the above equation. Instead, it was established that a high (CaO)/(FeO) ratio does not rephosphorise and a low (CaO)/(FeO) ratio does rephosphorise. Whether the ratio was high or low (before the finishings were added), free lime and free iron oxide in the slags considered were present in sufficient quantities to reduce the phosphorus content of the metal to less than 0·03%. It is believed that at the end of the refining period, when equilibrium is almost attained, the iron oxide in the metal is related to the free iron oxide in the slag according to the partition coefficient $L_{FeO} = \frac{[FeO]_{metal}}{(FeO)_{slag}}$. When the finishings are added, the iron oxide in the metal is reduced and thrown out of balance with the

free iron oxide in the slag. If the slag remains sufficiently liquid for long enough and the agitation of the slag-metal interface is sufficient, iron oxide will pass from the slag to the metal and disappear there owing to reaction with the deoxidants. Further, on tapping, the slag picks up silica from the launder and the ladle, and the chemical effect of the silica on the slags of the casts under consideration was to decrease the free lime and increase slightly the free iron oxide in the slag, as shown by Schenck's equilibrium diagrams. A slag of a lower lime content would suffer a reduction of the concentrations of both free lime and free iron oxide. According to Schenck and Riess, then, an increase in the silica content of slag would result in changes in the concentrations of free lime and free iron oxide in such a manner as to encourage tendencies to rephosphorisation. Therefore, on the basis of the above considerations alone, rephosphorisation will always be expected to occur in the ladle; and it may be argued that the absence of rephosphorisation which has been found with the high (CaO)/(FeO) ratios is due to the fact that such ratios associated with a high slag viscosity. Thus, the absence of rephosphorisation may be due to the physical condition of the slag.

Hypothesis B: *Based on the Basicity of the Slag.*

The chief differences shown by the analyses of the tapping and the after-casting slags were an increase in the silica and a consequent decrease in the lime and iron contents. It is possible that a slag of high (CaO)/(FeO) ratio does not attack the fireclay bricks of the ladle lining so readily as a slag of low (CaO)/(FeO) ratio; consequently the latter slag would increase in silica content as a result of attack on the brick lining. The increased silica content would effect a diminution in the lime and the total iron concentrations by dilution and by the possible breakdown of the lime ferrites and subsequent reduction of the iron oxide, thus liberated, by silicon and manganese at the slag-metal interface. If phosphorus exists in slags as a lime phosphate compound, as indeed is the current belief, then the decrease in the basicity of the slag as a result of the silica pick-up would render possible the liberation and subsequent reduction into the metal of a portion of the phosphoric acid originally held as lime phosphate.

Examination of Hypotheses.

In order to examine the relative soundness of these hypotheses, the phosphorus content of the metal at the end of casting was calculated on Schenck's basis from the composition and temperature of the corresponding after-casting ladle slag. The results of the calculations are shown in Table XVI.

TABLE XVI.—*Phosphorus Content of Metal at End of Casting.*

Cast Identification Number.	Temperature. ° C.	(CaO).	(FeO).	[P] Calc. %.	[P] Analysed. %.
T.1	1533	23·53	3·58	0·023	0·034
T.3	1527	24·17	3·52	0·025	0·040
T.6	1500	23·45	3·20	0·014	0·032
T.9	1524	17·57	4·61	0·022	0·063
R.1	1524 *	22·00	2·80	0·091	0·103
R.2	1516 *	22·70	3·39	0·025	0·037
R.3	1513 *	20·30	3·08	0·061	0·053
R.4	1524 *	21·40	2·82	0·115	0·090
R.5	1463 *	21·20	2·00	0·072	0·056
R.6	1502 *	17·70	2·85	0·101	0·059

* Calculated from the tapping temperature by Land's method.[21]

In considering the results shown in Table XVI., it may be said that the calculated values would indicate a high, medium or low rephosphorisation. When the practical difficulties in the way of taking representative samples of slag and metal at the end of casting are considered and also the fact that temperatures were either taken in the tundish or calculated from those taken in the furnace, the results sufficiently indicate that viscosity has not prevented chemical reaction and the operation of the partition law from taking place. Thus the absence of rephosphorisation with slags of high (CaO)/(FeO) ratio cannot be attributed to the action of viscosity as a governing factor, though it undoubtedly plays some part in controlling rephosphorisation, as will be seen in the following discussion.

Fig. 31 shows a correlation of the ratio free lime to free iron oxide in the final furnace slag with the increase of the silica content of the ladle slag from the beginning to the end of casting. Since the increase of this compound in the slag was very much greater than, and out of proportion to, that which can be accounted for by oxidation of the amount of silicon likely to come from the metal during the casting period, the increased silica must have come from the lining of the ladle. Therefore, the value of the increase in silica, shown as abscissa in Fig. 31, may be taken to represent approximately the degree of erosion of the brick lining of the ladle.

The upper curve in Fig. 31 relates to the higher-carbon casts, whilst the lower curve was drawn for the lower-carbon casts. It is evident that for

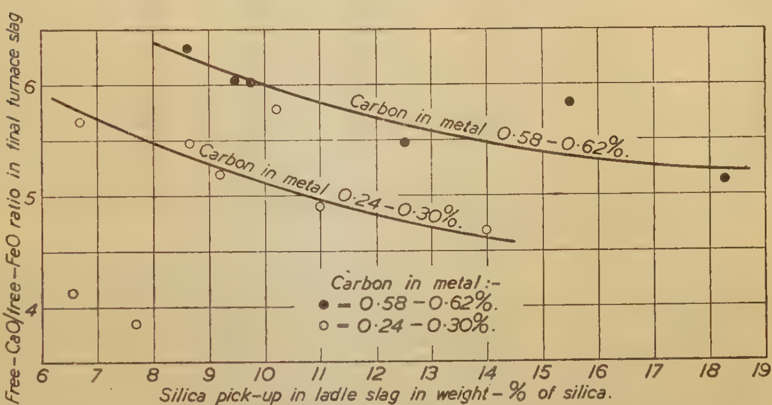


FIG. 31.—Correlation between Free-CaO/Free-FeO Ratio and Silica Increase in Ladle Slag.

the high-carbon casts there is a general tendency for the increase in silica in the ladle slag to intensify at decreasing values of the ratio (CaO)/(FeO) in the slag that was introduced into the ladle before the commencement of casting. The wide scatter shown by some of the points for the low-carbon casts may be attributed to the fact that the slag samples were not taken in all cases from the casting nozzle, as was done for the high-carbon casts; the only means of obtaining slag analyses in these instances was from the slag ingot used for slag volume determinations. Even so, there was a general tendency for the silica pick-up to increase with decreasing (CaO)/(FeO) ratio.

In the section " Application of Schenck's Equilibrium Diagrams to the Study of Ladle Rephosphorisation," it has been shown that the chief

factors in promoting a drop in the ratio (CaO)/(FeO) are an increase in the temperature of the slag and the metal, a decrease in the lime and an increase in the total iron content of the slag. It is certain that more erosion of the lining will take place at a high temperature than at a low; and a decrease in the lime and an increase in the iron content of a normal basic open-hearth slag will increase its fluidity, making it more reactive with the fire-clay brick lining. It would therefore seem that the experimental results given in Fig. 31 indicate the limited part played by viscosity in rephosphorisation.

In considering rephosphorisation in relation to the possible importance of compounds such as calcium silicate and calcium phosphate, the following relationship was found: The weight percentages of lime, silica and phosphoric acid in the slags under consideration were re-calculated so that $CaO + P_2O_5 + SiO_2 = 100\%$. The results are shown in Table XVII.

TABLE XVII.—*Slag Compositions (CaO, SiO_2 and P_2O_5).*

Cast Identification Number.	Tapping Slag.			Ladle Slag after Casting.		
	CaO. %.	SiO_2. %.	P_2O_5. %.	CaO. %.	SiO_2. %.	P_2O_5. %.
T.1	79·4	13·3	7·3	63·1	33·0	3·9
T.3	74·9	17·2	7·9	66·2	27·2	6·6
T.6	77·5	15·5	7·0	66·5	29·0	4·5
T.9	70·3	17·4	12·3	62·0	28·1	9·9
R.1	77·7	11·8	10·5	58·5	36·2	5·3
R.2	76·5	15·5	8·0	63·9	29·6	6·3
R.3	74·8	13·6	11·6	62·1	26·5	11·4
R.4	71·9	16·1	12·0	59·9	31·5	8·6
R.5	69·9	20·3	9·8	61·5	31·2	7·3
R.6	76·4	14·2	9·4	56·5	36·3	7·2
R.6	...	...	... *	74·8	17·4	7·8
R.6	...	...	... †	65·8	27·7	6·5

* Slag sampled 6 min. after the ladle was filled with metal and slag.
† Slag sampled 6 min. later.

The results given in Table XVII. are plotted in the three-component diagram, Fig. 32. The points denoted by solid circles and squares belong to those casts which showed pronounced rephosphorisation during the normal casting period. If a line is drawn in this figure joining the points *X* and *Y*, which correspond to the compositions dicalcium silicate and tricalcium phosphate respectively, it will be seen that for the after-casting ladle slags all the solid points appear on the right-hand side. All the points falling to the left-hand side of the line *XY* belong to the no-rephosphorisation casts, with the exception of one point (R.5) which was from a cast made at an exceptionally low temperature and in which the reaction between the slag and the metal was possibly not as great. The path of the change in the composition of the slag, in regard to the three components CaO, SiO_2 and P_2O_5, from tapping to the end of casting is also shown in Fig. 32 for one cast (R.6). The change in slag composition shown by the points R.6, R.6*a*, R.6*b*, &c., indicates that the erosion of brick lining was slow at first, then became more rapid and finally became slow again. During this attack on the lining the slag would become so enriched in silica that it could no longer hold the phosphoric acid as a stable compound against the reducing action of the metal; thus phosphorus would gradually pass from the slag to the metal.

Fig. 32 did not take into account, however, the effect of the alumina which was also present in the fireclay bricks. The increase in the alumina

content of the slag, resulting from its attack on the bricks, was from 2·5 to 3·5% for each cast. The effect of this increase in alumina content may

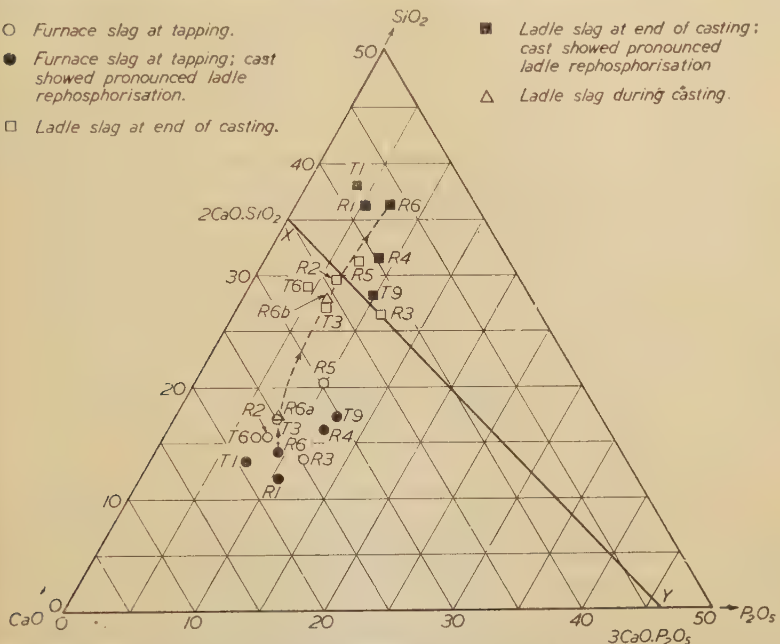


FIG. 32.—Three-Component Diagram Illustrating Changes in the Slag Composition (CaO, SiO_2, P_2O_5) between Tapping and End of Casting.

rotate the line XY in Fig. 32 in a downward direction about the point Y, owing to the formation of calcium aluminate, if such a compound exists in molten basic slag.

Attempts were made to obtain the relative degrees of attack on fireclay bricks by slags of various compositions. The experiment consists of melting a slag with separate known additions of lime, iron oxide and manganous oxide in fireclay pots. The melts were then analysed chemically and the silica figures were taken as a measure of the erosion of the pots. The results of the experiment showed severe attack on the fireclay pots by all the slags, but the difference in the degree of their attacks was negligible. The results seem to indicate, however, that there was a preferential attack by the slags on the free silica in the fireclay pots. The author realises that the conditions of this experiment were very different from those of steelmaking, and he believes that a study along more fundamental lines would be required to elucidate satisfactorily the problem of slag attack on refractory materials; but such an attempt would involve a very considerable amount of work and some practical difficulties might be encountered.

Consideration of the Present Position.

(a) *Constitution of Basic Slags.*

There is no direct means as yet available to determine the true constitution of molten basic slags, which will undoubtedly vary according to

the chemical composition of the slag and may possibly depend upon the temperature of the molten slag itself. In view of the unsatisfactory nature of our present state of knowledge on this subject, it was thought desirable to obtain some further information of the constitution of the solidified slags employed in the present study, so as to obtain a rough indication of the possible compounds present in these slags whilst in the molten state. It is realised that the rate of cooling of the samples may affect the results and that in the absence of high-temperature thermal data on the formation and dissociation of compounds, it will not be possible to conclude whether those compounds found in the solidified slags can be taken to represent the constitution of molten slags. Trömel,[22] in a recent investigation on the system $CaO–P_2O_5–SiO_2$, found that at temperatures above 1300° C. the compounds $3CaO.P_2O_5$ and $2CaO.SiO_2$ form a solid solution; and that $5CaO.P_2O_5.SiO_2$ (silico-carnotite) forms only when the melt is slowly cooled to below 1300° C.

A survey of the literature on the constitution of basic slags shows very divergent opinions. The following constitution forms the basis of Schenck's equilibrium diagrams:

Free Oxides.	Silicates.	Phosphates.	Ferrite.
CaO	$CaO.SiO_2$	$4CaO.P_2O_5$	$3CaO.Fe_3O_4$
MnO	$2MnO.SiO_2$	$3CaO.P_2O_5$	
FeO	$2FeO.SiO_2$		
SiO_2			

The general opinion of the various investigators appears to be, however, that in the presence of free lime the existence of iron and manganous silicates and free silica is doubtful. The compounds $2CaO.SiO_2$ and $2CaO.Fe_2O_3$ are generally believed to be the stable compounds at steel-making temperatures. Most of the investigators agree with Schenck in that they consider the phosphoric acid to be held in basic slags as a lime phosphate compound. A lime silica phosphate compound, corresponding to the formula $5CaO.P_2O_5.SiO_2$ (silico-carnotite), has, however, been found in solidified basic slags by Carnot and Richards,[23] Stead and Ridsdale,[24] Scott and McArthur,[25] and a few others.[26] The presence of the compound $9CaO.P_2O_5.3SiO_2$ in basic slags having certain SiO_2/P_2O_5 ratios has also been suggested by some investigators.[26] Kroll [27] claims to have separated the compound $6CaO.P_2O_5.2FeO.SiO_2$ (thomasite) from basic slags. In slags to which fluorspar has been added, part of the phosphoric acid is believed to be present as $3(3CaO.P_2O_5).CaF_2$ (fluorapatite).[28, 29]

In the opinion of the author, since the calculations based on Schenck's work have yielded results which agree with actual practice and there is not sufficient evidence to disprove entirely the basis upon which Schenck built his equilibrium diagrams, one can accept these diagrams as a reasonable basis to work upon. Schenck's hypotheses appear to be fundamentally sound, even though some of his minor assumptions require further confirmation. If, however, at a later date the slag constitution formulated by Schenck be proved incorrect, then modified calculations will have to be made.

During the present investigation, slag samples were forwarded for X-ray examination to Dr. A. H. Jay, who has since interpreted his results on slag constitution in relation to the slag/metal reaction. This new point of view is fully dealt with in Dr. Jay's paper.[31]

(b) *Opinion of Steelmakers.*

For the purpose of comparing the findings of the present study with practical observations, a number of prominent open-hearth steelmakers were asked to contribute their experience regarding the factors which they

found usually accompanied ladle rephosphorisation. A list of such factors is given below :

(1) A high tapping temperature.
(2) Insufficient lime content of the slag.
(3) Insufficient slag oxidation.
(4) A high residual manganese content in the bath.

Author's Notes on the Above.

A high tapping temperature and an insufficient lime content of the slag experienced in general practice with a rephosphorised cast agree completely with the present findings. In the case of insufficient slag oxidation, the author's opinion is that a certain minimum oxide content is necessary in the bath for dephosphorisation, which depends chiefly upon temperature and carbon content. In the ladle, a minimum free lime content appears to be the most important factor governing rephosphorisation; but this also implies a minimum iron oxide in the slag.

A high residual manganese content, assuming a given manganese content in the charge, may be the result of a high bath temperature, since the reaction represented by the equation : $FeO + Mn \rightleftharpoons MnO + Fe$ is known to be exothermic.[11] Schenck,[18] who studied the distribution of manganese in the basic open-hearth process after an extensive survey of the literature on this subject, arrived at the conclusion that the manganese content of the metal increased with rising temperature for a given slag composition.

(c) *Basic-Lined Ladle Experiment.*

As a final check on the hypothesis relating rephosphorisation to the basicity of the slag, a series of experiments were carried out on the Templeborough plant of Messrs. Steel, Peech and Tozer, using a ladle lined with dolomite bricks. The experiments consisted mainly of taking slag and metal samples from the furnace and from the ladle on a number of casts of steel in a manner similar to that employed in the foregoing work. A detailed account of these experiments is given in Appendix B.

Six casts of steel were studied in detail whilst using the basic-lined ladle. Two of these casts (36/6375 and 35/7922) were specially designed to tap with slags of compositions and temperatures which, on the basis of the present study, would produce marked rephosphorisation in the ladle. But as a basic-lined ladle was used, the results showed very little rephosphorisation at the end of casting. The slight increase in the phosphorus content of the last few hundredweights of metal in these instances is attributable to the increase of the silica content of the slag when it reacted with the bottom lining of the ladle and the stopper sleeves, which were made of the usual fireclay material for practical reasons. Confirmation of this effect may be seen from the analysis of the tapping and ladle slags, shown in Appendix B. A third cast (29/81) was tapped into the basic-lined ladle after a course of fireclay bricks had been placed on top of the basic bricks and held in position by means of ganister. The result was that the slag reacted with the fireclay material and there was a large increase in its silica content; rephosphorisation was observed on this cast. None of the remaining three casts nor three other casts of steel, ranging from 0·08 to 0·26% of carbon, using the basic-lined ladle before the above modification, showed rephosphorisation, as indicated by the analysis of the metal samples.

The concentrations of free lime and free iron oxide in the tapping slags from these basic-ladle casts, when interpolated from Schenck's equilibrium diagrams, are as shown in Table XVIII.

TABLE XVIII.—*(CaO), (FeO) and Ladle Rephosphorisation on Casts from a Basic-Lined Ladle.*

Cast No.	(CaO).	(FeO).	Ratio $\frac{(CaO)}{(FeO)}$.	Carbon in Steel. %.	Ladle Rephosphorisation "Comparable Period." Actual (Basic Ladle).	Ladle Rephosphorisation "Comparable Period." Estimated from Figs. 22 and 30, for Fireclay Brick Ladle.
36/6375	29·20	4·60	6·35	0·50	0·001% P	Nil.
40/2505	26·42	4·40	6·00	0·83	0·003% P	Well over 0·006% P
35/7922	30·80	6·14	5·00	0·33	0·002% P	0·002% P
34/3087	32·20	4·06	7·90	0·53	0·002% P	Nil.
31/1367	20·65	7·36	2·81	0·26	0·001% P	Over 0·020% P
29/81	23·27	5·78	4·04	0·29	0·006% P	Over 0·010% P

A distinction has already been made between rephosphorisation in the furnace and that in the ladle. The term "ladle rephosphorisation" was taken in the first place to mean the increase of phosphorus in the metal during casting, that is, between the first plate sample and last ingot teemed. It has now become apparent from the present work that phosphorus can return to the metal between taking the last sample in the furnace and taking the first plate sample. This may occur in the furnace owing to the disturbance of equilibrium conditions by the finishing additions or other changes occurring during tapping; but the present results do not afford the means of stating the precise mechanism by which this initial rephosphorisation takes place. Attempts made to correlate this occurrence with the analyses of tapping slags and rephosphorisation in the furnace and ladle do not indicate any definite relationship. It is believed, however, that the initial rephosphorisation, which in some instances reached a value of 0·008% of phosphorus by weight after a correction had been made for the amount of phosphorus contained in the ferro-manganese that was added in the ladle, may be related to the iron oxide in the metal and slag. It therefore appears that further study along this path would be required to elucidate the precise mechanism of rephosphorisation that occurs before the beginning of casting. The lack of means to determine the iron oxide content of the metal and the small magnitude of the phosphorus increase between tapping and casting, and that on furnace deoxidation, constitute the main difficulties in reaching a complete understanding of the mechanism of rephosphorisation in the basic open-hearth process. In general, however, the initial increase is small in comparison with the amount of rephosphorisation that may occur during the teeming of a cast, which has been shown by the present series of casts to be of any amount up to 0·04% of phosphorus by weight.

SUMMARY AND CONCLUSIONS.

(1) It has been shown that for a given set of working conditions, the rephosphorisation likely to occur in the ladle may be estimated in a quantitative manner by the application of Schenck's equilibrium diagrams to establishing a ratio of free lime to free iron oxide in the final furnace slag from its composition and the bath temperature.

(2) Rephosphorisation does not occur when the ladle is lined with basic material.

(3) Rephosphorisation of basic open-hearth steel during casting is due to a change in the composition of the slag caused by enrichment of the silica content, resulting from the reaction of the slag with the fireclay lining

of the ladle. It has not been possible to elucidate the mechanism involved in rephosphorisation and thereby determine whether the increase in silica from the ladle lining acts indirectly by altering the ratio of free lime to free iron oxide, or simply by rendering certain phosphorus compounds in the slag unstable. The difficulty in reaching any definite conclusion on this point is due to the fact that the ratio and the silica pick-up are so intimately related, that it is impossible to increase the silica content of the slag without decreasing the ratio.

(4) The degree of rephosphorisation taking place when a ladle lined with fireclay brick is used, may be controlled to a large extent by keeping the temperature of the slag and metal in the ladle as low as possible; and secondly, by adjusting the slag composition so that, in spite of its reaction with the ladle lining, the basicity of the slag will not fall below a certain minimum for holding phosphorus. The latter can be achieved, when circumstances permit, by increasing the lime content of the slag.

Appendix A.—*Details of the Casts Studied.*

The plant data were obtained on the basic open-hearth furnaces of the Templeborough and Rotherham melting shops of Messrs. Steel, Peech and Tozer (The United Steel Companies, Ltd.).

The furnaces used were fixed, cold-metal-charged, basic furnaces of 87 and 65 tons capacity. Brief descriptions of the general procedure of charging, melting, refining, tapping and casting these heats, and also of the methods for slag and metal sampling, are given below.

Procedure of Working a Cast.

Charging commenced on a hot furnace bottom and proceeded at such a rate that the furnace was neither overburdened with scrap, nor charged so slowly that excessive melting occurred before further additions of scrap were made. When all the steel scrap had been charged, sufficient time was allowed to melt down the charge reasonably low in the bath to accomodate comfortably the remainder of the materials. These consisted of approximately 60% of the total lime required, a small quantity of spar and the pig iron.

When the charge was completely melted, the balance of the lime was put into the furnace. With the assistance of a further small quantity of spar, as much lime as possible was permitted to be taken into solution in the slag before the addition of any oxide. The full quantity of oxide was then fed into the furnace without undue delay, in the form of mill scale or iron ore.

After a time the bath recovered from the depression of temperature due to the oxide feed, and developed a good consistent boil. This state is thought to signify the approach of equilibrium conditions and from this stage slag and metal samples were taken, and any further necessary additions to the furnace made.

With the slag and metal at the appropriate composition and the bath lying at the correct temperature, an addition of spiegel was made, followed quickly by the ferro-manganese finishing addition. The furnace was then tapped with the aid of an oxygen lance and the remainder of the deoxidants (ferro-silicon and ferro-manganese) were thrown into the stream as it flowed into the ladle.

The ladle, lined with high-alumina firebricks, was perfectly dry and fitted with a clay nozzle ($1\frac{1}{2}$ in.). After the ladle had filled and the metal was covered by a layer of slag, it was removed to the casting bay and the

metal was teemed uphill into groups of ingot moulds at a controlled speed.

Sampling Slag in the Furnace.

The slag sample was taken in a hemispherical steel spoon of 4 in. dia. The spoon was first warmed in the slag by being dipped into it vertically twice. A spoonful of slag from the main bulk was then taken through the middle door of the furnace. After withdrawal from the furnace, it was quickly poured into a "slag pan" 4 in. in dia. and 0·5 in. deep. The sample solidified rapidly and was left to cool in the "slag pan." The "pancake" sample was then crushed in a steel mortar and any metallic iron was removed by a magnet.

Metal Sampling in the Furnace.

The metal sample was taken immediately after the slag sample. The warmed spoon used for sampling the slag was used for the metal. The sample was taken through the middle door and from the main body of the bath. After withdrawal from the furnace, any slag attached to the surface of the metal in the spoon was removed quickly by a short stick; and the metal was then deoxidised with 3 in. of aluminium wire (16 S.W.G.) and poured into a tapered steel mould. Drillings taken from the bottom third of this test-piece were used for chemical analysis.

Bath Temperature Measurements.

Bath temperatures were taken at a standard position in the bath, through the middle door, by means of a quick-immersion pyrometer. Each temperature reading was made immediately before taking the corresponding set of slag and metal samples.

Slag Sampling in the Ladle.

Ladle slags were taken through the teeming nozzle as soon as the ladle was empty of metal. A preheated mould of the type used for sampling ladle metal was used for this purpose. It may be added that if a cold mould was used, the slag would foam in the mould and result in a spongy sample. On one cast (22/2309) ladle slag samples at two other stages during teeming were taken from the crane-box platform by lowering a sampling mould into the slag which bubbled through the space between the crust and the inside wall of the ladle. It is doubtful, however, whether these samples can be used to represent the slag at the slag-metal interface in the ladle.

Metal Sampling in the Ladle.

Ladle metal samples, or the so-called "pit samples," were taken at the casting bay from the stream of metal leaving the ladle nozzle. All samples, with the exception of the final ladle metal, were taken when the metal in the uphill-cast ingot moulds had risen to a predetermined height; this enabled the weight of metal still remaining in the ladle at the time a sample was taken, to be calculated from the actual weight of ingots teemed.

Temperature Measurements in the Ladle.

These temperatures were taken in the tundish of the "spare" ingot by means of a hand-manipulated immersion pyrometer, built along similar lines to the one used for taking bath temperatures.

Materials Charged.

Details of the metal charged on the observed casts are given in Table XIX.

Analytical Data of Slag and Metal Samples.

The slag and metal sample analyses, together with the corresponding immersion temperatures of the bath, taken when melted, during final refining and at tapping, are shown in Table XX.

Slag and metal analyses and temperatures of the metal taken at the casting bay during the process of teeming the metal are shown in Table XXI.

Details of the ferro-alloy finishings made to the furnace and to the ladle are given in Table XXII.

TABLE XIX.—*Composition of the Charges.*

Average Phosphorus Content of Charged Material:	1·5%.	0·15%.	0·8%.	0·05%.	Total Phosphorus Content of the Charge. %.
Plant and Cast Number.	Wt. of Pig Iron. Tons.	Wt. of Ingot Mould. Tons.	Wt. of Cast Iron. Tons.	Wt. of Steel Scrap. Tons.	
T.34/2238	19·1	4·0	...	64·3	$0·371_5$
T.30/4600	32·1 *	6·4	...	45·8	0·624
T.31/501	32·9 *	5·1	...	46·0	0·624
T.31/508	38·5 *	...	...	46·6	0·706
T.35/7148	21·1	6·6	...	59·6	0·408
T.35/7156	20·8	4·4	...	61·8	0·402
T.31/599	18·7	5·5	...	63·3	0·366
T.31/572	22·0	6·2	...	59·3	0·422
T.30/4591	32·6 *	5·5	...	46·7	0·614
R.25/1518	18·1	...	...	46·0	0·459
R.25/1521	18·0	...	...	47·0	0·450
R.23/3736	20·0 *	6·8	...	37·0	0·516
R.25/1527	19·6	...	...	45·0	0·490
R.25/1528	9·3	...	10·0	45·0	0·376
R.22/2309	15·0	4·0	...	38·0	0·439
R.C.22/2418	18·0	4·5	...	42·2	0·460
R.C.22/2420	16·1	...	4·0	44·6	0·457
R.C.22/2422	19·1	...	...	45·5	0·480
R.C.22/2436	18·0	...	...	46·0	0·458

* These casts contained a high percentage of pig iron, as they were "lying-in" charges at the week-end or due to a temporary shortage of steel scrap. To counteract the carbon balance in those charges due to shortage of scrap, varying amounts of oxide up to 5 tons were put in with the charge.

TABLE XX.—*Furnace Samples : Slag and Metal Analyses.*

Cast Identification No.	Plant and Cast No.	Condition of Bath.	Clock Time.	Bath Temperature. ° C.	Metal Analysis.					Slag Analysis.								
					C. %.	Si. %.	S. %.	P. %.	Mn. %.	SiO_2. %.	Fe. %.	CaO. %.	P_2O_5. %.	MnO. %.	MgO. %.	S. %.	FeO. %.	Fe_2O_3. %.
T.1	Templeboro' 34/2238	Melted	12.20	...	0·44	0·001	0·054	0·048	0·33$_5$	16·36	5·80	51·10	7·26	5·67	5·33	0·42	4·80	3·00
		Refining	13.45	1623	0·13	0·001	0·045	0·020	0·17	8·46	13·50	51·70	5·04	5·74	6·49	0·39	12·80	5·10
		Tapping	14.16	1620	0·14	0·003	0·040$_5$	0·017$_5$	0·62	8·40	14·80	49·90	4·58	6·00	6·05	0·39	13·50	6·10
T.2	,, 30/4600	Melted	13.45	1537	0·65	0·004	0·076	0·227	0·25	17·70	4·40	53·20	10·65	4·96	3·96	0·21	3·74	2·14
		Refining	15.52	1550	0·31	0·014	0·034$_5$	0·025	0·12	10·20	12·70	51·24	8·47	4·32	3·78	0·30	11·87	5·00
		Tapping	16.30	1575	0·22	0·004	0·037$_5$	0·020	0·46	9·54	11·60	50·68	7·68	4·63	4·04	0·33	12·64	2·57
T.3	,, 31/501	Melted	10.25	1564	0·25	0·000	0·043$_5$	0·248	0·12	13·00	11·70	50·40	6·22	4·70	3·34	0·40	11·00	4·40
		Refining	11.00	1613	0·15	0·004	0·035	0·018	0·17	12·30	11·70	52·40	5·49	4·33	4·00	0·39	11·70	3·70
		Tapping	11.15	1618	0·17	0·004	0·036	0·021	0·56	12·10	11·90	52·80	5·53	5·06	3·63	0·37	11·40	4·40
T.4	,, 31/508	Melted	11.05	...	0·33	0·001	0·045$_5$	0·017$_5$	0·18	14·50	8·60	54·10	3·41	4·68	6·63	0·42	8·40	3·00
		Refining	11.43	1619	0·19	0·002	0·046	0·012	0·17	12·56	12·30	50·65	3·43	5·15	6·73	0·41	12·50	3·70
		Tapping	11.50	1610	0·19	0·004	0·041	0·016	0·54	12·50	12·60	50·50	3·50	4·96	6·63	0·41	12·00	4·70
T.5	,, 35/7148	Melted	10.20	...	0·42	0·002	0·039	0·039$_5$	0·28	12·70	6·20	51·50	9·16	6·76	4·86	0·42	4·90	3·40
		Refining	11.00	1586	0·26	0·004	0·034	0·024	0·19	10·80	12·00	51·50	7·56	3·15	4·86	0·42	10·40	5·60
		Tapping	11.30	1599	0·27	0·004	0·036	0·027	1·41	10·50	12·40	50·00	6·64	5·21	6·80	0·39	11·00	5·60
T.6	,, 35/7156	Melted	14.30	1536	0·29	0·001	0·052	0·029	0·20	16·60	6·60	51·00	7·56	6·19	4·68	0·28	5·00	3·80
		Refining	15.35	1570	0·14	0·075	0·031	0·030	0·20	11·80	15·20	51·00	5·50	5·50	4·32	0·34	14·40	5·70
		Tapping	16.40	1581	0·11	0·016	0·026$_5$	0·016	0·24	10·20	16·00	51·00	4·58	6·32	5·22	0·35	15·60	5·60
T.7	,, 31/599	Melted	10.20	1574	0·48	0·009	0·032	0·027	0·24	14·76	7·92	52·90	6·75	5·53	3·42	0·33	6·45	4·15
		Refining	11.45	1574	0·13	0·016	0·025	0·023	0·13	9·56	14·44	49·60	5·68	5·83	4·17	0·33	14·20	4·86
		Tapping	12.10	1585	0·13	0·002	0·026	0·011	0·45	9·32	16·23	46·50	5·32	6·30	5·98	0·30	13·80	7·87
T.8	,, 31/572	Melted	12.00	1549	0·23	0·000	0·050	0·038	0·22$_5$	14·60	7·00	51·90	7·44	5·99	4·76	0·33	5·40	4·00
		Refining	13.10	1607	0·15	0·004	0·032	0·024	0·16	11·20	10·90	53·20	5·79	5·73	4·87	0·39	10·10	4·40
		Tapping	13.17	1599	0·19	0·002	0·037	0·022$_5$	0·84	10·40	11·30	52·75	5·45	6·81	5·57	0·38	10·30	4·70
T.9	,, 30/4591	Melted	10.15	1494	0·80	0·017	0·064	0·137	0·31	21·50	5·10	44·60	11·52	7·44	4·29	0·23	2·59	4·41
		Refining	12.50	1602	0·24	0·011	0·046	0·031	0·27	12·80	11·50	50·00	9·16	5·69	5·14	0·43	8·64	6·84
		Tapping	13.10	1604	0·27	0·012	0·042	0·042	0·87	12·46	12·00	50·40	8·81	5·69	5·98	0·44	...	...

R.1	Rotherham	25/1518	Melted	9.08	...	0·92	0·003	0·019	0·081	0·51	10·76	4·20	56·60	9·83	5·93	3·89	0·29	1·58	4·24
			Refining	10.35	1620	0·58	0·000	0·019	0·023	0·21	8·44	11·80	51·30	7·10	6·71	4·87	0·24	10·58	5·15
			Tapping	10.45	1614 *	0·58	0·004	0·017	0·028	0·40	7·78	11·70	51·00	6·90	7·92	5·55	0·23	10·32	5·30
R.2	,,	25/1521	Melted	15.45	...	0·62	0·004	0·032	0·025	0·28	14·00	6·80	53·60	7·45	6·04	6·10	0·31	4·68	4·52
			Refining	17.20	1601	0·36	0·000	0·026	0·016	0·17	10·04	11·80	51·70	5·41	6·66	6·31	0·27	12·00	3·58
			Tapping	17.28	1595 *	0·35	0·005	0·023	0·018	0·39	10·40	12·10	51·70	5·39	6·68	7·00	0·28	10·94	5·14
R.3	,,	23/3736	Melted	11.15	...	1·15	0·002	0·021	0·130	0·53	11·70	6·90	53·30	9·79	4·90	6·12	0·35	6·32	2·86
			Refining	14.00	1598	0·55	0·001	0·024	0·021	0·19	9·64	10·70	54·70	8·56	5·53	6·55	0·35	10·10	4·15
			Tapping	14.05	1592 *	0·57	0·004	0·023	0·022	0·42	9·94	10·50	54·80	8·43	5·70	6·44	0·36	9·93	4·00
R.4	,,	25/1527	Melted	1.55	...	0·87	0·004	0·028	0·079	0·41	12·90	6·70	50·50	9·58	4·25	4·54	0·29	4·77	4·29
			Refining	2.46	1610	0·54	0·001	0·025	0·029	0·26	11·32	9·20	51·50	8·54	5·98	6·33	0·34	8·51	3·72
			Tapping	2.55	1604 *	0·55	0·002	0·024	0·039	0·48	11·40	9·00	50·80	8·47	6·54	6·44	0·33	7·87	4·15
R.5	,,	25/1528	Melted	11.27	...	0·75	0·008	0·053	0·041	0·23	17·80	5·60	49·30	8·76	6·89	6·20	0·27	4·82	2·64
			Refining	12.28	1556	0·63	0·008	0·038	0·018	0·21	14·96	9·90	50·00	6·90	6·20	6·30	0·36	8·64	4·57
			Tapping	12.37	1550 *	0·66	0·004	0·031	0·018	0·86	14·48	9·30	49·80	6·94	5·95	6·20	0·36	9·03	3·29
R.6	,,	22/2309	Refining	11.15	1597	0·53	0·004	0·032	0·022	0·21	9·58	11·60	52·60	6·75	6·10	6·52	0·37	10·96	4·40
			Tapping	11.22	1591 *	0·56	0·000	0·031	0·027	0·58	9·80	11·20	52·60	6·48	6·70	6·94	0·36	9·94	4·96
R.C.1	,,	22/2418	Melted	17.43	...	...	...	...	0·219	...	...	...	...	...	...	...	...	...	...
			Refining	19.55	1600	...	...	...	0·021	0·20	8·96	11·80	52·50	8·89	6·75	5·23	...	...	...
			Tapping	20.04	1594 *	...	...	...	0·021	...	8·84	12·00	52·40	8·63	6·20	5·18	...	...	...
R.C.2	,,	22/2420	Melted	16.26	...	...	...	...	0·070	...	...	...	...	...	...	...	...	...	...
			Refining	17.58	1584	...	...	...	0·014	0·18	8·58	11·00	52·40	8·04	6·20	4·69	...	...	...
			Tapping	18.19	1578 *	...	...	...	0·020	...	8·28	10·90	52·20	7·72	6·76	4·77	...	...	...
R.C.3	,,	22/2422	Melted	15.00	...	...	...	...	0·027	...	...	...	...	...	...	...	...	...	...
			Refining	16.20	1570	...	...	...	0·012	0·16	9·62	12·30	52·70	6·87	5·95	4·37	...	...	...
			Tapping	16.28	1564 *	...	...	...	0·015	...	9·76	12·50	52·60	6·83	6·45	4·83	...	...	...
R.C.4	,,	22/2436	Melted	10.50	...	...	...	...	0·249	...	...	...	...	...	...	...	...	...	...
			Refining	12.45	1601	...	...	...	0·019	0·17	10·74	10·80	54·32	7·36	5·18	6·64	...	...	...
			Tapping	13.00	1595 *	...	...	...	0·022	...	10·52	10·50	53·80	7·01	5·67	6·67	...	...	...

* It has been determined in previous investigations that each hundredweight of finished additions (excluding large additions of ferro-silicon) caused a drop in bath temperature of 1·3° C. The tapping temperatures, when not otherwise available, have been calculated on this basis.

Table XXI.—*Ladle Samples : Slag and Metal Analyses.*

Cast Identification No.	Cast No.	Initial Weight of Metal in Ladle. Cwt.	Sample No.	Weight of Metal in Ladle at Time of Sampling. Cwt.	Metal Analysis. C. %.	Metal Analysis. Si. %.	Metal Analysis. S. %.	Metal Analysis. P. %.	Metal Analysis. Mn. %.	Metal Temperature. ° C.	Slag Analysis. Slag Sample.§	Slag Analysis. SiO_2. %.	Slag Analysis. Fe. %.	Slag Analysis. CaO. %.	Slag Analysis. P_2O_5. %.	Slag Analysis. MnO. %.	Slag Analysis. MgO. %.	Slag Analysis. S. %.	Slag Analysis. FeO. %.	Slag Analysis. Fe_2O_3. %.	Phosphorus in Metal at Tapping. %.
T.1	34/2238*	1593	1	905·0	0·24	0·205	0·037	0·026	1·06	...	T	8·40	14·80	49·90	4·58	6·00	6·05	0·39	13·50	6·10	0·0175
			2	467·0	...	...	0·039	0·031	1·06	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·24	0·116	0·045	0·034	0·90	...	FL	22·40	8·90	42·80	2·62	9·61	4·72	0·28	8·00	3·80	
T.2	30/4600*	1628	1	923·0	0·33	0·211	0·033	0·026	1·16	...	T	9·54	11·60	50·68	7·68	4·63	4·04	0·33	12·64	2·57	0·020
			2	109·0	...	...	0·033	0·028	1·12	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·30	0·103	0·041	0·046	0·84	1502	FL	18·20	8·40	46·34	6·53	6·88	3·60	0·27	9·16	1·86	
T.3	31/501*	1628	1	923·0	0·26	0·172	0·033	0·023	1·06	...	T	12·10	11·90	52·80	5·53	5·06	3·63	0·37	11·40	4·40	0·021
			2	109·0	...	...	0·035	0·025	1·05	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·26	0·074	0·047	0·040	0·71	1527	FL	18·80	9·30	46·00	4·54	6·45	3·33	0·35	8·80	3·60	
T.4	31/508*	1671	1	910·0	0·26	0·190	0·046	0·024	1·02	...	T	12·50	12·60	50·50	3·50	4·96	6·63	0·41	12·00	4·70	0·016
			2	472·0	...	...	0·046	0·026	0·98	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·28	0·138	0·046	0·030	0·89	1533	FL	22·74	8·00	45·55	3·16	8·55	3·75	0·33	8·90	1·60	
T.5	35/7148*	1704	1	943·0	0·26	0·204	0·039	0·039	1·04	...	T	10·5	12·40	50·00	6·64	5·21	6·80	0·39	11·00	5·60	0·027
			2	505·0	...	...	0·039	0·041	1·00	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·24	0·132	0·036	0·042	0·92	1533	FL	21·50	8·70	47·00	5·27	8·87	5·22	0·33	8·30	3·30	
T.6	35/7156*	1626	1	865·0	0·28	0·196	0·035	0·020	1·10	...	T	10·20	16·00	51·00	4·58	6·32	5·22	0·35	15·60	5·60	0·016
			2	55·0	...	...	0·037	0·020	1·04	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·28	0·105	0·034	0·032	0·89	1500	FL	19·40	11·40	44·50	2·98	8·82	4·86	0·29	11·00	4·10	
T.7	31/599*	1675	1	892·0	0·27	0·186	0·029	0·019	1·02	...	T	9·32	16·23	46·50	5·32	6·30	5·98	0·30	13·80	7·87	0·011
			2	60·0	...	...	0·030	0·032	1·01	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·26	0·080	0·031	0·034	0·76	1529	FL	17·00	13·52	41·50	3·92	7·26	5·83	0·31	14·70	3·00	
T.8	31/572*‡	1623	1	851·0	0·30	0·208	0·040	0·037	1·08	...	T	10·40	11·30	52·75	5·45	6·81	5·57	0·38	10·30	4·70	0·0225
			2	38·0	...	...	0·041	0·079	1·45	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·36	0·151	0·036	0·087	1·64	...	FL	16·90	9·00	47·50	4·78	7·27	3·78	0·36	8·80	3·10	
T.9	30/4591*	1725	1	935·0	0·30	0·164	0·043	0·048	1·12	...	T	12·46	12·00	50·40	8·81	5·69	5·98	0·44	...	...	0·042
			2	185·0	...	...	0·043	0·057	1·11	...	...	...	...	...	...	...	...	...	...	...	
			3	Nil	0·31	0·166	0·043	0·063	0·99	1524	FL	19·00	12·10	41·80	6·69	8·68	5·15	0·35	...	...	
R.1	25/1518†	1172	1	888·5	0·58	0·257	0·028	0·027	0·70	...	T	7·78	11·70	51·00	6·90	7·92	5·55	0·23	10·32	5·30	0·028
			2	577·0	0·59	0·242	0·027	0·033	0·73	...	...	...	...	...	...	...	...	...	...	...	
			3	306·0	0·59	0·235	0·026	0·039	0·70	...	...	...	...	...	...	...	...	...	...	...	
			4	79·5	0·59	0·229	0·025	0·051	0·71	...	...	...	...	...	...	...	...	...	...	...	
			5	19·0	0·59	0·216	0·027	0·065	0·70	...	...	...	...	...	...	...	...	...	...	...	
			6	Nil	0·56	0·105	0·023	0·103	0·58	...	FL	26·08	4·50	42·40	3·82	8·85	5·58	0·15	5·54	0·29	
R.2	25/1521†	1150·75	1	844·75	0·47	0·222	0·033	0·022	0·88	...	T	10·40	12·10	51·70	5·39	6·68	7·00	0·28	10·94	5·14	0·018
			2	545·0	0·46	0·222	0·034	0·022	0·84	...	...	...	...	...	...	...	...	...	...	...	
			3	287·5	0·47	0·220	0·030	0·023	0·82	...	...	...	...	...	...	...	...	...	...	...	
			4	34·0	0·46	0·170	0·031	0·023	0·72	...	...	...	...	...	...	...	...	...	...	...	
			5	Nil	0·44	0·168	0·035	0·037	0·46	...	FL	20·16	9·10	43·60	4·28	7·49	5·95	0·25	11·22	0·57	

R.3	23/3736†	1165·75	1	859·75	0·63	0·289	0·031	0·030	0·84	...	T	9·94	10·50	54·80	8·43	5·70	6·44	0·36	9·93	4·00	0·022
			2	560·0	0·61	0·283	0·030	0·032	0·79	...	...	...	...	...	...	...	...	...	...	...	
			3	302·5	0·63	0·270	0·026	0·032	0·79	...	...	...	...	...	...	...	...	...	...	...	
			4	49·0	0·61	0·237	0·029	0·032	0·73	...	...	...	...	...	...	...	...	...	...	...	
			5	4·0	0·63	0·209	0·046	0·037	0·68	...	...	...	...	...	...	...	...	...	...	...	
			6	Nil	0·63	0·121	0·044	0·053	0·49	...	FL	19·42	7·00	45·70	8·36	8·24	6·08	0·26	8·13	1·00	
R.4	25/1527†	1128·75	1	830·75	0·55	0·233	0·031	0·047	0·78	...	T	11·40	9·00	50·80	8·47	6·54	6·44	0·33	7·87	4·15	0·039
			2	286·5	0·56	0·233	0·030	0·051	0·76	...	...	...	...	...	...	...	...	...	...	...	
			3	76·2	0·57	0·224	0·029	0·064	0·74	...	...	...	...	...	...	...	...	...	...	...	
			4	19·0	0·56	0·147	0·032	0·086	0·60	...	...	...	...	...	...	...	...	...	...	...	
			5	Nil	0·54	0·116	0·034	0·090	0·52	...	FL	23·94	5·00	45·50	6·53	7·16	5·20	0·24	6·06	0·43	
R.5	25/1528†	1274·5	1	976·5	0·62	0·227	0·033	0·028	0·85	...	T	14·48	9·30	49·80	6·94	5·95	6·20	0·36	9·03	3·29	0·018
			2	683·75	0·61	0·227	0·035	0·029	0·86	...	...	...	...	...	...	...	...	...	...	...	
			3	432·25	0·61	0·224	0·034	0·029	0·86	...	...	...	...	...	...	...	...	...	...	...	
			4	184·75	0·60	0·211	0·034	0·029	0·83	...	...	...	...	...	...	...	...	...	...	...	
			5	Nil	0·53	0·049	0·058	0·056	0·33	...	FL	23·10	6·00	45·60	5·39	7·45	4·97	0·29	6·84	1·00	
R.6	22/2309†	1192	1	896·0	0·65	0·248	0·031	0·031	0·75	...	T	9·80	11·20	52·60	6·48	6·70	6·94	0·36	9·94	4·96	0·027
			2	563·0	0·63	0·255	0·031	0·032	0·74	...	...	...	...	...	...	...	...	...	...	...	
			3	277·0	0·64	0·205	0·031	0·036	0·73	...	L2	11·64	9·90	50·00	5·20	6·38	9·75	0·34	10·58	2·43	
			4	124·0	0·62	0·172	0·034	0·042	0·66	...	L7	19·40	7·10	45·60	4·56	6·45	4·14	0·27	8·51	0·72	
			5	Nil	0·60	0·166	0·047	0·059	0·43	...	FL	25·30	6·10	39·50	5·06	7·80	3·67	0·19	...	...	
R.C.1	22/2418†	1225·5	1	903·5	0·64	0·272	...	0·026	0·72	...	T	8·84	12·00	52·40	8·63	6·20	5·18	...	...	...	0·021
			2	549·5	0·63	0·272	0·027	0·028	0·73	...	...	...	...	...	...	...	...	...	...	...	
			3	195·5	...	...	0·025	0·031	0·69	...	...	...	...	...	...	...	...	...	...	...	
			4	30·0	...	...	...	0·057	...	...	...	...	...	...	...	...	...	...	...	...	
			5	Nil	...	...	...	0·072	...	...	...	...	...	...	...	...	...	...	...	...	
R.C.2	22/2420†	1249	1	967·25	0·59	0·261	...	0·030	0·80	...	T	8·28	10·90	52·20	7·72	6·76	4·77	...	...	...	0·020
			2	657·5	0·59	0·263	0·024	0·030	0·80	...	...	...	...	...	...	...	...	...	...	...	
			3	388·0	...	...	0·024	0·030	0·76	...	...	...	...	...	...	...	...	...	...	...	
			4	162·75	...	...	...	0·030	...	...	...	...	...	...	...	...	...	...	...	...	
			5	36·0	...	...	...	0·035	...	...	...	...	...	...	...	...	...	...	...	...	
			6	Nil	...	...	...	0·046	...	...	...	...	...	...	...	...	...	...	...	...	
R.C.3	22/2422†	1269	1	987·25	0·63	0·242	...	0·022	0·80	...	T	9·76	12·50	52·60	6·83	6·45	4·83	...	...	...	0·015
			2	677·5	0·63	0·242	0·031	0·022	0·78	...	...	...	...	...	...	...	...	...	...	...	
			3	408·0	...	...	0·031	0·023	0·76	...	...	...	...	...	...	...	...	...	...	...	
			4	182·75	...	...	...	0·026	...	...	...	...	...	...	...	...	...	...	...	...	
			5	122·5	0·62	0·224	...	0·026	0·76	...	...	...	...	...	...	...	...	...	...	...	
			6	Nil	...	...	...	0·048	...	...	...	...	...	...	...	...	...	...	...	...	
R.C.4	22/2436†	1136	1	792·0	0·59	0·297	...	0·026	0·88	...	T	10·52	10·50	53·80	7·01	5·67	6·67	...	...	...	0·022
			2	416·0	0·59	...	0·027	0·028	0·87	...	...	...	...	...	...	...	...	...	...	...	
			3	38·0	...	...	0·029	0·033	0·76	...	...	...	...	...	...	...	...	...	...	...	
			4	Nil	...	...	...	0·040	...	...	...	...	...	...	...	...	...	...	...	...	

* Templeborough plant. † Rotherham plant.
‡ Templeborough cast 31/572.—Ferro-alloy additions to the ladle made later than normal; some trapped ferro-manganese in slag suspected.
§ T = Tapping slag. L2 = Ladle slag 2 min. after commencincing teeming. L7 = Ladle slag 7 min. after commencing teeming. FL = Final ladle slag.

TABLE XXII.—*Ferro-Alloy Additions.*

Plant and Cast No.		In the Furnace.	In the Ladle.	
		Ferro-Manganese.* Cwt.	Ferro-Manganese.* Cwt.	Ferro-Silicon. † Cwt.
Templeboro'	34/2238	14·5	10·0	4·5
,,	30/4600	15·0	11·5	4·5
,,	31/501	15·0	11·0	4·5
,,	31/508	14·5	9·5	4·5
,,	35/7148	14·5	9·5	4·5
,,	35/7156	14·5	12·0	4·5
,,	31/599	14·5	10·0	4·5
,,	31/572	14·5	11·0	4·5
,,	30/4591	15·0	9·5	4·5
Rotherham	25/1518	5·0	6·0	4·5
,,	25/1521	5·0	8·0	4·5
,,	23/3736	5·0	6·0	5·0
,,	25/1527	5·0	6·0	5·0
,,	25/1528	5·0	8·5	5·0
,,	22/2309	5·0	5·5	8·0
Rotherham (checking)	22/2418	5·0	7·0	8·5
,, ,,	22/2420	5·0	7·5	8·5
,, ,,	22/2422	5·0	7·5	8·0
,, ,,	22/2436	5·0	7·5	8·5

* Typical analysis of the ferro-manganese used is as follows:

Carbon.	Silicon.	Sulphur.	Phosphorus.	Manganese.
7·0%	1·0%	trace	0·30%	78·0%

† Ferro-silicon used contained 75·0% silicon.

APPENDIX B.—*Details of the Basic-Lined Ladle Experiments.*

The refractory lining of the ladle was composed of stabilised dolomite, with the exception of the bottom and the backing adjacent to the metal casing. Fireclay material was used for lining the bottom and for the stopper-rod sleeves. It was expected that less slag and skull would form on the firebrick bottom than on a dolomite brick bottom. This skull would have to be removed before resetting the next nozzle and also to allow following casts to drain properly. Any reaction between the slag and the bottom could only take place when the ladle was almost emptied of metal. This bottom was therefore considered sufficiently suitable for these experiments.

The basic lining consisted of a 4½-in. thickness of unfired stabilised dolomite brick and a basic cement essentially of finely ground stabilised dolomite clinker. The manufacturing processes of the brick and the cement were described by Swinden and Chesters.[30] A typical analysis of the dolomite clinker, from which the brick and the cement used in the ladle in question were made, is recorded below:

Stabilised Dolomite Clinker.

CaO	40·04%
MgO	40·30%
SiO_2	14·44%
Fe_2O_3	3·44%
Al_2O_3	1·50%
Loss on ignition	0·25%

During the trial of this ladle, it was observed that after each cast a considerable amount of slag had gathered over the entire surface of the dolomite bricks and that there was a tendency for the metal to form a skull. The amount of slag and metal growth was such that at the end of the eighth cast, the capacity of the ladle had to be restored by means of an extra course of bricks built on the top of the existing lining. For this purpose a top course of fireclay bricks was introduced and these were held in position by the usual brick-setting ganister, which contained about 84% of silica. The result was that, on the ninth cast (29/81) the basic slag reacted with the ganister and its silica content increased from 10·8% at tapping to 21·8% at the end of casting; and that the analysis of the metal samples showed rephosphorisation had occurred in the ladle. The fireclay brick top course was then removed.

The accumulated slag and metal were next removed, and in so doing two small patches of the dolomite lining were damaged. These were, however, repaired with new dolomite bricks and the ladle was again put into commission. After a few more casts the slag-skull formed and again reduced the capacity. A top course of dolomite bricks was built and no fireclay brick was used anywhere above the bottom.

Among the casts studied, two (36/6375 and 35/7922) were specially designed to tap with slags of such composition and temperature as would, on the basis of the foregoing work, produce pronounced rephosphorisation in the ladle had it been lined with the usual fireclay brick. The analyses of the metal samples taken during the various stages of teeming the twenty casts made in this ladle showed that the rephosphorisation between the first and the last plate on any of the casts, other than the one already mentioned, was negligible. The slight increase in the phosphorus content of the last few hundredweights of metal from the ladle is attributable to the increase in the silica content of the slag when it reacted with the acid lining of the ladle bottom. The fireclay brick bottom slopes 3 in. down to the nozzle hole; therefore there is a period when the slag is in contact with the upper portion of the slope whilst some metal is still in the ladle covering the lower portion of the slope.

The analytical data relating to slag and metal samples taken on six casts are shown in Table XXIII. The remaining casts were not studied in detail, as routine analysis of the "pit" samples showed no evidence of rephosphorisation.

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To The United Steel Companies and to Leeds University the author wishes to express his gratitude for their kind permission to publish this paper. He is also greatly indebted to the management, staffs and operatives of the Steel, Peech and Tozer Works and the Central Research Department of The United Steel Companies, Ltd., for their co-operation and guidance both in the original investigation and in the preparation of this paper.

TABLE XXIII.—*Analytical Data of Slag and Metal Samples—Basic-Lined Ladle Experiments.*

Cast No.	Slag Sampling Period.	Slag Analysis.								Metal Sampling Period (in terms of cwt. of metal left in ladle).	Metal Analysis.					Bath Temp. on Tapping. ° C.	Ladle Additions.	
		SiO_2. %.	Fe. %.	CaO. %.	P_2O_5. %.	MnO. %.	MgO. %.	FeO. %.	Fe_2O_3. %.		C. %.	Si. %.	S. %.	P. %.	Mn. %.		Ferro-Manganese. Cwt.	Ferro-Silicon. Cwt.
36/6375†	Tapping	18·10	7·7	50·40	5·50	5·10	5·15	7·61	2·57	1640*	0·37	0·016	0·037	0·034	0·32	1612	11	7·5
										1190 (1st plate)	0·50	0·168	0·039	0·038	0·77			
										830 (2nd plate)	0·49	0·158	0·039	0·036	0·76			
										470 (3rd plate)	0·49	0·148	0·039	0·039	0·76			
										110 (4th plate)	0·48	0·140	0·037	0·039	0·76			
										35 (spare ingot)	0·45	0·117	0·036	0·047	0·77			
	End of casting	22·70	4·5	48·72	4·63	6·71	4·43	5·16	0·71	Nil	0·45	0·084	0·035	0·056	0·75			
40/2505	Tapping	15·21	9·3	50·00	6·97	5·22	3·89	7·61	4·72	1640*	0·83	0·011	0·022	0·034	0·29	1572	9	7·5
										1232 (1st plate)	0·83	0·158	0·024	0·032	0·67			
										782 (2nd plate)	0·83	0·160	0·025	0·035	0·66			
										400 (3rd plate)	...	0·151	0·024	0·035	0·65			
										93 (4th plate)	$0{\cdot}80_5$	0·145	0·025	0·037	0·67			
	End of casting	(Sample unsound.)								Nil	0·80	0·114	0·024	0·046	0·66			
35/7922‡	Tapping	11·10	14·8	52·90	4·70	5·17	4·34	14·83	4·72	1570*	0·28	...	0·029	0·014	0·16	1616	5·5 10·0 added in bath	8·5
										1315 (1st plate)	0·33	0·214	0·028	0·016	0·74			
										989 (2nd plate)	0·33	0·207	0·027	0·017	0·75			
										659 (3rd plate)	...	0·205	0·027	0·018	0·75			
										329 (4th plate)	...	0·196	0·028	0·018	0·75			
										50 (spare ingot)	0·31	0·164	0·029	0·020	0·75			
	End of casting	15·70	12·2	49·2	4·42	6·86	4·96	12·0	4·15	Nil	0·29	0·130	0·030	0·020	0·67			

34/3087	Tapping	14·30	8·3	54·6	6·27	4·82	3·98	7·71	3·15	1560*	0·53	...	0·033	0·035	0·20	1586	10·75	7·25
											(Sampled 20 min. before tapping.)							
										780 (2nd plate)	0·52	...	0·028	0·031	0·74			
										390 (3rd plate)	0·53	0·161	0·028	0·033	0·72			
										65 (4th plate)	...	...	0·029	0·033	0·72			
	End of casting	(Sample unsound.)								Nil	(Not sampled.)							
31/1367 §	Tapping	7·82	15·1	41·5	6·48	14·94	5·40	14·96	5·05	1590*	...	...	0·030	0·022	0·32	1602	8·5	6·25
										1326 (1st plate)	0·26	...	0·028	0·037	1·23		15·0	
										902 (2nd plate)	0·26	0·119	0·026	0·037	1·23		added	
										464 (3rd plate)	...	...	0·026	0·038	1·20		in bath	
										2 (4th plate)	0·26	0·094	0·027	0·043	1·17			
	End of casting	14·00	13·4	39·8	6·06	12·68	4·07	13·35	4·36	Nil	0·25	0·090	0·028	0·056	1·09			
29/81 ‖	Tapping	10·80	11·79	49·0	9·28	5·51	5·04	9·80	5·10	1558*	...	...	...	0·026	0·25	1590	11·0	6·25
										1294 (1st plate)	0·27	0·117	...	0·037	1·12		15·0	
										936 (2nd plate)	0·29	0·114	0·034	0·040	1·13		added	
										505 (3rd plate)	...	0·112	0·035	0·042	1·12		in bath	
										25 (spare ingot)	0·28	0·104	...	0·044	1·12			
	End of casting	21·80	7·00	40·46	6·16	9·48	5·58	6·70	2·57	Nil	...	...	...	0·081	0·86			

* Samples taken in the furnace on tapping. On casts which had part of the ferro-manganese added in the bath, the tapping samples were taken immediately before furnace deoxidation.

† Cast designed to tap at a high temperature and with a high silica content in the slag.

‡ Cast designed to tap at a high temperature and with a high iron content in the slag.

§ Cast designed to tap with a slag containing low silica and high manganous oxide.

‖ A top course of fireclay brick, held in position by means of brick-settting ganister, placed over the dolomite lining of the ladle to restore its capacity, which was reduced by the slag and metal growth on the dolomite lining.

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A STUDY OF THE BASIC OPEN-HEARTH PROCESS, WITH PARTICULAR REFERENCE TO SLAG CONSTITUTION.*

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(Figs. 10 to 15 = Plate L.)

Synopsis.

An account is given of experiments carried out to investigate the constitution of basic open-hearth slags in relation to the mechanism of dephosphorisation in the furnace and rephosphorisation in the casting pit.

The work described involved the X-ray examination and chemical analysis of slag samples taken during the refining period and the casting of the ingots. Differences in constitution were found, depending on the rate of cooling of the slag samples. Changes in bath composition have been studied by means of metal samples taken during the same period. The X-ray work has been supplemented by the examination of oxide mixtures prepared in the laboratory, the results of which have greatly contributed to the interpretation of the slag constitution. A hypothesis has been evolved to relate rephosphorisation in the pit to the composition of the tapping slag. A suggestion is made to explain the mechanism of dephosphorisation in the furnace.

Our first X-ray examination was made of some slag-pit lumps intended for crushed road-making material; these showed undesirable qualities—low strength and/or high porosity—and the X-ray photographs of them revealed striking differences from those of slags of normal quality as regards both mineralogical constitution and degree of crystallinity. The results were sufficiently interesting to encourage a study of furnace slags, and spoon-cooled basic slags were found to contain lime-rich and ferrous-oxide-rich phases, a calcium silicate, and in a few cases a small amount of dicalcium ferrite; there were also other constituents present which at first could not be identified. In spite of its rather low degree of crystallinity, the calcium silicate was believed to be the β-dicalcium silicate, β-$2CaO.SiO_2$, and not the tri-calcium silicate, $3CaO.SiO_2$.

Subsequently Dr. Zea and his colleagues at the works of Messrs. Steel, Peech and Tozer requested an X-ray examination of slags they had taken for work on rephosphorisation.[1] The results of the X-ray work were examined in relation to rephosphorisation data and a hypothesis was put forward relating rephosphorisation in the casting pit to the composition of the tapping slag. In view of the new approach to the problem, it was, perhaps, not surprising to find that this hypothesis was at variance in some respects with the fundamental assumptions made by Schenck in his interpretation of slag–metal reactions and used by Zea in the application of Schenck's theory to basic steelmaking.[2]

Later a study was made at the works of Messrs. Samuel Fox and Company, Ltd., of basic slags obtained during the working of the charge from the clear-melted stage to tapping and supplemented in some cases by ladle

* Received February 15, 1945.

slags. This examination has provided information on the dephosphorisation of the metal and confirmed previous deductions regarding slag/metal reactions.

In assessing the merits of this contribution in relation to Zea's paper, it should be remembered that the problem has been approached from a different point of view from his, and each paper should therefore be regarded as a contribution to our knowledge on a problem which still warrants considerable study.

Experimental Procedure.

The aim of the present work was the determination of slag constitution in an attempt to throw some light on the problem of rephosphorisation in pit-metal samples. Slag and metal samples were taken from the furnace at definite times during the refining period; metal samples were also taken during the casting of the ingots and a final slag sample was obtained by drainage through the ladle nozzle. Slag samples were taken from the furnace in a hot spoon and either allowed to cool in the spoon for 10 min. before chilling on the stage floor, or quickly poured into cold water.

Casts at Messrs. Steel, Peech and Tozer's works were sampled by Dr. Zea, and the slags were allowed to cool in the spoon before transference to the stage floor.

The taking of samples of rapidly quenched furnace slags was a development arising from the examination of the X-ray results on Dr. Zea's samples. The first tests of this kind were made at the works of Messrs. Samuel Fox and Company, Ltd., on slag samples taken about 15 min. before tapping. Details of the water-quenching experiments are given in Appendix IV.

Subsequently at this works, one metal and two slag samples, water-quenched and spoon-cooled, were taken at selected times during the refining period; samples of the pit metal and the final ladle slag were also obtained.

Analyses were provided by the Chemical Laboratories of Messrs. Steel, Peech and Tozer and the Central Research Department of The United Steel Companies, Ltd.

While certain phases could easily be recognised on the X-ray photographs of the basic slags, it was found that many of the reflections could not be accounted for by reference to well-known standards. (One list has been compiled recently by the American Society for Testing Materials.*) It therefore became clear that new standards would have to be investigated—standards chosen from a knowledge of the chemical analysis of the slags. This work was carried out during the early stages of the examination of basic slags and involved the making of oxide mixtures in varying proportions and heating them under a controlled atmosphere at selected temperatures. From this fundamental study it has been possible to obtain phases which fully account for all the lines on the X-ray photographs of basic slags. The experimental procedure and results are given in detail in Appendix I. Laboratory work also included experiments arranged to simulate the filling of the ladle by metal and slag, since it appeared, from data on certain casts, that a return of phosphorus, say up to 0·010%, could take place in the comparatively short time between taking the final furnace sample and first pit sample. This experiment is described in Appendix III.

The chemical analyses of metal and slag samples from casts made at Messrs. Steel, Peech and Tozer's works are given in Tables I., II. and III. A selection of X-ray data on slag samples is given in Table IV.

* "X-Ray Diffraction Data Cards."

Cast.	Sample.	Metal Analysis. %.					Slag Analysis. %.							
		C.	Mn.	Si.	S.	P.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.
25/1518	Before manganese.	0·58	0·21	...	0·019	0·023	51·3	8·4	7·1	6·7	4·9	10·6	5·2	11·8
	Tapping.	0·58	0·40	0·005	0·017	0·028	51·0	7·8	6·9	7·9	5·5	10·3	5·3	11·7
	1st pit.	0·58	0·70	0·26	0·028	0·027								
	2nd pit.	0·59	0·73	0·24	0·027	0·033								
	3rd pit.	0·59	0·70	0·24	0·026	0·039								
	4th pit.	0·59	0·71	0·23	0·025	0·051	*Final ladle slag.*							
	Spare ingot.	0·59	0·70	0·22	0·027	0·065	42·4	26·1	3·8	8·8	5·6	5·5	0·3	4·5
25/1521	Before manganese.	0·36	0·17	...	0·026	0·016	51·7	10·0	5·4	6·7	6·3	12·0	3·6	11·8
	Tapping.	0·35	0·39	0·00$_5$	0·023	0·018	51·7	10·4	5·4	6·7	7·0	11·0	5·1	12·1
	1st pit.	0·47	0·88	0·22	0·033	0·022								
	2nd pit.	0·46	0·84	0·22	0·034	0·022								
	3rd pit.	0·47	0·82	0·22	0·030	0·023								
	4th pit.	0·46	0·72	0·17	0·031	0·023	*Final ladle slag.*							
	Last metal.	0·44	0·46	0·17	0·035	0·037	43·6	20·2	4·3	7·5	6·0	11·2	0·6	9·1
23/3736	Before manganese.	0·55	0·19	...	0·024	0·021	54·7	9·6	8·6	5·5	6·5	10·1	4·2	10·7
	Tapping.	0·57	0·42	0·00$_5$	0·023	0·022	54·8	9·9	8·4	5·7	6·4	9·9	4·0	10·5
	1st pit.	0·63	0·84	0·29	0·031	0·030								
	2nd pit.	0·61	0·79	0·28	0·030	0·032								
	3rd pit.	0·63	0·79	0·27	0·026	0·032								
	4th pit.	0·61	0·73	0·24	0·029	0·032	*Final ladle slag.*							
	Spare ingot.	0·63	0·68	0·21	0·046	0·037	45·7	19·4	8·3	8·2	6·1	8·1	1·0	7·0
25/1527	Before manganese.	0·54	0·26	...	0·025	0·029	51·5	11·3	8·5	6·0	6·3	8·5	3·7	9·2
	Tapping.	0·55	0·48	...	0·024	0·039	50·8	11·4	8·5	6·5	6·4	7·9	4 1	9·0
	1st pit.	0·55	0·78	0·23	0·031	0·047								
	2nd pit.	0·56	0·76	0·23	0·030	0·051								
	3rd pit.	0·57	0·74	0·22	0·029	0·064								
	4th pit.	0·56	0·60	0·15	0·032	0·086	*Final ladle slag.*							
	Last metal.	0·54	0·52	0·12	0·034	0·090	45·5	23·9	6·5	7·1	5·2	6·1	0·4	5·0
25/1528	Before manganese.	0·63	0·21	0·01	0·038	0·018	50·0	15·0	6·9	6·2	6·3	8·6	4·6	9·9
	Tapping.	0·66	0·86	0·00$_5$	0·031	0·018	49·8	14·5	6·9	6·0	6·2	9·0	3·3	9·3
	1st pit.	0·62	0·85	0·23	0·033	0·028								
	2nd pit.	0·61	0·86	0·23	0·035	0·029								
	3rd pit.	0·61	0·86	0·22	0·034	0·029								
	4th pit.	0·60	0·83	0·21	0·034	0·029	*Final ladle slag.*							
	Last metal.	0·53	0·33	0·05	0·058	0·056	45·6	23·1	5·4	7·4	5·0	6·8	1·0	6·0
22/2309	Before manganese.	0·53	0·21	0·00$_5$	0·032	0·022	52·6	9·6	6·7	6·1	6·5	11·0	4·4	11·6
	Tapping.	0·56	0·58	0·00	0·031	0·027	52·6	9·8	6·5	6·7	6·9	10·0	5·0	11·2
	1st pit.	0·65	0·75	0·25	0·031	0·031								
	2nd pit.	0·63	0·74	0·25	0·031	0·032								
	3rd pit.	0·64	0·73	0·21	0·031	0·036								
	4th pit.	0·62	0·66	0·17	0·034	0·042	*Final ladle slag.*							
	Last metal.	0·60	0·43	0·17	0·047	0·059	39·5	25·3	5·1	7·8	3·7	...	...	6·1

TABLE II.—*Analytical Data for Metal and Slag Samples.*

Cast.	Sample.	Metal Analysis. %.					Slag Analysis. %.							
		C.	Mn.	Si.	S.	P.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.
34/2238	Before manganese.	0·13	0·17	...	0·045	0·020	51·7	8·5	5·0	5·5	6·5	12·8	5·1	13·5
	Tapping.	0·14	0·62	0·00$_5$	0·041	0·018	49·9	8·4	4·6	6·0	6·1	13·5	6·1	14·8
	2nd pit.	0·24	1·06	0·21	0·037	0·026								
	3rd pit.	...	1·06	...	0·039	0·031	*Final ladle slag.*							
	Last metal.	0·24	0·90	0·12	0·045	0·034	42·8	22·4	2·6	9·6	4·7	8·0	3·8	8·9
30/4600	Before manganese.	0·31	0·12	0·01$_5$	0·035	0·025	51·2	10·2	8·5	4·3	3·8	11·9	5·0	12·7
	Tapping.	0·22	0·46	0·00$_5$	0·038	0·020	50·7	9·5	7·7	4·6	4·0	12·6	2·6	11·6
	2nd pit.	0·33	1·16	0·21	0·033	0·026								
	4th pit.	...	1·12	...	0·033	0·028	*Final ladle slag.*							
	Last metal.	0·30	0·84	0·10	0·041	0·046	46·3	18·2	6·5	6·9	3·6	9·2	1·9	8·4
31/501	Before manganese.	0·15	0·17	0·00$_5$	0·035	0·018	52·4	12·3	5·5	4·3	4·0	11·7	3·7	11·7
	Tapping.	0·17	0·56	0·00$_5$	0·036	0·021	52·8	12·1	5·5	5·1	3·6	11·4	4·4	11·9
	2nd pit.	0·26	1·06	0·17	0·033	0·023								
	4th pit.	...	1·05	...	0·035	0·025	*Final ladle slag.*							
	Last metal.	0·26	0·71	0·07	0·047	0·040	46·0	18·8	4·5	6·5	3·3	8·8	3·6	9·3
31/508	Before manganese.	0·19	0·17	...	0·046	0·012	50·7	12·6	3·4	5·2	6·7	12·5	3·7	12·3
	Tapping.	0·19	0·54	0·00$_5$	0·041	0·016	50·5	12·5	3·5	5·0	6·6	12·0	4·7	12·6
	2nd pit.	0·26	1·02	0·19	0·046	0·024								
	3rd pit.	...	0·98	...	0·046	0·026	*Final ladle slag.*							
	Last metal.	0·28	0·89	0·14	0·046	0·030	45·6	22·7	3·2	8·6	3·8	8·9	1·6	8·0
35/7148	Before manganese.	0·26	0·19	0·00$_5$	0·034	0·024	51·5	10·8	7·6	3·2	4·9	10·4	5·6	12·0
	Tapping.	0·27	1·41	0·00$_5$	0·036	0·027	50·0	10·5	6·6	5·2	6·8	11·0	5·6	12·4
	2nd pit.	0·26	1·04	0·20	0·039	0·039								
	3rd pit.	...	1·00	...	0·039	0·041	*Final ladle slag.*							
	Last metal.	0·24	0·92	0·13	0·036	0·042	47·0	21·5	5·3	8·9	5·2	8·3	3·3	8·7
35/7156	Before manganese.	0·14	0·20	0·07$_5$	0·031	0·030	51·0	11·8	5·5	5·5	4·3	14·4	5·7	15·2
	Tapping.	0·11	0·24	0·01$_5$	0·027	0·016	51·0	10·2	4·6	6·3	5·2	15·6	5·6	16·0
	2nd pit.	0·28	1·10	0·20	0·035	0·020								
	4th pit.	...	1·04	...	0·037	0·020	*Final ladle slag.*							
	Last metal.	0·28	0·89	0·11	0·034	0·032	44·5	19·4	3·0	8·8	4·9	11·0	4·1	11·4
31/599	Before manganese.	0·13	0·13	0·01$_5$	0·025	0·023	49·6	9·6	5·7	5·8	4·2	14·2	4·9	14·4
	Tapping.	0·13	0·45	...	0·026	0·011	46·5	9·3	5·3	6·3	6·0	13·8	7·9	16·2
	2nd pit.	0·27	1·02	0·19	0·029	0·019								
	4th pit.	...	1·01	...	0·030	0·032	*Final ladle slag.*							
	Last metal.	0·26	0·76	0·08	0·031	0·034	41·5	17·0	3·9	7·3	5·8	14·7	3·0	13·5
31/572	Before manganese.	0·15	0·16	0·00$_5$	0·032	0·024	53·2	11·2	5·8	5·7	4·9	10·1	4·4	10·9
	Tapping.	0·19	0·84	...	0·037	0·023	52·8	10·4	5·5	6·8	5·6	10·3	4·7	11·3
	2nd pit.	0·30	1·08	0·21	0·040	0·037								
	4th pit.	...	1·45	...	0·041	0·079	*Final ladle slag.*							
	Last metal.	0·36	1·64	0·15	0·036	0·087	47·5	16·9	4·8	7·3	3·8	8·8	3·1	9·0
30/4591	Before manganese.	0·24	0·27	0·01	0·046	0·031	50·0	12·8	9·2	5·7	5·1	8·6	6·8	11·5
	Tapping.	0·27	0·87	0·01	0·042	0·042	50·4	12·5	8·8	5·7	6·0	...	...	12·0
	2nd pit.	0·30	1·12	0·16	0·043	0·048								
	4th pit.	...	1·11	...	0·043	0·057	*Final ladle slag.*							
	Last metal.	0·31	0·99	0·17	0·043	0·063	41·8	19·0	6·7	8·7	5·2	...	...	12·1

TABLE III.—*Analytical Data for Metal and Slag Samples.*

Cast.	Sample.	Metal Analysis. %.					Slag Analysis. %.							
		C.	Mn.	Si.	S.	P.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.
22/2418	Before manganese.	...	0·20	...	...	0·021	52·5	9·0	8·9	6·8	5·2	...	...	11·8
	Tapping.	...	...	...	...	0·021	52·4	8·8	8·6	6·2	5·2	...	...	12·0
	1st pit.	0·64	0·72	0·27	...	0·026								
	2nd pit.	0·63	0·73	0·27	0·27	0·028								
	3rd pit.	...	...	...	...	0·031								
	4th pit.	...	...	...	...	0·057								
22/2420	Before manganese.	...	0·18	...	...	0·014	52·4	8·6	8·0	6·2	4·7	...	...	11·0
	Tapping	...	...	...	...	0·020	52·2	8·3	7·7	6·8	4·8	...	...	10·9
	1st pit.	0·59	0·80	0·26	...	0·030								
	2nd pit.	0·59	0·80	0·26	0·024	0·030								
	3rd pit.	...	...	...	...	0·030								
	4th pit.	...	...	...	...	0·030								
22/2422	Before manganese.	...	0·16	...	...	0·012	52·7	9·6	6·9	6·0	4·4	...	...	12·3
	Tapping	...	...	...	...	0·015	52·6	9·8	6·8	6·5	4·8	...	...	12·5
	1st pit.	0·63	0·80	0·24	...	0·022								
	2nd pit.	0·63	0·78	0·24	0·031	0·022								
	3rd pit.	...	...	...	...	0·023								
	4th pit.	...	...	...	...	0·026								
22/2436	Before manganese.	...	0·17	...	...	0·019	54·3	10·7	7·4	5·2	6·6	...	...	10·8
	Tapping.	...	...	...	...	0·022	53·8	10·5	7·0	5·7	6·7	...	...	10·5
	1st pit.	0·59	0·88	0·30	...	0·026								
	2nd pit.	0·59	0·87	...	0·027	0·028								
	3rd pit.	...	...	...	...	0·033								

TABLE IV.—*Mineralogical Constitution of Spoon-Cooled Slags. X-Ray Results.*

Sample No.	Phase.	Amount.	Spacing kX. Units.
25/1518	CaO S.S.	Medium/large	4·730 ± 0·002
	FeO S.S.	Small/medium	4·328 ± 0·003
	Silico-fluorapatite	Medium/large	...
25/1521	CaO S.S.	Medium/large	4·739 ± 0·002
	FeO S.S.	Medium	4·330 ± 0·005
	β-2CaO.SiO$_2$	Medium	...
	Silico-fluorapatite	Small	...
25/1527	CaO S.S.	Medium	4·715 ± 0·003
	FeO S.S.	Small/medium	4·305 ± 0·005
	β-2CaO.SiO$_2$	Medium	...
	Silico-fluorapatite	Medium	...
25/1528	CaO S.S.	Small	4·720 ± 0·005
	FeO S.S.	Medium	4·320 ± 0·003
	β-2CaO.SiO$_2$	Medium/large	...
	Silico-fluorapatite	Small	...
22/2418	CaO S.S.	Medium	4·735 ± 0·002
	FeO S.S.	Small/medium	4·340 ± 0·005
	β-2CaO.SiO$_2$	Small	...
	Silico-fluorapatite	Medium	...
22/2420	CaO S.S.	Medium	4·741 ± 0·002
	FeO S.S.	Medium	4·334 ± 0·005
	β-2CaO.SiO$_2$	?	...
	Silico-fluorapatite	Medium	...
22/2422	CaO S.S.	Medium	4·740 ± 0·002
	FeO S.S.	Medium	4·342 ± 0·005
	β-2CaO.SiO$_2$	Small	...
	Silico-fluorapatite	Small/medium	...
22/2436	CaO S.S.	Medium	4·738 ± 0·002
	FeO S.S.	Small/medium	4·32 ± 0·005
	β-2CaO.SiO$_2$	Medium	...
	Silico-fluorapatite	Small	...

The data on casts at the works of Messrs. Samuel Fox and Company, Ltd. (*see* Tables V. to IX.) include details of the metal charge, additions to furnace and ladle, refining procedure, and analyses of metal and slag samples.

SUMMARY OF X-RAY RESULTS.

Constitution of Spoon-Cooled Furnace Slags.
(Messrs. Steel, Peech and Tozer.)

X-ray results on slag samples are given for the respective charges in Tables V. to IX. The X-ray findings indicated the presence of the following substances:

(1) A lime-rich phase in small to medium/large amount,* believed to contain manganous oxide in solid solution.

(2) A ferrous-oxide-rich phase in small to medium amount, believed to contain manganous oxide and magnesia in solid solution.

* Very small = <10 weight-%.
Small = 10–15 „
Medium = 20–25 „
Large = > 30 „

(3) β-Dicalcium silicate in very small to medium/large amount.

(4) A lime–silica phosphate (based on a range of compositions between $2CaO.SiO_2$ and $3CaO.P_2O_5$), and found by Nagelschmidt[3] by the use of petrographic methods.

(5) An apatite-like mineral containing silica, here regarded as a silica fluorapatite, in variable (small to medium) amount.

Constitution of Spoon-Cooled Ladle Slags.
(Messrs. Steel, Peech and Tozer).

β-Dicalcium silicate and a ferrous-oxide-type phase predominated.

Constitution of Rapidly Water-Quenched Slags Taken Just Before Tapping: Preliminary Study. (Messrs. Samuel Fox and Company, Ltd.)

The photographs of these slags showed the following crystalline phases :

(1) A lime-rich phase.

(2) The lime–silica phosphate nagelschmidtite, together with a non-crystalline component indicative of a supercooled liquid.

It was concluded that the two crystalline phases were present in solid form in the molten furnace slag.

(3) Another crystalline phase of the (FeO, MgO, MnO) type, which appeared when the quenching operation was carried out more slowly.

Constitution of Water-Quenched and Spoon-Cooled Slags Taken during the Refining Period: Second Study. (Messrs. Samuel Fox and Company, Ltd.)

All the water-quenched slags showed the presence of the lime–silica phosphate nagelschmidtite, which is present as a finely dispersed solid phase in the molten furnace slag. The phase appears before the slag is "truly basic," *i.e.* before free lime is present. The poor crystallinity of the ferrous-oxide-type phase suggested that this component was not present in this form, *i.e.* as a separate phase, in the molten slag. With the appearance of the free-lime phase, also regarded as being present in a dispersed solid form, the ferrous-oxide-type phase was very small in amount and indefinite in character.

The spoon-cooled slags confirmed the deductions made previously on slags from Messrs. Steel, Peech and Tozer's works.

Comments on Metal and Slag Analyses during the Refining Period. (Messrs. Samuel Fox and Company, Ltd.)

Detailed observations on the variations in slag composition, as shown by the results in Tables V. to IX., are outside the scope of this paper; it may be said, however, that they followed the expected course. The salient features can be outlined as follows : The lime and silica were normally low and high respectively at the clear-melted stage, the former increasing and the latter decreasing as refining proceeded. The iron content increased during the working of the charge, but showed marked variations in the clear-melted sample. These variations appeared to be associated with the content of residual phosphorus in the metal when clear-melted; this was high when the iron content was low, but was reduced to less than 0·04% phosphorus when the iron content approached 5%. The contents of manganous oxide and phosphorus pentoxide decreased from "at melted" owing to dilution of the bath by the increasing bulk of slag. Magnesia normally decreased from melting to tapping for the above reason, but may occasionally have increased when a hearth boil occurred.

TABLE V.—*Cast History of Cast No. V. 4887, with Analyses of Metal and Slag Samples.*

Type of Steel: 0·3% C, 1% Mn, 0·9% Cr, 0·05% S and P.
Charge: Basic Iron 30·9 tons, Steel Scrap 53·6 tons, Lime 3·3 tons.

Cast History.																						
Time After Melting.		Additions.					Samples No.	Metal Analysis. %.							Slag Analysis. %.								
Hr.	Min.	Addition.	Ton.	Cwt.	Qrt.	Lb.		C.	Mn.	Si.	S.	P.	Ni.	Cr.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.	F_2.
...		Spar	...	15	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
0	0	*Clear melted*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
0	5	...	...	...	...	...	1	0·63	0·27	tr.	0·055	0·040	0·17	0·12	47·6	15·3	10·1	10·5	4·6	5·9	3·0	6·7	...
0	15	Lime	1	13	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
0	35	Scale	...	5	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	0	...	...	...	...	...	2	0·48	0·30	tr.	0·029	0·023	0·17	0·12	53·6	12·9	7·5	7·6	4·3	6·4	4·1	7·9	...
2	0	...	...	...	...	...	3	0·19	0·30	tr.	0·028	0·035	$0·17_5$	0·05	51·4	12·4	7·2	6·0	6·6	8·1	4·8	9·7	...
2	5	Lime	1	13	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	10	Spiegel	1	4	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	15	...	...	...	...	...	4 (slag only)	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	20	Fe–Cr	1	7	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	35	Fe–Mn	...	4	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	40	...	...	...	...	...	5 (slag only)	...	...	...	...	...	...	...	54·0	12·5	6·0	6·6	6·0	7·0	4·9	8·9	...
2	45	*Tapped*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
...		*Ladle*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
...		Fe–Mn	...	2	0	0	Pit No. 1	0·34	1·08	0·22	0·027	0·040	$0·17_5$	0·88	...	...	...	...	...	...	...	...	...
...		Si–Mn	...	6	3	4	2	...	...	...	...	0·044	...	...	...	...	...	...	...	...	...	...	...
...		Fe–Si	...	2	3	0	3	$0·34_5$	1·07	$0·21_5$	0·030	0·045	...	...	...	...	...	...	...	...	...	...	...

X-Ray Results. Mineralogical Constitution of Slags: Amount, Crystallinity, Spacing (kX. Units).

Sample No.		Water-Quenched.					Spoon-Cooled.				
		CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).	CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).
1	Amount	...	Medium	Large	Small	?	...	Medium/ large	...	Medium	Medium
	Crystallinity	...	Moderate	Good	...	...	...	Good	...	Poor	...
	Spacing	...	4·313 ± 0·001	...	...	...	...	4·319 ± 0·001	...	...	...
2	Amount	Very small	Small	Large	Small	...	...	...	...	...	...
	Crystallinity	...	Poor	Good	...	...	...	...	...	...	...
	Spacing	...	4·31 ± 0·005	...	...	...	...	...	...	...	...
3	Amount	...	...	...	...	...	Medium	Medium	...	Small	Medium/ large
	Crystallinity	...	...	...	...	...	Moderate	Poor	...	...	...
	Spacing	...	...	...	...	...	4·745	4·27–4·30 ± 0·005	...	...	...
4	Amount	Medium	?	Large	Small	...	Medium/ large	Small/ medium	...	Small	Medium/ large
	Crystallinity	Good	...	Good	...	...	Moderate	Poor	...	...	...
	Spacing	4·687 ± 0·001	...	...	...	...	...	...	...	...	...
5	Amount	Medium/ large	...	Large	Small	...	Medium/ large	Medium	...	Medium	Medium/ large
	Crystallinity	Good	...	Good	...	...	Moderate	Poor	...	Normal	...
	Spacing	4·692 ± 0·001	...	...	...	...	4·719 ± 0·002	4·297 ± 0·005	...	...	...
Ladle	Amount	...	...	...	...	...	...	Medium/ large	...	Medium/ large	...
	Crystallinity	...	...	...	...	...	...	Poor	...	...	...
	Spacing	...	...	...	...	...	...	4·31 ± 0·005	...	...	...

Note.—(C,S,P) = Lime–silica phosphate compound nagelschmidtite.
β-2C.S. = β-2CaO.SiO$_2$.
(S-FA) = Silico-fluorapatite.

TABLE VI.—*Cast History of Cast No. Z. 3982, with Analyses of Metal and Slag Samples.*

Type of Steel: 0·4% C, 0·8% Mn, 1% Cr, 0·05% S and P.

Charge: Basic Iron 24·7 tons, Steel Scrap 53·0 tons, Lime 4·4 tons.

Cast History.																							
Time After Melting.		Additions.					Samples No.	Metal Analysis. %.							Slag Analysis. %.								
Hr.	Min.	Addition.	Ton.	Cwt.	Qrt.	Lb.		C.	Mn.	Si.	S.	P.	Ni.	Cr.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.	F_2.
...		Spar	...	15	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
	0	*Clear melted*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
							1	0·63	0·18	tr.	0·056	0·027	$0·24_5$	0·24	49·3	14·0	9·4	7·7	6·2	6·4	3·2	7·2	...
	10	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
	40	Lime	2	4	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
							2	0·40	0·24	tr.	0·041	0·033	0·24	0·21	55·3	12·3	8·1	5·2	5·8	4·4	4·1	6·3	...
	55	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	0	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	5	Scale	...	7	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
							3	0·37	0·21	tr.	0·033	0·027	0·24	0·18	54·9	11·7	7·6	4·6	5·6	7·2	4·2	8·5	...
1	20	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	25	Lime	1	13	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	30	Spiegel	1	4	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	50	Silica bricks	...	...	2	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1	55	Silica bricks	...	...	2	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
							4 (slag only)	...	...	...	...	...	...	...	54·2	12·1	6·4	4·2	5·4	7·7	5·4	9·8	...
2	13	...	...	...	...	...		...	...	...	...	...	...	...									
2	15	Fe–Cr	1	4	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
							5 (slag only)	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	25	...	...	...	...	...																	
2	30	Fe–Mn	...	6	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2	35	*Tapped*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
...		*Ladle*	...	...	...	...	Pit No. 1	0·46	0·88	0·21	0·035	0·032	...	...	...	...	...	...	...	...	...	...	...
...		Fe–Mn	...	2	0	0	2	...	...	...	...	0·033	0·25	1·06	...	...	...	...	...	...	...	...	...
...		Fe–Si	...	2	3	0	3	0·46	0·87	$0·19_5$	0·035	0·033	...	...	...	...	...	...	...	...	...	...	...
...		Si–Mn	...	6	3	0	4	$0·45_5$	0·85	$0·19_5$	0·037	0·032	...	...	...	...	...	...	...	...	...	...	...
...		...	...	...	...	...	Ladle slag	...	...	...	...	...	...	...	50·8	17·3	4·7	5·7	4·7	7·6	1·8	7·2	...

X-Ray Results. Mineralogical Constitution of Slags: Amount, Crystallinity, Spacing (kX. Units).

Sample No.		Water-Quenched.					Spoon-Cooled.				
		CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).	CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).
1	Amount	...	...	...	...	...	...	Medium	Small	Medium	Medium/ large
	Crystallinity	...	...	...	...	...	...	Moderate	...	Normal	...
	Spacing	...	...	...	...	...	...	4·296 ± 0·001	...	...	...
2	Amount	Medium	Medium	Large	Small	...	...	...	...	...	...
	Crystallinity	Good	Very poor	Good	...	...	...	...	...	...	...
	Spacing	4·701 ± 0·0005	4·23* ± 0·01 4·28 ± 0·01	...	...	...	...	...	...	...	...
3	Amount	Medium/ large	?	Large	Small	...	Medium/ large	Medium	...	Medium	Medium/ large
	Crystallinity	Good	...	Good	...	...	Moderate	Poor	...	Normal	...
	Spacing	...	...	...	...	...	4·736 ± 0·001	4·255* ± 0·003	...	...	...
4	Amount	Medium/ large	Small	Large	Very small/ small	...	Medium/ large	Small/ medium	?	Medium/ large	Medium
	Crystallinity	Good	...	Good	...	...	Moderate	Poor	...	Normal	...
	Spacing	4·718 ± 0·0005	4·26* ± 0·01	...	...	...	...	...	...	...	...
5	Amount	...	...	...	...	...	Medium/ large	Small/ medium	?	Medium/ large	Medium
	Crystallinity	...	...	...	...	...	Moderate	Poor	...	Normal	...
	Spacing	...	...	...	...	...	4·750 ± 0·001	4·305 ± 0·005	...	...	...
Ladle	Amount	...	...	...	...	...	...	Medium	...	Large	...
	Crystallinity	...	...	...	...	...	...	Poor	...	Normal	...
	Spacing	...	...	...	...	...	...	4·295 ± 0·003	...	...	...

* Indication of high MgO and/or low FeO content.

Note.—(C,S,P) = Lime-silica phosphate compound nagelschmidtite.
β-2C.S. = β-$2CaO.SiO_2$.
(S-FA) = Silico-fluorapatite.

TABLE VII.—*Cast History of Cast No. V. 4921, with Analyses of Metal and Slag Samples.*

Type of Steel: 0·7% C, 0·3% Mn, 0·04% S and P.
Charge: Basic Iron 30·5 tons, Steel Scrap 46·0 tons, Lime 3·5 tons.

Cast History.

Time After Melting. Hr. Min.	Additions.					Samples No.	Metal Analysis. %.							Slag Analysis. %.								
	Addition.	Ton.	Cwt.	Qrt.	Lb.		C.	Mn.	Si.	S.	P.	Ni.	Cr.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.	F_2.
...	Spar	...	15	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
0	*Clear melted*	...	...	...	...	1	1·20	0·45	tr.	0·046	0·20	0·20	0·09	54·1	16·6	7·3	5·0	7·4	2·1	1·5	2·7	2·8
10	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
20	Lime	2	4	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
30	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
35	Scale	1	0	0	0	2	1·06	0·22	tr.	0·037	0·02	...	...	56·0	10·9	6·2	5·0	5·7	5·9	6·5	9·1	...
1 25	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1 50	Lime	...	9	0	0	3	0·96	0·19	tr.	0·036	0·02	...	...	...	...	...	...	...	...	...	...	...
2 10	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 20	Scale	...	...	4	0	4	0·79	0·16	tr.	0·035	0·016	...	...	...	...	...	...	...	...	...	...	...
2 40	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 45	Lime	...	9	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 50	Spiegel	...	12	0	0	5	0·79	0·17	tr.	0·034	0·017	0·17	0·08	55·0	11·3	6·1	4·5	5·0	9·1	5·9	11·2	1·4
3 15	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
3 20	Fe–Mn	...	4	2	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
3 25	*Tapped*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
...	*Ladle*	...	...	...	...	Pit No. 1	$0·73_5$	0·35	0·17	0·033	0·019	...	...	...	...	...	...	...	...	...	...	...
...	Fe–Si	...	5	3	14	2	...	0·34	...	...	0·019	...	...	...	...	...	...	...	...	...	...	...
...	...	...	...	...	...	3	0·73	0·33	$0·16_5$	0·035	0·020	0·17	0·11	...	...	...	...	...	...	...	...	...
...	...	...	...	...	...	4	...	0·32	...	...	0·021	...	...	...	...	...	...	...	...	...	...	...
...	...	...	...	...	...	5	$0·73_5$	0·30	0·11	0·041	0·022	...	...	...	...	...	...	...	...	...	...	...
...	...	...	...	...	...	Ladle slag	...	...	...	...	...	...	...	47·8	21·7	4·8	5·2	5·3	5·0	2·5	5·7	...

X-Ray Results. Mineralogical Constitution of Slags: Amount, Crystallinity, Spacing (kX. Units).

Sample No.		Water-Quenched.					Spoon-Cooled.				
		CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).	CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).
1	Amount	...	Small	Large	Small	...	...	Small	Small	Large	...
	Crystallinity	...	Moderate	Good	...	...	...	Very poor	...	Normal	...
	Spacing	...	4·288 ± 0·002	...	...	...	...	4·27* ± 0·01	...	...	...
2	Amount	Medium	Small	Medium/ large	?	...	Medium	Small	Small	Medium	Small
	Crystallinity	Very good	Poor	Good	...	...	Moderate	Poor	...	Normal	...
	Spacing	4·723 ± 0·0005	4·26* ± 0·005	...	...	...	4·744 ± 0·001	4·25* ± 0·005	...	...	...
4	Amount	Medium +	Small	Medium/ large	...	...	Medium +	Medium	?	Medium	Medium
	Crystallinity	Very good	Very poor	Good	...	...	Moderate	Poor	...	Normal	...
	Spacing	4·724 ± 0·0005	4·28 ± 0·01	...	...	...	4·753 ± 0·002	4·24* ± 4·28	...	...	...
5	Amount	Medium/ large	...	Medium/ large	...	...	Medium/ large	Medium	...	Small/ medium	Medium
	Crystallinity	Very good	...	...	...	...	Moderate	Poor	...	...	...
	Spacing	...	...	...	...	...	4·749 ± 0·002	4·255* ± 0·005	...	...	...
Ladle	Amount	...	...	...	...	...	...	Medium	?	Medium/ large	?
	Crystallinity	...	...	...	...	...	...	Moderate	...	Normal	...
	Spacing	...	...	...	...	...	...	4·294 ± 0·002	...	...	...

* Indication of high MgO and/or low FeO content.

Note.—(C,S,P) = Lime-silica phosphate compound nagelschmidtite.
β-2C.S. = β-2CaO.SiO_2.
(S-FA) = Silico-fluorapatite.

TABLE VIII.—*Cast History of Cast No. V. 4953, with Analyses of Metal and Slag Samples.*

Type of Steel: 0·3% C, 1·4% Mn, 0·05% S and P.

Charge: Basic Iron 19·4 tons, Steel Scrap 48·4 tons, Lime 3·8 tons.

Cast History.																								
Time After Melting.		Additions.					Samples No.	Metal Analysis. %.							Slag Analysis. %.									
Hr.	Min.	Addition.	Ton.	Cwt.	Qrt.	Lb.		C.	Mn.	Si.	S.	P.	Ni.	Cr.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.	F_2.	
...		Spar	...	12	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
	0	*Clear melted*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
...		Lime	1	13	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
	10	Scale	1	0	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
	20	Scale	...	5	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
	35	Scale	...	5	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
1	5	...	...	...	...	...	1	0·68	0·26	tr.	0·038	0·029	0·32	0·13	55·6	14·9	6·0	5·6	4·4	6·9	2·8	7·3	1·69	
1	15	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
1	20	Ore	...	5	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
1	45	...	...	...	...	...	2	0·47	0·24	tr.	0·034	0·028	...	...	...	...	...	...	...	...	...	...	...	
2	0	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
2	25	...	...	...	...	...	3	0·28	0·20	tr.	0·034	0·028	...	...	55·8	12·9	4·7	4·6	6·6	8·5	2·8	8·6	...	
2	30	Lime	1	2	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
2	35	Spiegel	...	12	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
3	0	Fe–Mn	...	15	2	14	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
3	10	...	...	...	...	...	4	0·32	1·18	tr.	0·033	0·024	0·32	0·13	55·0	11·8	4·0	5·0	7·3	8·5	4·7	9·9	...	
3	15	*Tapped*	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	
...		*Ladle*	...	...	...	...	Pit No. 1	0·31	1·38	$0·17_5$	0·036	0·027	...	...	...	...	...	...	...	...	...	...	...	
...		Fe–Mn	...	1	3	0	2	...	...	...	...	0·029	0·32	0·14	...	...	...	...	...	...	...	...	...	
...		Si–Mn	...	9	2	0	3	0·31	1·36	...	0·036	$0·028_5$	...	...	...	...	...	...	...	...	...	...	...	
...		Fe–Si	...	1	1	0	4	...	...	...	...	0·031	...	...	...	...	...	...	...	...	...	...	...	
...		...	...	...	...	...	5	0·31	1·33	$0·16_5$	0·037	0·034	...	...	...	...	...	...	...	...	...	...	...	
...		...	...	...	...	...	6	...	...	...	...	0·037	0·32	0·13	...	...	...	...	...	...	...	...	...	
...		...	...	...	...	...	7	0·30	1·30	0·15	0·035	0·040	...	...	...	...	...	...	...	...	...	...	...	
...		...	...	...	...	...	Ladle slag	...	...	...	...	...	...	...	50·8	16·4	3·7	5·4	6·0	7·8	3·6	8·6	...	

X-Ray Results. **Mineralogical Constitution of Slags:** Amount, Crystallinity, Spacing (kX. Units)

Sample No.		Spoon-Cooled.				
		CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA)
1	Amount	Small/medium	Medium	Small	Medium/large	Medium
	Crystallinity	Moderate	Poor	...	Normal	...
	Spacing	4·711 ± 0·001	(*a*) 4·307 ± 0·002 (*b*) 4·255* ± 0·005	...	...	...
3	Amount	Medium	Medium	Small/medium	Medium	?
	Crystallinity	Moderate	Moderate	...	Normal	...
	Spacing	4·745 ± 0·002	4·25* ± 0·003	...	...	...
4	Amount	Medium/large	Medium	Small	Medium	Small
	Crystallinity	Good	Moderate	...	Normal	...
	Spacing	4·742 ± 0·001	4·245* ± 0·002	...	...	...
Ladle	Amount	...	Medium/large	Small	Very large	...
	Crystallinity	...	Moderate	...	Normal	...
	Spacing	...	4·300 ± 0·002	...	...	...

* Indication of high MgO and/or low FeO content.

Note.—(C,S,P) = Lime-silica phosphate compound nagelschmidtite.
β-2C.S. = β-$2CaO.SiO_2$.
(S-FA) = Silico-fluorapatite.

TABLE IX.—*Cast History of Cast No. V. 4960, with Analyses of Metal and Slag Samples.*

Type of Steel: 0·8% C, 0·8% Mn, 0·04% S. and P.

Charge: Basic Iron 31·0 tons, Steel Scrap 44·5 tons, Lime 3·9 tons.

Cast History.																						
Time After Melting.	Additions.					Samples No.	Metal Analysis. %.							Slag Analysis. %.								
Hr. Min.	Addition.	Ton.	Cwt.	Qrt.	Lb.		C.	Mn.	Si.	S.	P.	Ni.	Cr.	CaO.	SiO_2.	P_2O_5.	MnO.	MgO.	FeO.	Fe_2O_3.	Fe.	F_2.
...	Spar	...	15	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
0	*Clear melted*	...	...	...	...	...	...	...	...	...	...	...	...	56·6	21·9	5·1	5·7	4·7	1·3	1·7	2·2	2·92
25	...	...	...	...	...	1	1·17	0·48	tr.	0·044	0·24	0·17	0·12	...	...	...	...	...	...	...	...	...
25	Lime	1	13	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1 0	...	...	...	...	...	2	1·12	0·50	tr.	0·029	0·18	...	...	...	...	...	...	...	...	...	...	...
1 0	Scale	1	0	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1 30	Lime	...	11	0	0	...	1·03	0·32	tr.	0·026	0·04	...	...	58·3	15·8	6·5	4·6	4·0	5·0	1·9	5·2	...
1 40	...	...	...	...	...		...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1 45	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
1 55	Scale	...	5	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 15	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 25	Scale	...	5	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 45	...	...	...	...	...	4	0·84	0·21	tr.	0·026	0·021	...	...	...	...	...	...	...	...	...	...	...
2 45	Lime	...	11	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
2 50	Spiegel	...	12	0	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
3 20	Fe–Mn	...	6	2	0	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
3 25	...	...	...	...	...	5	0·76	0·42	tr.	0·026	0·029	0·20	0·10	58·0	13·5	5·2	5·0	3·7	8·4	3·4	8·9	1·20
3 30	*Tapped*	...	...	...	...	...	...	...	...	...	...	...	...	...	...		...	...	...	...	...	...
...	*Ladle*	...	...	...	...	Pit No. 1	0·81	0·82	...	...	0·027	...	...	...	...	...	...	...	...	...	...	...
...	Fe–Mn	...	2	0	0	2	0·80	0·82	0·15	0·029	0·028	0·19	0·12	...	...	...	...	...	...	...	...	...
...	Si–Mn	...	6	2	0	3	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...	...
...	Fe–Si	...	2	3	0	4	0·81	0·77	0·14	...	0·033	...	...	...	...	...	...	...	...	...	...	...
...	...	...	...	...	...	5	...	0·66	...	...	0·047	...	...	...	...	...	...	...	...	...	...	...
...	...	...	...	...	...	Ladle slag	...	...	...	...	...	...	...	50·0	21·4	4·2	7·1	3·8	4·6	2·8	5·5	...

X-Ray Results. Mineralogical Constitution of Slags: Amount, Crystallinity, Spacing (kX. Units).

Sample No.		Spoon-Cooled.				
		CaO. S.S.	FeO. S.S.	(C,S,P).	β-2C.S.	(S-FA).
1	Amount	...	Medium	?	Large	...
	Crystallinity	...	Moderate	...	Normal	...
	Spacing	...	$4 \cdot 278 \pm 0 \cdot 001$	...	...	...
2a	Amount	Medium/large	Small	...	Medium	Medium
	Crystallinity	Moderate	...	...	Normal	...
	Spacing	$4 \cdot 762 \pm 0 \cdot 001$	4·26	...	...	...
2b Only partially fluxed	Amount	Large	...	...	$3CaO.SiO_2$/large	...
	Crystallinity	Good	...	...	Good	...
	Spacing	$4 \cdot 764 \pm 0 \cdot 0005$	...	...	...	...
4	Amount	Medium/large	Medium	Small	Medium	Small
	Crystallinity	Good	Moderate	...	Normal	...
	Spacing	4·756	$4 \cdot 311 \pm 0 \cdot 002$	...	...	...

Note.—(C,S,P) = Lime–silica phosphate compound nagelschmidtite.
β-2C.S. = β-$2CaO.SiO_2$.
(S-FA) = Silico-fluorapatite.

An Interpretation of Pit Rephosphorisation in Relation to Analyses of Tapping Slag. (Messrs. Steel, Peech and Tozer.)

In view of the fact that the six casts containing carbon 0·45–0·65%, manganese 0·7–0·85%, showed distinctive characteristics as regards the amount of rephosphorisation in the pit, the results from these were chosen for first study. The deductions were certainly encouraging, and an examination of the nine casts containing carbon 0·3%, manganese $1 \cdot 0_5$%, provided substantial agreement. A hypothesis was thus evolved which expressed the probability of rephosphorisation of the pit samples (between first and fourth pit) in terms of the analysis of the tapping slag, with particular reference to the ratio SiO_2/P_2O_5, and of the oxidising power of the slag, represented by the ratio slag-iron/MnO. (It should be noted that the above statement only applies when free lime—excess or uncombined lime—is present in the slag.)

The development of this hypothesis concerning the fixation of phosphorus in the slag and the factors controlling the tendency towards rephosphorisation in the pit was based on the following considerations. It seemed clear from the X-ray findings that a $2CaO.SiO_2$-type phase was formed in preference to $2FeO.SiO_2$ (fayalite), $2MnO.SiO_2$ (tephroite) or $MnO.SiO_2$ (rhodonite). This conclusion accorded well with the known relative stabilities at high temperatures. Striking variations in the silica contents of the slags were indicated by chemical analysis. Thus, as a starting point, it appeared reasonable to suppose that the ratio CaO/SiO_2 would be a most important factor. The presence of a ferrous-oxide-rich phase in the spoon-cooled slag suggested that the ferrous-oxide content of the tapping slag was a second important factor. The ratio of the weight percentage of CaO to that of SiO_2 is plotted against the weight percentage of FeO for each tapping slag in Fig. 1. The grouping of the points in

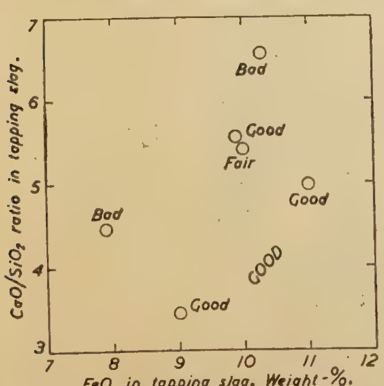

Fig. 1.—Relation Between CaO/SiO_2 Ratio and FeO Content in Tapping Slags.

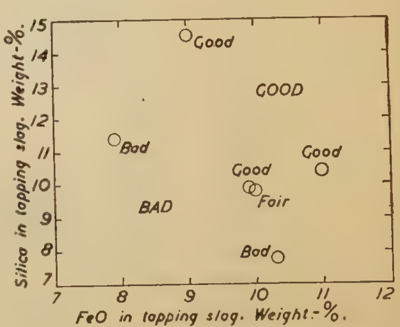

Fig. 2.—Relation Between SiO_2 and FeO Contents in Tapping Slags.

relation to the degree of rephosphorisation experienced in the pit indicates, within the range of slag compositions examined:

(1) that for a given percentage of FeO the slags which show rephosphorisation contain a higher CaO/SiO_2 ratio than those which show negligible rephosphorisation;

(2) that for a given CaO/SiO_2 ratio the slags which show rephosphorisation contain a lower percentage of FeO than those which show negligible rephosphorisation; and

(3) that other factors should be considered, as evidenced by the unsatisfactory position of one sample marked "good."

In view of the fact that the variation in the CaO/SiO_2 ratio is due largely to changes in the silica content, a plot of percentage of silica against percentage of ferrous oxide is shown in Fig. 2. The grouping of points indicates:

(1) that for a given FeO content the SiO_2 content is higher for the good casts than for the bad casts, and
(2) that for a given SiO_2 content the FeO content is higher for the good casts than for the bad casts.

It should be remembered that free lime is present in these tapping slags. The grouping of the points in Figs. 1 and 2 is satisfactory with the exception of samples 23/3736. In this connection, the chemical analysis shows a lower manganous oxide content, the effect of which has so far been neglected. It thus appears advisable to take into account the possible effect of the MnO content of the slag. Manganous oxide may be regarded as having a retarding effect on the oxidising power of the iron oxide for the following reasons:

(*a*) here it is not regarded as an oxidising agent,
(*b*) it is a possible diluent towards FeO, since MnO and FeO are mutually soluble, and
(*c*) the activity of FeO–MnO mixtures for increased MnO concentrations will doubtless decrease owing to the higher melting point.

It was therefore decided initially to represent the oxidising power of the slag by the ratio FeO/MnO. A plot of silica contents against the ratios FeO/MnO is shown in Fig. 3. The grouping, compared with that of Fig. 2, is considerably improved.

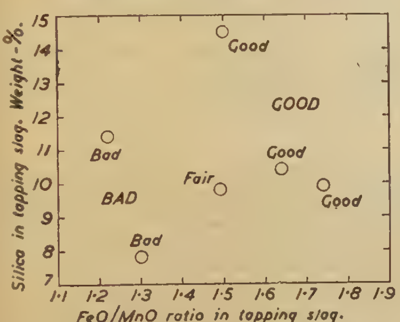

FIG. 3.—Relation Between SiO_2 Content and FeO/MnO Ratio in Tapping Slags.

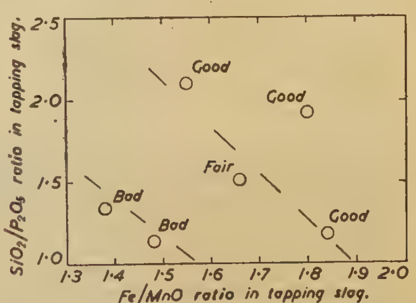

FIG. 4.—Relation Between SiO_2/P_2O_5 Ratio and Total-Fe/MnO Ratio in Tapping Slags from steels containing 0·45–0·65% of carbon and 0·7–0·85% of manganese.

In the foregoing graphical treatment, the content of P_2O_5 in the slag has been ignored, though it is fairly obvious that it will have an important bearing on the question of rephosphorisation. The analytical results show that, provided the slag contains an excess of lime, the presence of a high silica content is beneficial from the viewpoint of stabilising the phosphorus in the slag. Thus, it appeared reasonable to associate the fixation of the P_2O_5 in the slag with the amount of lime silicate, $2CaO.SiO_2$. In other words, a slag containing high silica, *i.e.* a high lime silicate content,

would appear capable of holding more P_2O_5 than one with a low silica content. Fig. 4 shows a graphical relationship between the values of the ratio SiO_2/P_2O_5 and the oxidising power of the slag as represented by the ratio slag-iron/MnO. This latter expression differs from that of Fig. 3, in that total iron is considered instead of ferrous oxide. This substitution does not unduly modify the position of the points on the graph and is more in keeping with normal laboratory methods of analysing slag samples.

The grouping was regarded as satisfactory. A certain amount of latitude must be granted in the placing of these points, since slight variations in chemical analysis often show an enhanced effect on the precise value of the ratio involved.

The results of the higher-manganese series (9 casts) are contained in Fig. 6 and show equally satisfactory agreement.

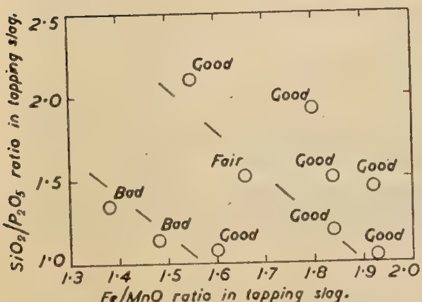

Fig. 5.—Relation Between SiO_2/P_2O_5 Ratio and Total-Fe/MnO Ratio in Tapping Slags from 10 casts of steel containing 0·45–0·65% of carbon and 0·7–0·85% of manganese.

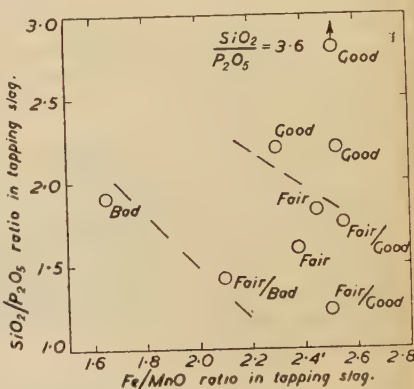

Fig. 6.—Relation Between SiO_2/P_2O_5 Ratio and Total-Fe/MnO Ratio in Tapping Slags from steels containing 0·25–0·30% of carbon and 1·05–1·10% of manganese.

Later, four casts of lower manganese content were examined and the results are given with those of the six similar casts in Fig. 5. This plot shows one obvious inconsistency, namely: Cast No. 22/2420, in which negligible rephosphorisation occurred, whereas, from the interpretation of the chemical analysis, marked rephosphorisation should have occurred. On the other hand, the constitution of this slag (*see* Table IV.) suggests a slag condition similar to that of slags 22/2418, and 22/2436, which showed negligible rephosphorisation.

A comparison of the grouping of points in Figs. 5 and 6 for medium- and high-manganese steels reveals some interesting facts (*see* Table X.).

Whatever the true significance of the ratios in the last column, it is clear that slags with higher Fe/MnO values are required for finished steels of higher manganese content in order to restrict rephosphorisation to negligible amounts.

The results of the examination of the slags taken at the works of Messrs. Samuel Fox and Company, Ltd., accord very satisfactorily with the considerations outlined above. It was perhaps unfortunate, from the point of view of this theory, that no outstanding case of rephosphorisation occurred in this series; nevertheless, the work has provided additional confirmation of the validity of the foregoing hypothesis.

TABLE X.

Type of Cast.	Ratio SiO_2/P_2O_5.	Ratio Total-Iron/Manganese-Oxide. Steel 1. 0·75–0·85% Mn.	Steel 2. 1·05–1·1% Mn.	Ratio $\frac{(Fe/MnO)_2}{(Fe/MnO)_1}$.
Bad	1·4	1·4	2·0	1·45
Fair	1·4	1·6/1·7	2·4/2·5	1 5
Good	2·0	1·6	2·4	1·5
Mean				1·5
Ratio of Mn % in finish steel				1·4
Ratio of Mn % added				1·5

THEORETICAL CONSIDERATIONS REGARDING THE MECHANISM OF SLAG FORMATION AND THE RETENTION OF PHOSPHORUS.

The presence of the phases lime, lime silicate and ferrous oxide in slowly cooled slags suggested initially a simple way of representing the formation of basic slag. The process may be outlined as below :

(1) $Si \rightarrow SiO_2$ $\quad Mn \rightarrow MnO \quad Fe \rightarrow FeO$.
(2) $FeO + MnO + SiO_2 \rightarrow (FeO,MnO,SiO_2)$ low-melting-point liquid.
(3) (FeO,MnO,SiO_2) liquid $+ CaO \rightarrow (FeO,MnO,CaO,SiO_2)$ liquid.
$(FeO,MnO,CaO,SiO_2) + CaO \rightarrow 2(CaO,MnO,FeO).SiO_2 + FeO$.
$2(CaO,MnO,FeO).SiO_2 + CaO \rightarrow 2CaO.SiO_2 + FeO + MnO$.
$2CaO.SiO_2 + FeO + MnO + CaO \rightarrow CaO + 2CaO.SiO_2 + (FeO,MnO)$.

The significance of the above scheme is that the addition of lime, strictly the fluxing of lime, displaces (sets free) ferrous and manganous oxides molecule for molecule from ferrous and manganous silicates. The following equations are therefore believed to be fundamental to basic steel-making :

$$2FeO.SiO_2 + 2CaO \rightarrow 2CaO.SiO_2 + 2FeO.$$
$$2MnO.SiO_2 + 2CaO \rightarrow 2CaO.SiO_2 + MnO.$$

This conception has been studied in the laboratory by means of synthetic oxide mixtures (*see* Appendix II.), the results of which show that the equations are correct. In other words, in the presence of free or excess lime, ferrous and manganous silicates are not stable and cease to exist.

The main features of the hypothesis already given, relating the probability of rephosphorisation of pit samples to the analyses of the tapping slag, can be expressed by the following points :

(1) For a given slag-iron/MnO ratio, an improvement from "bad" to "good" is effected by increasing the SiO_2/P_2O_5 ratio.

(2) For a given SiO_2/P_2O_5 ratio, an improvement is effected by increasing the slag-iron/MnO ratio. This means that a slag with a relatively high oxidising power must be maintained.

For the above theory to be regarded as a true statement and not merely as a working hypothesis, it seemed necessary that at least a portion of the P_2O_5 should be held as a solid lime–silica phosphate in the molten slag. This reasoning appeared to be further strengthened by the fact that, considering slags of the same total iron, manganous oxide and phosphorus pentoxide contents, a slag with a greater silica content retained phosphorus better than one with a lower silica content (free lime being, of course, always present). Thus it was inferred that molten basic slag is a two-phase system—solid and liquid—not unlike a suspension. Further

examination of the rephosphorisation process showed that this could be considered as taking place in three stages:

(1) Furnace to first pit sample.
(2) First pit sample to fourth pit sample.
(3) Fourth pit sample to last metal draining from the ladle.

It was observed that where the phosphorus content of the metal was approximately constant between the first and fourth pit samples, there was an appreciable increase of phosphorus between the furnace and first pit sample (*see* Figs. 8 and 9). Incidentally, a return of phosphorus can take place at this stage when there is a steady and large rise during the second period, but it was felt that here initial changes could be masked by the very pronounced return of phosphorus (*see* Figs. 7 and 9). In view of the fact that "good" casts, characterised by an initial phosphorus return followed by steady conditions, appeared to be connected with slags of high acidic content, it was likely that the initial phosphorus return was related to the amount of liquid component in the slag at tapping. The assumption of this simple and direct relationship has much to commend it, but the more complete examination of subsequent slag samples did not fully substantiate it. It is clear, therefore, that other factors are involved which tend to obscure the above simple relationship. One such factor appeared to be the degree of turbulence between metal and slag during casting from

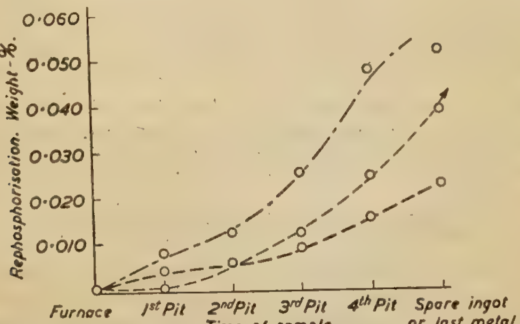

FIG. 7.—Course of Rephosphorisation of Pit Samples for Three Poor Casts.

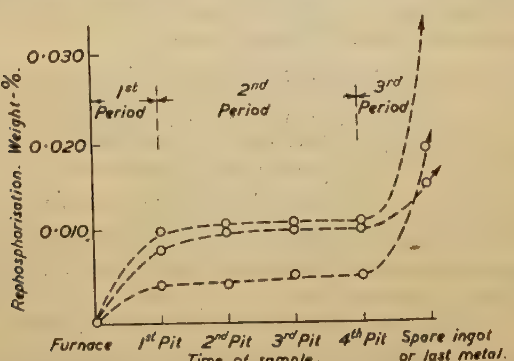

FIG. 8.—Course of Rephosphorisation of Pit Samples for Three Good Casts.

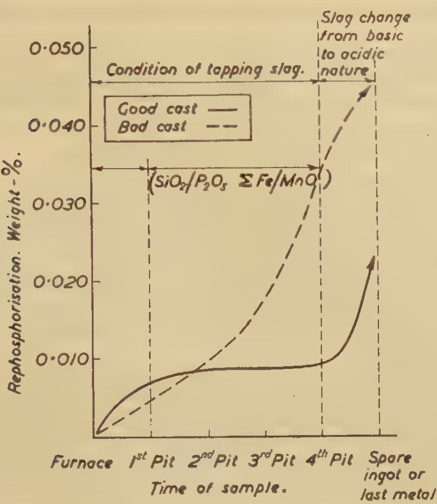

Fig. 9.—Course of Rephosphorisation of Pit Samples for Good and Bad Casts.

the furnace. The effect of this factor is discussed later. Moreover, the precise composition of the liquid component cannot, at present, be determined, but variations in it must undoubtedly affect the problem of rephosphorisation.

There still remained the problem of why a slag which held the phosphorus during the second period—from first to fourth pit sample—allowed phosphorus to return to the metal during the filling of the ladle and in particular during the period when the metal was covered by slag before ingot casting. It appeared likely that this rephosphorisation would be largely determined by the extent of mixing of metal and slag during the filling of the ladle. It was then postulated that during the filling of the ladle pronounced turbulence existed in the ladle and that the slag stream, following the metal stream, was drawn well beneath the metal surface and so for a few minutes there was intimate mixing of slag and metal. It was realised that when the ladle had been filled, the oncoming slag would not contribute much to this effect, owing to the shorter drop and the blanketing effect of the slag layer already present.

To test this postulate, an experiment was devised, details of which are given in Appendix III. Briefly, the metal and slag were represented by two immiscible liquids, the one of lower density being coloured. The two liquids were contained in a large vessel and allowed to run out through a side tube into a beaker. The amount of the heavier liquid was controlled to ensure that the beaker was three-quarters full when the lighter liquid entered the beaker.

The following facts were established :

(1) There was violent turbulence, as evidenced by the movement of air bubbles during the pouring in of the first (heavy) liquid.

(2) The flow of the second liquid was seen to go initially in a dispersed form to, or very near, the bottom of the beaker.

(3) As the beaker filled with the second liquid, the penetration below the top surface of the heavier liquid rapidly became less, until,

when the beaker was full and overflowing, the penetration was only just visible.

(4) Under these ideal experimental conditions the change between the two liquids was instantaneous. When the liquids come together for an appreciable time, the mixing will be more pronounced.

(5) The absence of any overflow of the heavier liquid was probably due to the ideal conditions of the experiment.

These experiments showed clearly that in the pouring of two immiscible liquids from a vessel into a beaker, violent agitation and mixing occur. The same must also occur between slag and metal during the filling of a ladle. It is therefore a reasonable assumption that the return of phosphorus to the metal is associated with this intimate mixing, being probably due to the release of phosphorus from the liquid component of the slag. While the presence of these phases could be inferred, their precise nature still remained unknown.

Experiments were then carried out in an attempt to study the constitution of molten furnace slag by the method of water-quenching (*see* Appendix IV.). The test consisted in taking a spoonful of slag from the furnace, about 15 min. before tapping, and pouring a quantity as quickly as possible into a bucketful of water. The time of transfer was 3 sec. The test was repeated after a few minutes' interval. The crystalline constituents were :

(*a*) a lime-rich phase and
(*b*) the lime–silica phosphate nagelschmidtite, together with
(*c*) a non-crystalline phase, *i.e.* a supercooled liquid.

It was concluded from a knowledge of their melting points, and especially the high degree of crystallinity exhibited by the patterns on the X-ray photographs, that the crystalline phases were present in solid form in the molten slag. It has not been possible to determine the composition of the liquid phase. Further tests with molten slag were made to observe any changes in constitution arising from delayed quenching of the slag. Slag samples were poured into water after holding in the spoon for 8 or 20 sec. The results showed that changes had taken place, the nagelschmidtite phase being less pronounced. This is explained by a reaction between the nagelschmidtite phase and the liquid phase, leading finally to the formation of dicalcium silicate and a silico-fluorapatite.

Dephosphorisation of Metal in the Furnace and the Probability of Rephosphorisation in the Pit.

The work on the constitution of quenched furnace slags and of spoon-cooled slags lends itself to a consideration of the process of dephosphorisation. From the foregoing, it is considered that the bulk of the phosphorus pentoxide is held in the slag as a solid lime–silica phosphate phase—" nagelschmidtite "; thus dephosphorisation becomes effective when this phase is formed. The process does not necessarily require the presence of a free-lime phase in the molten furnace slag (*see* data on casts made at the works of Messrs. Samuel Fox and Company, Ltd.). The addition of more lime will promote the formation of more nagelschmidtite (solid) and so enable more phosphorus to be held in the slag. Although basic open-hearth finishing or tapping slags contain more lime than is required to satisfy chemically the acid components, this does not necessarily mean that all the acids are combined into the lime–silica phosphate; in fact, a partition exists between the solid and liquid phases in respect to silica and phosphorus pentoxide.

It appears justifiable to regard the fluorspar as being in the liquid phase.

The occurrence of phosphorus in the slag as a phosphate indicates that a certain degree of oxidation is required. There are two sources of this oxygen—oxide additions and the furnace atmosphere—and both must be responsible for maintaining the state of oxidation of the slag. On transference of the metal and slag to the ladle, one important source of oxygen, the furnace atmosphere, is removed. Since the fixation of phosphorus as a lime–silica phosphate in the slag is a dynamic process influenced by the oxygen pressure as one factor, it follows that a slag which dephosphorises the metal to a certain low limit in the furnace will not maintain this low limit in the metal on transference to the ladle, even on the assumption that no reaction with the fireclay lining of the ladle takes place.

The return of phosphorus occurs during the filling of the ladle, as a result of the intimate mixing of the liquid component of the slag and the metal. Later, when the ladle is full, the return of phosphorus is controlled by the reaction taking place across the metal–slag interface, and the phosphorus is distributed in the metal by convection and diffusion.

On this basis the initial return of phosphorus to the metal between the furnace and the first ingot casting should be related to the amount of liquid phase, and the return of the phosphorus during the pouring of subsequent ingots should be controlled by the stability of the lime–silica phosphate phase.

The rise in phosphorus near the end of casting is presumably due to a change in the character of the slag from basic to acidic, as a result of reaction with the firebrick of the ladle.

The presence of an excessive lime content has been found to affect the oxidising power of the slag as instanced by :

(1) a relatively high phosphorus content in the metal at tapping, followed by a return in pit-metal samples, and

(2) the very small content of the ferrous-oxide-type phase in the spoon-cooled slag.

It would appear therefore that there is an optimum amount of free-lime phase in a well-balanced slag.

An increase in the amount of magnesia in the slag can cause rephosphorisation in the furnace. An explanation is that the magnesia reacts with the liquid phase normally containing the bulk of the iron oxide, and renders the slag less reactive by virtue of its increased viscosity. Furthermore, it should be noted that the magnesia is assciated with ferrous oxide rather than with lime, and unusually high contents of magnesia should not be regarded as essentially increasing the basicity of the slag, but rather as acting as an inhibitor of the oxidising power of the slag.

Acknowledgments.

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The author would like to place on record his appreciation of the great interest of the late Dr. T. Swinden in the progress of this work.

APPENDIX I.—*Synthesis and Examination of Substances to Explain the Full Details of the X-Ray Pattern of Solid Slag.*

The X-ray line pattern of spoon-cooled furnace slags indicated the possible presence of the following substances:

Lime	CaO-rich phase.
Ferrous oxide . .	FeO-rich phase.
Lime silicate . .	Very similar to β-$CaO.SiO_2$.
Lime–silica phosphate .	Very similar to that shown by the composition $7CaO.2SiO_2.P_2O_5$, known as nagelschmidtite.
Lime ferrite . .	$2CaO.Fe_2O_3$ with possible small alumina substitution.

It was apparent that, in addition to the above substances, new mixtures would have to be examined, as the presence of these phases did not fully account for the X-ray line pattern.

(1) *The Spacing of the Lime-Rich Phase.*

The spacing of this lime phase was found to be of the order of 4·74 kX. units (pure lime = 4·798 kX. units). The smaller spacing is due to solid solution of other oxides of lower spacing. The most likely of these is manganous oxide, since lime and magnesia do not combine in the pure state and the existence of the lime–ferrous oxide binary system is not onsidered possible, at least not in slowly cooled samples.

xperimental.

Mixture: CaO 80 weight-%, MnO 20 weight-%.

Heat-Treatment: Heated in a vacuum, temperature rising to 1300° C. in 2½ hr. Temperature maintained at 1300–1350° C. for 1 hr. Sample cooled quickly by withdrawal of evacuated tube from furnace.

Result.

The mixture showed a single phase, highly crystalline, with a spacing of 4·7412 kX. units. This result indicated that manganous oxide enters into solid solution in lime and that the observed lime-rich phase can be regarded, to a first approximation, as being a CaO–MnO phase. Further work has shown that CaO and MnO [5] are mutually soluble and the system CaO–MnO–MgO is being studied.

(2) *The Spacing of the Ferrous-Oxide-Rich Phase.*

Variation between 4·33 and 4·35 kX. units was found among the many samples examined. The spacings of the following relevant oxides, magnesium, ferrous and manganous oxides, are 4·20, 4·28–4·30 and 4·44 kX. units respectively. It appeared that the observed spacing of the FeO-rich phase was due mainly to a ferrous–manganous oxide phase.

Experimental.

Mixture: FeO 67 weight-%, MnO 33 weight-%.

Heat-Treatment: Heated in a vacuum, temperature rising to 1200° C. in 2½ hr. Temperature maintained at 1200–1250° C. for 1 hr. Sample cooled quickly by withdrawal of evacuated tube from furnace.

Result.

The mixture showed a single phase, highly crystalline, with a spacing of 4·3505 kX. units. Thus, the phase can be regarded as the ferrous-oxide-rich phase in the ferrous–manganous oxide system. Further work has shown that magnesia can be associated with this phase.

(3) *Tricalcium Phosphate.*

This substance was photographed and its pattern checked by known data; it was decided that it was not present in the spoon-cooled slag samples.

(4) *Tetracalcium Phosphate.*

Materials : Calcined lime and calcined tricalcium phosphate.

Heat-Treatment : Heated in air up to 1500° C. Temperature kept at 1500° C. for 1 hr.

It was decided that tetracalcium phosphate was not present in spoon-cooled slag samples.

(5) *Calcium Ferrite* ($CaO.Fe_2O_3$).

Materials : Calcined lime and calcined ferric oxide.

Heat-Treatment : Heated in air up to 1150° C. Temperature kept at 1150° C. for 1 hr.

It was decided that this substance was not present in these spoon-cooled slag samples.

(6) *Calcium Manganite.*

This substance was prepared, with the composition $2CaO.MnO_2$, having in mind the possibility that the manganese in the slag, reported chemically as MnO, may be present also in a higher-oxide form.

Materials : Calcined lime and calcined manganese dioxide.

Heat-Treatment : Heated in air up to 1500° C. Temperature kept at 1500° C. for 1 hr., cooled slowly to 1100° C.

The mixture contained free lime and a second phase which was not observed in the slag samples.

(7) *Lime–Silica Phosphates.*

(a) *Pure Lime–Silica Phosphates.*

Samples were prepared from calcined lime, precipitated silica and tricalcium phosphate or ammonium phosphate. One series was made to compositions on the line between $2CaO.SiO_2$ and $3CaO.P_2O_5$, the other series with excess lime. The tests prove that with ratios of SiO_2 toP_2O_5 between 3 : 1 and 2 : 3 by weight there was no change in the silica phosphate on addition of excess lime. (It was noted that the results of these X-ray measurements did not agree with the binary system $2CaO.SiO_2$–$3CaO.P_2O_5$ as given by Barrett and McCaughey.[4] In the first place, the solid solubility of P_2O_5 in β-$2CaO.SiO_2$ appeared to be appreciably smaller; secondly, a new phase was indicated between β-$2CaO.SiO_2$ and the nagelschmidtite phase; and thirdly, the nagelschmidtite phase extended into the suggested silico-carnotite region. These results were obtained on samples heated to 1500° and 1550° C. for 1 hr. and air cooled by withdrawal from the furnace. It may be that further changes will take place on slow cooling.)

In all these samples the mineral known as nagelschmidtite was found. Small changes in lattice dimensions were observed on passing from low to high phosphate content.

It was decided that the pure lime–silica phosphate phase was present to a varying degree in spoon-cooled slags. It was one of the main constituents in rapidly cooled molten furnace slags.

(b) *Lime–Silica Phosphates with Iron Oxide.*

Mixtures were heated in air on the assumption that it was ferric oxide rather than ferrous oxide which was associated with the lime-rich corner of the CaO–SiO_2–P_2O_5 system. Samples were heated to 1500–1550° C. in 3 hr.,

maintained at that temperature for 1 hr. and air-cooled. The compositions were based on analyses of furnace slag and a typical analysis was as follows :

CaO 68%		CaO 50%
SiO_2 16%	equivalent in	SiO_2 11%
P_2O_5 10%	the slag to :	P_2O_5 7%
Fe_2O_3 6%		Fe_2O_3 4%

The X-ray photographs showed lime and the nagelschmidtite phase.

(c) *Lime–Silica Phosphates with Fluorspar.*

Mixtures were made up to compositions along the line $2CaO.SiO_2$–$3CaO.P_2O_5$ with the ratios of SiO_2 to P_2O_5 of 2 : 1 and 1 : 1 by weight. Fluorspar was added in excess of that required to form an apatite-like mineral, *i.e.* the silica content was for this purpose regarded as phosphate.

In the first series the maximum temperature was 1500° C. for 1 hr. In the second series the temperature was only allowed to reach 1370° C. and maintained for ½ hr. Samples were air-cooled.

Series 1 : The pattern showed free lime and the lime–silica phosphate nagelschmidtite. The fluorspar had evidently dissociated and changed to lime.

Series 2 : The pattern was distinctive; no excess lime was found.

It was concluded that the fluorspar had reacted with the lime–silica phosphate.

The pattern was similar to that for fluorapatite, although with increased lattice dimensions. It was decided that the larger dimensions were consistent with the partial substitution of silicon for phosphorus in the fundamental apatite structure. It appeared that a silico-fluorapatite had been formed.

This pattern satisfied the lines on the X-ray patterns of spoon-slag samples hitherto unaccounted for.

Appendix II.—*Reaction of Lime on Ferrous and Manganous Silicates.*

This study was made to test the validity of the expressions :

$$2FeO.SiO_2 + 2CaO \longrightarrow 2CaO.SiO_2 + 2FeO$$
$$2MnO.SiO_2 + 2CaO \longrightarrow 2CaO.SiO_2 + 2MnO,$$

which state that lime replaces ferrous and manganous oxides in their silicates molecule for molecule.

In order that the ferrous and manganous silicates should contain no excess, *i.e.* uncombined, ferrous or manganous oxides, it was decided to use a slight excess of silica. Further, to ensure that no uncombined lime was present in the heated mixture, it was decided to use less lime than that required to replace all the ferrous and manganous oxides.

(1) *Preparation of Ferrous and Manganous Silicates.*

Materials.

Ferrous oxide : Prepared from ferrous oxalate. (Examined by X-rays and found to contain less than 1% ferrous and 1% ferric oxide.)

Manganous oxide : Prepared from manganous oxalate. (No trace of impurity revealed by X-ray examination.)

Silica : Pure, calcined at 900° C.

Experimental.

The powders were ground together in an agate mortar and put into an iron crucible supported vertically inside a refractory tube. The tube was

evacuated and then heated to the required temperature, *viz.* : (*a*) ferrous silicate 1150–1200° C., (*b*) manganous silicate 1250° C. After 1 hr. at this temperature, the tube was quickly withdrawn from the furnace and allowed to cool in air. The samples were examined in the X-ray camera to test for completion of the reaction.

Results.

The silicates were highly crystalline and possessed the olivine structure. The manganous silicate also contained some rhodonite, $MnO.SiO_2$. Free ferrous or manganous oxides were not observed.

(2) *Reaction of Silicates with Lime.*

Materials.

Ferrous silicate : $5(2FeO.SiO_2) + SiO_2$.
Manganous silicate : $3(2MnO.SiO_2) + 1(MnO.SiO_2)$.
Lime : Pure carbonate, calcined at 950° C.

Experimental.

The procedure was similar to that used in the preparation of the ferrous and manganous silicates. The temperatures used were : ferrous silicate and lime 1150–1200° C.; manganous silicate and lime 1250° C.

Results.

(a) *Ferrous Silicate and Lime.*—The photograph showed two olivines and ferrous oxide with no free lime. The ratio of the amount of silicates to ferrous oxide was 2 : 1, as indicated by the intensities of the patterns.

The ferrous oxide phase was found to have a lattice dimension of 4·298 kX., a value corresponding to that for ferrous oxide when in contact with iron and quickly cooled from, say, above 1000° C. (It should be noted that an iron crucible was used.)

The lattice dimension of the silicates corresponded to that of β-$2CaO.SiO_2$ and a lime–ferrous silicate of composition $(CaO)_{1·3}.(FeO)_{0·7}.SiO_2$ or 60 weight-% of $2CaO.SiO_2$ plus 40 weight-% of $2FeO.SiO_2$, as deduced from our X-ray survey of the $2CaO.SiO_2$–$2FeO.SiO_2$ system. The intensities of the two olivine phases suggested the following relative amounts : 2 parts of the lime–ferrous silicate and 1 part of the lime silicate. (On account of the poor crystallinity of the lime silicate it is possible that the amount is higher.) The results indicated an average composition of $(CaO)_{1·5}.(FeO)_{0·5}.SiO_2$ A mean composition of the silicate phases was calculated to be $(CaO)_{1·6}.(FeO)_{0·4}.SiO_2$, on the assumption that ferrous oxide was replaced by lime molecule for molecule. Further, a calcined mixture of lime silicate and ferrous silicate made up to the composition $(CaO)_{1·6}.(FeO)_{0·4}.SiO_2$ gave an X-ray pattern which contained no ferrous oxide but the patterns of β-$2CaO.SiO_2$ and the lime ferrous silicate, $(CaO)_{1·3}.(FeO)_{0·7}.SiO_2$, in the proportions given by the original mixture of ferrous silicate and lime.

The experimental results were therefore in excellent agreement with the calculated data based on the theoretical expression.

(b) *Manganous Silicate and Lime.*—This test showed that manganous oxide in a manganous silicate was replaced by lime to form a lime–manganous silicate and manganous oxide in the free state.

The manganous oxide was found to have a lattice dimension of 4·4355 kX. and is regarded as pure manganous oxide. (Pure MnO was found previously to have a spacing of 4·4354 kX.)

The line pattern of the lime-rich olivine was not sufficiently definite for the purpose of accurate identification of its chemical composition.

Summary of the Reaction of Lime with Ferrous and Manganous Silicates.

(1) Lime replaces ferrous oxide in ferrous silicate molecule for molecule.
(2) Lime replaces manganous oxide in manganous silicate molecule for molecule.

$$2FeO.SiO_2 + 2CaO \longrightarrow 2CaO.SiO_2 + 2FeO$$
$$2MnO.SiO_2 + 2CaO \longrightarrow 2CaO.SiO_2 + 2MnO$$

Appendix III.—*Experiment to Illustrate the Filling of the Ladle with Metal and Slag.*

(1) *First Tests.*

The furnace and launder are represented by a large glass vessel with side outlet and glass tube; the metal and slag by two immiscible liquids of different densities, the less dense liquid being coloured; the ladle by a 1000-ml. beaker. The arrangement is shown in Fig. 10. The exit tube is arranged so that its end projects slightly over the lip of the beaker and at an angle which allows the first stream of liquid to strike the bottom corner of the beaker. A pinchcock is clipped on a short length of rubber tubing attached to the end of the glass tube.

(2) *Second Tests.*

The second arrangement (*see* Fig. 13) was made to represent more nearly the works conditions. The furnace is represented by a large shallow dish, its base sloping towards the centre of the long side, at which point is situated a small exit tube. This tube projects into a U-shaped trough inclined towards the beaker.

The amount of the heavier liquid can be adjusted to give any desired depth in the beaker before the entry of the second liquid.

Tests were made with the following liquids:

(*a*) oil and water,
(*b*) water and carbon tetrachloride,
(*c*) water and bromoform.

The photographs (Figs. 11–12 and Figs. 14–15) show the filling of the beaker 75 and 90% full of the heavier liquid. The turbulence in the heavier liquid and the deep penetration of the less-dense liquid into the heavier liquid are clearly seen.

It was observed that the higher the level of the heavier liquid in the beaker before the entry of the lighter liquid, the less pronounced was the penetration of the lighter liquid towards the bottom of the beaker.

Appendix IV.—*Experiments Made to Determine the Constitution of Molten Furnace Slag.*

It was decided to obtain solid slag samples by quenching a small amount of molten slag in cold water. The time chosen was about 15 min. before the tapping of the furnace, when conditions would be favourable for a "good fluxed slag."

A slag-spoon was first heated in the furnace flame and then filled with slag. The spoon was then quickly withdrawn and a sample poured into a bucket of water. This was repeated with a new slag sample after a period of 5 min. The time between withdrawal and the quenching of the slag was 3 sec. (maximum).

X-Ray Examination.

The two slags gave identical patterns and showed two phases, a lime-rich phase and the lime–silica phosphate phase nagelschmidtite. No other

EXPERIMENTAL ARRANGEMENTS ILLUSTRATING THE FILLING OF A LADLE.

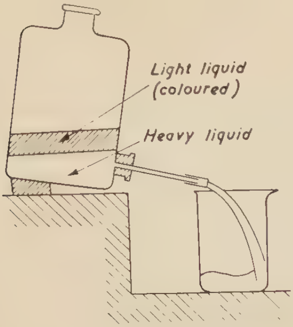

FIG. 10.—First Arrangement.

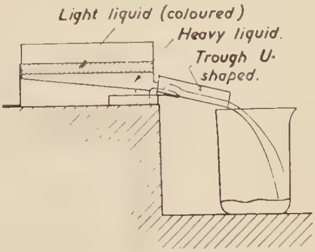

FIG. 13.—Second Arrangement.

FIG. 11.—Beaker 75% Full of First Liquid. Entry of second liquid in large stream.

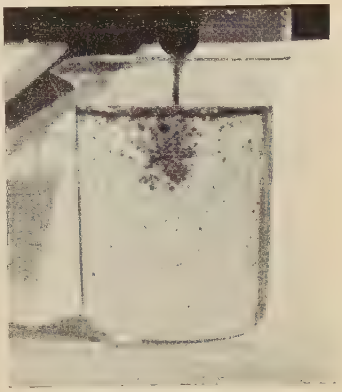

FIG. 14.—Beaker 90% Full of First Liquid. Entry of second liquid in small stream.

FIG. 12.—Later Stage in the Filling Process.

FIG. 15.—Later Stage in the Filling Process.

[*Jay.*

[*To face p.* 534 P.

pattern was observed. A non-crystalline (supercooled liquid) phase was deduced to be present in the sample from the high X-ray scattering at the high angles.

The lime-rich phase was strikingly well crystalline, showing clearly marked $K\alpha_1\alpha_2$ doublets. (Incidentally, this was the first occasion that this had been found with slags, probably owing to the fact that previous tests had been made on spoon or pit-cooled samples.) This degree of crystallinity indicated lime crystals having uniform composition and in equilibrium with the remainder of the slag. This fact, together with the observed dimension of 4·708 kX., which was indicative of a higher state of solid solution of other oxides, very probably MnO and possibly FeO and MgO, than in spoon-cooled samples, was very strong evidence that this lime-rich phase was present as a solid component in molten furnace slag. Further confirmation came from a consideration of the melting-point curves with respect to the deduced composition of this lime-rich phase.

In the case of the observed nagelschmidtite phase, it was only possible to identify its presence and not to make any comments on its crystallinity, because of the weakness of its characteristic pattern at high angles where the general scattering is high. However, in view of the fact that this phase has a melting point greatly in excess of the furnace working temperature [(4)] and that it was the only other crystalline phase observed, it seemed reasonable to conclude that this phase is also present as a solid phase in molten furnace slag. The presence of the nagelschmidtite phase at the working temperature appeared to be significant, since it indicated that fluorspar was associated with the liquid component. As the constitution of molten slag is different from that of slowly cooled slag, it appeared that, on cooling, this nagelschmidtite phase reacted with the liquid to form the silico-fluorapatite. The extent of this reaction will depend on the rate of cooling. It is possible that the presence of dicalcium silicate in slowly cooled slags is associated with this reaction.

In view of the success of this experiment, further tests were made later on another charge, in which the slag was held in the spoon for definite times—2, 8 and 20 sec.—before quenching by pouring into water. The constitution of the 8- and 20-sec. slags was found to be different from the sample quenched after 2 sec.; the lime-rich phase showed higher dimensions than the first sample, *viz.*: 4·707 (2 sec.), 4·711 (8 sec.) and 4.714 kX. units (20 sec.). (Spoon-cooled slags gave dimensions 4·73–4·74.) It was noted that whereas the lime-rich phase in the 2-sec. sample showed a high degree of uniformity of composition, the lime phase showed a progressive blurring for slags held longer in the spoon. This blurring of the high-angle reflection is here an indication of changing composition during cooling. The results are interpreted as showing that the solubility of MnO, and possibly FeO and MgO, is higher at the steelmaking temperature than at lower temperatures.

The pattern of the lime–silica phosphate phase was definitely weaker in the 20-sec. sample than in the 2-sec. sample. The photographs of the second (8 sec.) and third (20 sec.) sample showed other lines, which although diffuse were distinctly measurable on the third sample. These lines correspond to those of a cubic phase with lattice dimensions of the FeO, MnO, MgO solid solution. The pattern of this phase is usually clearly evident in spoon-cooled and slowly cooled samples.

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[This paper was discussed jointly with the preceding two by J. R. Rait and H. J. Goldschmidt on "The Constitution of Basic Steel Furnace Slags" and by Y. K. Zea on "The Phosphorus Reaction in Basic Open-Hearth Practice."]

JOINT DISCUSSION.

Dr. C. H. Desch, F.R.S. (President-Elect: Messrs. Richard Thomas and Baldwins, Ltd., London) : These are very interesting papers. It must be borne in mind that in the works one is mainly concerned with the liquid slag. The conclusion of Dr. Jay, that some of the active constituents are in solid suspension at steelmaking temperatures, is a matter which I hope we shall have fully discussed by the steelmaking representatives here, but ordinarily we are concerned with liquid slag in contact with liquid metal. By the time the slag has solidified no further reactions are taking place, so that the information given by the mineralogical constitution of the solid slag can give us only indirect information as to the reactive substances in the liquid.

It is generally assumed that a great degree of dissociation takes place, so that an ordinary silicate like Ca_2SiO_4 is actually largely dissociated into

$2CaO$ and SiO_2; but in papers by Herasymenko a few years ago we have also had the suggestion that a certain amount of electrolytic dissociation takes place into 2Ca and SiO_4 ions. It is very difficult to get direct information as to that. It does occur to me, however, that where it is possible to determine the freezing-point surfaces of a system such as a liquid slag one can infer something as to the degree of dissociation; it has been done in the case of metals, and could be done here.

A compound completely undissociated would give a sharp maximum on the freezing-point curve of the type shown in (*a*), Fig. A. That is theoretically impossible, because there must always be some dissociation, and the curve must be rounded at the top, as in (*b*)—you can have everything from a very sharp maximum to a flat maximum, as in (*c*), and one with a peritectic point, where the compound is completely broken up over a range of composition before reaching the melting point, as in (*d*). By examining a number of closely allied systems, and plotting on the same scale, one should be able to infer, from the degree of flattening of the maximum, the degree of dissociation into component oxides, but not, of course, the ionic dissociation, which would have to be determined by an entirely different method.

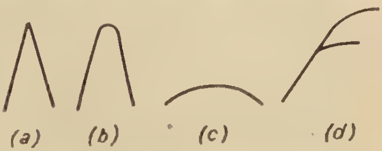

Fig. A.—Freezing-Point Curves.

I am not sure how far the freezing-point work on these slags has been carried, and whether it has gone far enough to allow of any such comparison as that to be made. We know that some of these diagrams are very incomplete. There is one on p. 434 P of the paper by Dr. Rait and Mr. Goldschmidt where the maxima have simply been dotted in because they have not been practically determined; a series of maxima have been filled in hypothetically, because the experimental work has not been done. But now that we have a good deal of information about the liquidus curves and liquidus surfaces of these systems, it might be found worth while to make a comparison of that kind. As a mere chemist, I can only express my admiration for the way in which the X-ray workers, dealing with very diffused lines of this kind and obtaining photometric curves such as these, are able to decide which of these things means a compound and which are only experimental variations, but I know that the technique has been very carefully worked out.

We shall have, I hope, a practical discussion on the question of rephosphorization, and it will be interesting to hear what operators have to say on the question of using a basic lining for the ladle to avoid this rephosphorization.

Unfortunately, one useful method does not seem to be applicable in these experiments, the microscopical method. Owing to spoon samples being used, the structure seems too fine for the use of a microscope; but in previous papers, such as that by Schneiderhöhn on Bessemer slags, the microstructure was so well developed that one could identify the phases quite easily, and check the X-ray results in that way.

Mr. A. Robinson (Appleby-Frodingham Steel Co., Ltd., Scunthorpe): I propose to confine my remarks principally to Dr. Jay's paper. I think it is rather unfortunate that Dr. Zea's most laborious work in dealing with the phosphorus reaction should have been confined to very low-phosphorus steels, the errors in sampling and analysis in such cases having a greater effect on his findings than would have been the case otherwise, the discrepancies and standard deviation of each method of calculation being of the same order.

The production of good basic steel can be approached only through a knowledge of the basic slag, the vehicle through which the refining is done, and a new technique has been used to give us that knowledge. The rephosphorisation of steel may take place owing to two causes : First, the upsetting of the balance between the slag and metal due to temperature changes or additions to the metal; secondly, the change of composition of the ladle slag due to its attack on the ladle lining. In the first case it occurs just before and at tapping, and this reaction is no doubt intensified by the mixing of slag and metal which takes place when tapping into the ladle, and which is shown by Dr. Jay's experiments to take place especially if the tap-hole becomes large, allowing the slag to come too soon. In the second case, the variable attack on the ladle lining causes a more acid slag, which is quite out of balance with the steel composition and is no doubt made worse by very hot and fluid slags and inferior ladle bricks. Fortunately this slag is in surface contact only.

The plots shown by Dr. Jay show some fairly good relation for a particular practice and composition of steel, but they are not universal, although the relation SiO_2/P_2O_5 against Fe/MnO is certainly the best and holds for a particular practice and similar steel compositions, and with a similar slag basicity and tapping temperature would no doubt give even better co-ordination. What does it mean? The basicity being fairly constant—

> A low-phosphorus charge can be tapped with a lower slag iron content, the manganese being constant.
>
> A high-silicon charge with constant phosphorus will require a larger slag bulk and this will dilute the P_2O_5 in the slag.
>
> A high iron content in the slag will hold the phosphorus in the slag.
>
> A high manganese content in the charge slows down the activity of the iron oxides and is liable to cause rephosphorisation, other things being constant.

All these things are confirmed by practice. In a previous paper given to the Institute, Mr. Jackson showed that in tilting-furnace practice, making the same type of mild steel, the sum of the SiO_2 and P_2O_5 in the slag bore a strict relation to the total iron in the slag, and that rephosphorisation could be avoided with a fairly constant and rather low lime content if the $SiO_2 + P_2O_5$ was kept to 23%, as measured by 11% of iron in the slag. No slag was allowed to go into the ladle. This total acid figure depends upon the quality of steel being made, high-carbon steels requiring a lower figure of from 18% to 20%. The manganese in the charges varied little, a fairly constant proportion of mixer iron being used.

It is shown by Dr. Jay that the mineral composition of the slag is not the same when molten as when cold, and that the fluorides present interact with the silico-phosphates on cooling, forming apatites. This may be important economically. Dr. Jay kindly provided me with samples of well-melted fluorspar slag taken at the same time, both granulated and cooled in the spoon. These were analysed, the 2% citric acid solubility was estimated, and the residues were analysed. The results were as shown in Table A. Whilst this cannot be claimed to be an accurate separation, it shows some difference in constitution. It will be noted that in both cases there is sufficient lime in each fraction to form the required silicates and phosphates, and that the silica is in the soluble portion, tending to disprove the presence of silico-fluorapatites or pointing to a very low form of combination.

Seeing that silico-phosphates are soluble and apatites virtually insoluble in the 2% citric acid test, this is a confirmation of Dr. Jay's find-

ings and shows that fluorspar slag could be improved as a fertiliser by granulating in water.

TABLE A.—*Slag Analyses.*

	Water-Quenched.			Spoon-Cooled.		
	Total.	Soluble.	Insoluble.	Total.	Soluble.	Insoluble.
SiO_2, %	16·40	14·93	1·47	16·87	15·14	1·73
FeO, %	6·22	5·40	0·82	6·55	4·72	1·83
Fe_2O_3, %	2·80	1·77	1·03	2·87	1·15	1·72
Total iron, %	6·80	5·45	1·35	7·10	4·48	2·62
CaO, %	44·80	33·74	11·06	45·20	31·27	13·93
P_2O_5, %	7·79	2·93	4·86	7·72	0·69	7·03
2% citric-acid-soluble P_2O_5	2·93			0·69		
Solubility of P_2O_5	37·60			8·90		
Insoluble slag	...	...	24·31	...	...	33·25

The theory that silico-phosphates and excess lime may exist in the molten slag principally as a dispersed solid is very interesting. The precipitation of silico-phosphates may explain the rapidity of dephosphorisation and the sluggishness of rephosphorisation. The liquid phase apparently contains lime, alumina, iron oxides, manganese and magnesium oxides and fluorides, and probably some silicates or silico-phosphates in solution. The effect of lime and/or iron oxides in helping dephosphorisation may be due to some precipitating effect on silico-phosphates. As it can be only the liquid phase of the slag that enters into the reactions with the steel, and the oxidation of the phosphorus may withdraw lime and silica from the liquid phase, we may be quite wrong in assessing the working slag basicity and oxidising power, and consequently the oxygen partition coefficient between slag and metal, which in practice is always higher in the steel than theory warrants.

Dr. Jay has whetted our appetite, and we want to know more. Can we in effect produce a lower silica content in the liquid phase or working part of the slag by having sufficient P_2O_5 present to precipitate it as a solid phase, or is it an advantage to have sufficient silica present in the slag to precipitate the phosphates and, if so, what precautions are necessary? This seems to be the logical conclusion to be drawn from Dr. Jay's paper.

Dr. T. FAIRLEY (Park Gate Iron and Steel Co., Ltd., Rotherham): I have been very interested in these papers on slags, and particularly in Dr. Jay's conclusions relative to rephosphorisation. Experience tells the steelmaker that if the basicity and oxidising power of the slag are adequate there is little danger of appreciable rephosphorisation. Dr. Jay has analysed his results, but he has not included any curve on which are given the lime, silica, phosphorus pentoxide and iron contents of the slags. With these points in mind, I have analysed the results which he gives and have drawn the following four graphs:

(1) Basicity against total iron content of the slags before the manganese addition for the plain carbon steels.

(2) Basicity against total iron content of the final ladle slags for the plain carbon steels.

(3) Basicity against total iron content of the slags before the manganese addition for the higher-manganese steels.

(4) Basicity against total iron content of the final ladle slags for the higher-manganese steels.

The basicity is expressed as the ratio of CaO to (SiO_2 + P_2O_5).

From these graphs it appears that for those casts of which both the "slag before manganese" and the "final ladle slag" have high basicities and oxidising powers, the rephosphorisation is small. If the "slag before manganese" has a good basicity and oxidising power and the "final ladle slag" has low basicity and/or oxidising power, then there is, of course, a danger of rephosphorisation. Cast 25/1518 illustrates this point. I feel that insufficient results are quoted on which to base a theory of rephosphorisation. For example, in Figs. 1 to 5 the results from two casts are used to define the conditions which are "bad" as regards rephosphorisation.

With reference to Fig. 1, Dr. Jay makes the comment that "for a given percentage of FeO the slags which show rephosphorisation contain a higher CaO/SiO_2 ratio than those which show negligible rephosphorisation." If, however, we take this a stage further and consider the basicities of the final ladle slags, then it is found that what I should expect does happen, namely, that for a given percentage of FeO the slags which show rephosphorisation contain a lower CaO/(SiO_2 + P_2O_5) ratio in the final ladle slags than those which show negligible rephosphorisation. An examination of the results based on these views also brings into line the unsatisfactory position of one sample to which Dr. Jay refers in item 3, p. 523 P.

On p. 524 P it is stated that cast 22/2420 is inconsistent in that negligible rephosphorisation occurred, whereas, from Dr. Jay's interpretation of the chemical analysis, marked rephosphorisation should have occurred. A final ladle slag is not given for this cast, but the basicity and oxidising power of the tapping slag are such that no marked rephosphorisation would be expected, and, unless something unusual happened between taking the tapping slag and teeming the ingots, rephosphorisation should not occur.

Since rephosphorisation occurs towards the end of a cast, in addition to studying the basicities and oxidising powers of the tapping slags, close attention should be given to the final ladle slags. The results given support the idea that if the basicity and oxidising power are adequate and are sufficiently maintained, then negligible rephosphorisation occurs.

Mr. F. L. Robertson (Messrs. John Summers & Sons, Ltd., Shotton, Chester): These three papers are so very different that I wish to deal with them separately. Dr. Zea has made a very careful examination of the use of Schenck's theories and constants for doing phosphorus sums, and he has found that within a range of P_2O_5 of 3–10%, of SiO_2 of 8–15%, and of total CaO of 49–55%, and at temperatures of 1550–1623° C., Schenck's formula works. That formula is:

$$\text{Log}\,[P] = \log K + \log(\Sigma P_2O_5) - 4\log(CaO) - 5\log(FeO) - 0{\cdot}06(P_2O_5).$$

He gets log K out of a book, and it varies with the temperature very markedly. The CaO and FeO terms he gets from tables prepared by Schenck.

He finds that the formula works within those ranges, and that means that the steelmaker, knowing his temperature and slag analysis, knows the phosphorus in his metal. Alternatively—and it is very important to grasp the alternative—if he knows the phosphorus in his steel he will know how much lime to add to his charge to reduce his phosphorus to specification. He has no control over his iron oxide; that is fixed by

other necessities of steelmaking. But he can add or withhold lime, and so we come back to the old-fashioned way of making steel, but now we have some moral support, and we can go on adding lime until the phosphorus in the steel drops to specification.

I explored Schenck's formula over a slightly wider range of composition ten years ago, when his work was first published, and I found such extraordinary agreement with my furnace work that I adopted the whole of his theories. If I was using that equation, it is obvious that I was accepting his principle of the partition of free oxide between steel and slag. I was swallowing that, so that I might as well use it for other things. I was also swallowing his division of the lime at different temperatures and different analyses into combined and free. When I applied those things to my other steelmaking difficulties, I found an increasing ease in handling factory problems.

Because Schenck's formulæ and tables seem to give useful results within the limits of our experience, and because his theories give us distraught melters a pretty picture, simple enough for us to understand and use, their scientific truth is by no means proved. There are some hard things to swallow in this phosphorus formula. We know, for instance, that log P should be 2 log P, which is rather a big difficulty to start with, and the argument advanced for using log P and not 2 log P seems to me rather weak. There is also this 0·06 as a correction factor of the total P_2O_5 in the slag. When you go up to 24% of P_2O_5 that is quite a big factor. If, however, we find that a formula of that kind works, we must go on using it as long as it helps us to make steel, but we must always be looking for the new ideas which will upset it and give us a better and easier way of making steel. In that connection, I think it is a pity that Dr. Zea has not pointed out and offered an explanation of the fact that some of his melted samples—I refer especially to *T*.2 and *T*.3, and in a lesser degree to *T*.9—do not obey Schenck's rules. These slags should have under them, in the case of *T*.2, melted, 0·083% of phosphorus instead of 0·22%; in the case of *T*.3, melted, 0·027% instead of 0·25%; and in the case of *T*.9, melted, 0·11% instead of 0·137% (*see* Table XX).

There is another very important exception. In the English translation of Schenck's work, which is to be issued shortly, on p. 159 there is an important exception in the Hoesch process, which does not come out according to those tables.

I should like to ask Dr. Zea to consider these, and to see whether he can give us some explanation for them. I myself have brushed the exceptions aside quite easily, by saying (1) before the boil the slag is not a liquid but only a paste. The partition law and the law of mass action cannot, therefore, be expected to work; the law of mass action is concerned with two liquid phases, and if one is semi-solid it does not apply. (2) I consider that the slag samples before the boil are not representative of the whole bath, and are not homogeneous. I think that Dr. Jay's work, in which he finds solid lime, unfluxed lime, bears that out. However, that on my part is rather wishful thinking; I want to think it, and so I do. I feel that a proper study of the cases where Schenck's formulæ do not work would be useful. I would ask Dr. Zea, if he has the opportunity, to consider his study of those cases, and especially of the higher P_2O_5 values and of pre-boil times. The discovery for which we are looking is how to dephosphorise a 4% carbon iron at 1400° C.—that is to say, at a more convenient temperature from the engineering point of view than 1600° C.

After reading Dr. Zea's paper, I am struck by the fact that Dr. Jay's paper comes as a flat contradiction to it. If we accept, as I believe we must, the fact that calcium silico-phosphate is the phosphorus com-

pound with which we are dealing, we have little right to pay such minor attention to SiO_2 in our calculations. It is a little strong on our part to take a formula such as the Schenck phosphorus formula, which does not include silica at all, and say that that is the law, when the compound about which we are talking is a silico-phosphate. That statement of mine is not quite true, because in that formula, with the free lime and free oxide, the silica is taken account of, but as far as the law of mass action goes it is simply not there. I am completely at a loss to reconcile Dr. Jay's and Dr. Zea's papers, and I feel that we must ask the authors of those papers to help us there. I suspect that they are not using the terms "free lime" and "free iron oxide" in the same sense. Schenck's "free lime" and "free iron oxide" are in solution; that is, single molecules of FeO and single molecules of CaO are moving about among molecules of calcium phosphate, calcium ferrite, calcium silicate and iron silicate. I feel that Dr. Jay, on the other hand, when he speaks about "free lime" is really referring to aggregates of molecules of CaO that are floating as solids through the liquid slag. That is quite a different idea from solution. I wonder whether Dr. Jay would help us on that point?

Let us see, in a commonly occurring slag analysis, what the effect of dissociation (according to Schenck) will be. A 100° C. difference of temperature will change the free FeO from 3·4% to 5·8%, and the free CaO from 26% to 31%, which is a terrific difference. If this dissociation, to which Dr. Desch has referred, is to be ignored, then Dr. Zea's paper drops into the same category as the rule-of-thumb happy accident of being able to add dissimilar acids and do sums with the results. That is a line which as steelmakers we should consider unfortunate. If we are forced to take it we shall have to do so, but we must try to get on to some chemical theories and ideas which will stand on their own legs, which obviously adding acids with different chemical properties will not.

I beg Dr. Jay and Dr. Rait to help us poor furnacemen with a little more arithmetic. In the end, we have to do sums, and if in the past we have been doing them wrongly, it is only men like Dr. Jay and Dr. Rait who can help us, and we look to them to show us the right way.

Mr. D. Manterfield (Messrs. Steel, Peech and Tozer, Sheffield): It has been my pleasure to be associated with Dr. Zea while he was carrying out his studies, and to be in touch with Dr. Jay while his work was in progress, so that I feel rather ungracious in saying one or two of the things that I have to say in commenting on their papers. I know that they will receive my remarks in the right spirit, and that any criticism which I have to offer will be regarded as constructive. Nothing that I may say detracts from my opinion of the excellence of their papers, and of that of Dr. Rait and Mr. Goldschmidt.

Dr. Zea has studied plant data very carefully, and has tried to correlate the many variants with the degree of rephosphorisation, eventually finding direct correlation between this and a certain free-CaO/free-FeO ratio, and that temperature also plays an important part. As he states in the paper, he has used Schenck's method in making these computations. In my opinion, and in that of some others, Schenck's methods are not above criticism, and certain of his analogies and assumptions are not regarded as valid. Dr. Jay's work, for instance, throws considerable doubt upon the existence of $FeO.SiO_2$ or $MnO.SiO_2$ in a truly basic slag. Nevertheless, using these methods of computation Dr. Zea has found this direct relationship between a free-CaO/free-FeO ratio and the degree of rephosphorisation.

An examination of the reactions concerned in dephosphorisation and rephosphorisation suggests at once that lime and iron oxide do determine the equilibrium position of the reaction, and, as temperature also affects

the equilibrium constant, obviously it plays a part too. The difficulty is in assessing these concentrations or activities. In my opinion, we do not yet possess a reliable method of assessing these free, active, or uncombined compounds, however you may describe them. When we can do that, we shall have a method whereby we can quantitatively calculate the equilibrium constants and the phosphorus content of steel from slag composition. I have recently calculated these free-CaO/free-FeO figures for ten slags according to Dr. Zea's application of Schenck's methods, and the results are quite good. They are not quite as accurate as Dr. Zea has found, but the calculated results are tolerably near the analytical figures.

One obvious question arises out of Dr. Zea's paper: Why is the ratio of 6·4 required for high-carbon steels and only 5·2 for the lower-carbon steels? It would appear that this is concerned with the "oxygen level," if I may use that term. The essential difference between high-carbon and low-carbon steels, as far as the FeO is concerned, is the distribution of this FeO between the slag and the metal, this being chiefly dependent upon the carbon and temperature. It occurs to me that if we could accurately determine this distribution, then, instead of using the FeO in the slag or the free FeO as computed by Schenck or by other methods, either the FeO content of the metal or the partition coefficient between the slag and metal would probably be a primary factor in our calculations and equations.

This seems to point to the fact that Dr. Zea's two variables, the free lime and the free iron oxide, are (as calculated) possibly secondary or dependent variables, pointing the way to some primary variable. If we could discover what that primary variable is, or what those primary variables are, we could probably construct a simple formula which would do for all carbon contents, and not have a different formula for 0·60% of carbon from that for 0·25% and 0·30% of carbon.

As for Dr. Jay's paper, like Mr. Robertson, I feel that we shall have to accept without any reserve the existence of lime silico-phosphates in the solid slags, and very probably in the liquid slags too. Some confirmation of that is given in the paper by Dr. Rait and Mr. Goldschmidt. But this does not affect our conceptions of the fundamental equations concerned in the phosphorus reaction; it is simply a further step forward in our knowledge of this slag-metal reaction.

I agree with Dr. Jay that there are three stages of rephosphorisation such as he states, but I cannot agree that the first stage is due to turbulence, as he suggests, nor do I think that his laboratory experiment altogether simulates actual conditions. Recently I took two samples from two casts, one from the furnace before the addition of ferro-manganese and the other from the lander before the steel went into the ladle (and after the addition of ferro-manganese to the bath), and further samples during teeming. I chose high-manganese casts because there were quite large additions of ferro-manganese. The results were as follows:

	Cast 25/2647.	Cast 31/2244.
Sample before manganese addition	0·013% P.	0·018% P.
	10 cwt. Fe–Mn (bath).	15 cwt. Fe–Mn (bath).
Lander sample . . .	0·017% P.	0·024% P.
	9½ cwt. Fe–Mn (ladle).	14 cwt. Fe–Mn (ladle).
	11 cwt. Si–Mn (ladle).	4 cwt. Fe–Si (ladle).
1st pit sample . . .	0·021% P.	0·029% P.
2nd ,, . . .	0·022% P.	0·030% P.
3rd ,, . . .	0·023% P.	0·031% P.

This points to the fact that the initial stage of rephosphorisation is simply due to deoxidation, by the addition made, shifting the equilibrium

position to the left and causing some reversion of phosphorus from slag to metal. We must bear in mind, too, the different conditions, particularly with regard to the oxygen pressure, when the steel is in the ladle, compared with when it is in the furnace. Those factors too, in my opinion, account for the higher Fe/MnO ratio quoted by Dr. Jay in his last diagram for high-manganese steels than for low-manganese steels; more ferro-manganese is added and there is a greater degree of deoxidation, requiring a higher iron/manganese ratio.

Dr. Jay does not claim any finality for his theory with regard to SiO_2/P_2O_5 and Fe/MnO ratios and rephosphorisation. It occurred to me that, of these four variables, three (SiO_2, P_2O_5 and MnO) are determined largely by their respective burdens in the charge, and in a particular charge they are invariable in normal practice. The only real variable is the iron, which is varied by temperature, by the addition of iron oxide in the form of ore or scale, and by the addition of lime. When one considers this, one comes back again to basicity, state of oxidation and temperature.

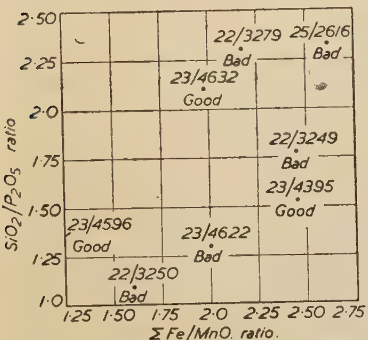

FIG. B.—Relation between SiO_2/P_2O_5 Ratio and Fe/MnO Ratio in Tapping Slags.

I have calculated on the basis of Dr. Jay's final charts the SiO_2/P_2O_5 and Fe/MnO ratios for eight recent slags of which I have complete data; the figures are given in Table B and Fig. B. Four of the samples, 25/2616, 22/3279, 22/3249 and 23/4596, show serious discrepancies in that the first three "bad" casts fall into Dr. Jay's good grouping and the last, a "good" cast, into the bad grouping. Dr. Jay, I know, can and will reply to this and tell us why these discrepancies have occurred. This does not detract in any way from the value of his work; it simply points to the fact that, as he stated in introducing his paper, other factors are involved. It appears to me, then, that Dr. Jay's variables are also secondary or dependent variables, pointing to some primary variables which are probably the same as those to which Dr. Zea's two variables point. If we could, as we probably shall in the near future, discover those primary variables, that will probably reconcile the apparent contradictions in Dr. Jay's and Dr. Zea's papers.

TABLE B.—*Slag Analyses.*

Cast No.	Phosphorus, %, in Pit Samples.		SiO_2, %.	Fe, %.	CaO, %.	P_2O_5, %.	MnO, %.	SiO_2/P_2O_5 Ratio.	Fe/MnO Ratio.
	First.	Last.							
23/4395, good .	0·033	0·041	9·8	12·0	55·3	6·25	4·99	1·56	2·4
23/4596, good .	0·033	0·038	10·5	9·0	51·66	7·79	7·13	1·35	1·26
22/3250, bad .	0·046	0·061	10·9	9·8	50·82	9·78	6·10	1·11	1·60
22/3249, bad .	0·040	0·055	9·12	12·6	54·04	5·15	5·10	1·77	2·47
23/4622, bad .	0·035	0·051	9·8	10·7	55·3	7·43	5·44	1·32	2·0
23/4632, good .	0·030	0·033	12·4	10·4	54·3	5·93	5·24	2·09	1·99
22/3279, bad .	0·032	0·051	15·0	9·7	53·0	6·48	4·46	2·31	2·18
25/2616, bad .	0·035	0·060	9·9	14·8	50·58	4·24	5·67	2·33	2·61

Mr. T. F. Pearson (Messrs. Colvilles, Ltd., Cambuslang): These three papers are difficult to comment upon comprehensively, since the subject matter is so extensive and complex, particularly in the paper by Dr. Rait and Mr. Goldschmidt. It is a very considerable step indeed from the early simple acceptance of phosphorus existing in a basic open-hearth slag as calcium phosphate, to the present day, when we have considerable knowledge of some very complex compounds, containing, in addition to lime and phosphorus, silica, fluorine and hydroxyl groups. However much Mr. Robertson dislikes the idea, we shall have to accept the fact that phosphorus is present as silico-phosphate, because, in addition to the X-ray evidence, there is considerable chemical evidence. The extraction of phosphorus by any selective solvent is always accompanied by silica, and the two are inseparable.

With particular reference to the paper by Dr. Rait and Mr. Goldschmidt, I am interested in the complex phosphorus compounds, and especially, as Mr. Robinson suggested, from the point of view of the citric acid solubility of slags; there are several features on which I should welcome more information.

I find that if we examine slags from many open-hearth charges for citric acid solubility by sampling at frequent intervals from beginning to end, the citric acid solubility values display a peculiarity which I do not think is fully realised. Early slags are low in solubility. During the working of the heat they climb to a maximum which may be of the order of 96%, and then may sink at tapping down to 80–85%. This applies to individual charges to which no fluorspar has been added, and also to consecutive charges on a fixed furnace where any possibility of fluorspar is discounted altogether. It applies also to the tilting furnace. The present papers are of considerable interest from the point of view of this alteration in solubility. I would ask Dr. Rait or Dr. Jay to examine by X-ray methods low-solubility "early" slags, because I believe that some of the phosphorus is there as iron phosphate. There seems little doubt that the highest solubilities obtainable coincide with some maximum value of the complex silico-phosphates in the slag. I can interpret the sinking of the solubility towards the end (in the absence of fluorspar) only by assuming that, in addition to the silico-phosphate and the free lime, a small proportion of either $3CaO.P_2O_5$ or $4CaO.P_2O_5$ (hilgenstockite) does form.

Dr. Rait and Mr. Goldschmidt, however, appear to discount the possibility of the co-existence of $3CaO.P_2O_5$ (steadite), and I should like to ask whether this is so?

Dr. Rait: Yes.

Mr. Pearson (continuing): The presence of hilgenstockite would certainly account for the lowered solubility, since its solubility value when cooled in air is stated by White to be only 70·5%, and it is true that its co-existence with silico-phosphates occurs only in low SiO_2/P_2O_5 ratio slags. I should welcome more information, however, and remain puzzled by the statement that the lower solubility of the tetra-calcium phosphate is due to the formation of a hydroxy compound. It is difficult to visualise —under normal cooling conditions on the stage, with a full spoon of slag with but a limited quiescent surface exposed to the atmosphere—the penetration of moisture throughout the mass. Hence my questioning the complete absence of steadite.

Interesting subsidiary evidence with regard to this lowered solubility is obtainable from the behaviour of natural rock phosphates, all of which contain fluorine and are not completely citric acid soluble. Whilst the

bulk of the mineral is the insoluble fluorapatite, the portion which is soluble appears to exist as the di- or mono-calcium phosphate, since the natural mineral is to some extent hydrated. If the rock phosphate is heated to 1050° C., its citric acid solubility is completely destroyed. My explanation is that the di- and mono-calcium phosphates have now been dehydrated and converted into insoluble $3CaO.P_2O_5$. Such evidence suggests that " insolubility " may arise through dehydration rather than by the formation of a hydroxy compound. In addition, the acceptance of the formation of the hydroxy compound as being responsible for the lowered solubility is complicated by the fact that quenching in water improves the solubility. The effect of quenching may be due, however, to a retention of the α-form of the $3CaO.P_2O_5$, which is known to be almost completely soluble. Quenching from above 1400° C. is necessary. I would welcome the authors' opinion on this apparently contradictory evidence.

Incidentally—and both Mr. Robinson and Mr. Manterfield have touched on this point—it appears probable that the co-existence of more than one phosphate prevents a more satisfactory formulation of the phosphorus equilibrium expression which Dr. Zea has examined. There seems little doubt that the Schenck and Riess formula, whilst giving the most satisfactory results to date, is inadequate; and Schenck himself acknowledges a weakness in assuming the validity of the tetracalcium phosphate by employing an empirical correction factor of 0·06.

Turning, now, to the question of slag control, Dr. Rait and Mr. Goldschmidt make an interesting contribution to this problem with their X-ray methods, but I would suggest that the ultimate aim is not control of the slag during melting, when we cannot do much about it and when it is better left alone, but rather control of the slag during refining. To myself, the problem is not what one should add to a slag in order to adjust it, but how much one should add. The melter is not overburdened when he has a choice of three materials only—lime, iron oxide or fluorspar—and it is of more importance to know how much to add (particularly iron oxide) rather than which one of the three to add. I suggest that a method by which this problem might be tackled is by adding to a freely-moving slag, or charging with the burden, an accurately known weight of some material which is not normally present, such as a barium salt, and then rapidly determining its concentration in the slag when completely formed, and from that working out the weight of slag present. Such a method might conceivably yield more accurate results than a dip-stick method, because of the differences occurring from charge to charge in the erosion of banks and the lack of precise knowledge of the bath area. An accurate knowledge of slag weight would enable quantitative additions to be made.

Dr. Jay's evidence of the existence of solid phases, and particularly of the nagelschmidtite phase, in the molten slag is very interesting, and possibly the improved citric acid solubility associated with the quenching of slags containing fluorspar may be due to the retention of a greater proportion of this phase and a prevention of its complete reaction with the fluorspar associated with the liquid component, which gives fluorapatite on slower cooling. Such an explanation is obviously an alternative to the suggestion that I made earlier regarding the presence of α-$3CaO.P_2O_5$. I would ask Dr. Jay what was the lowest lime content of a slag in which he has found solid phases present? All his slags tabulated contain over 50% of lime. I would also suggest to him that rephosphorisation in the ladle, in addition to being aided by mechanical turbulence in filling, is also assisted by the normal convection currents present afterwards, as described by Land in his paper on the mechanism of cooling in the ladle.

Mr. D. A. Oliver (Director of Research, Messrs. William Jessop & Sons, Ltd., Sheffield) : I should like to welcome Dr. Zea's paper as being, I believe, the first on the physical chemistry of steelmaking which has been presented since we have had reliable temperature measurements. Mr. Manterfield recently gave us an excellent paper on quick-immersion temperature measurement technique in very large furnaces, and it is good to see that Dr. Zea has used that new tool in the study of the physical chemistry of steelmaking. This has been one of the objectives of the Liquid Steel Temperature Sub-Committee for years, and we are beginning to see the fruits of past work.

With regard to the application of Schenck's work and the alleged agreement, I believe that Schenck's temperatures are in error; how much I do not know, but I should estimate by at least 30° C. I would ask Dr. Zea whether he corrected for possible errors in one or other direction, or whether he took his own accurate temperatures and imposed them on the Schenck data, on the assumption that Schenck was correct; because if that is so it may be that on further study the agreement will not be found to be quite so close. Also, Schenck was quite honest in his work and said that he put in various empirical and numerical corrections to make the formula agree with a large mass of practical steelmaking data. That is important, because all we are saying is that steelmaking in Britain is agreeing with steelmaking elsewhere, provided that there is some empirical formula to link up the variables. In discussing this work on the physical chemistry of steelmaking we must bear in mind that in striving to get some basic laws adjustments have been made, and the field is still open for a more fundamental study.

I should like to emphasise the point to which Dr. Desch referred in his opening remarks. We are all aware that molten slags are dissociated to a certain degree and that they are ionic, because their electrical conductivity rises with temperature. Just how ionic we do not know precisely; but going forward boldly on the lines of the X-ray workers we can get a vast amount of reliable information on the assumption that the compounds are present as molecular phases and that they are not highly dissociated. The justification for a patient study of slags in the solid state rests on this, that until we know which phases to study we cannot study them in detail. Future work will undoubtedly be concerned with studying some of these phases and their temperature dependence in the presence of different co-operating variables.

I believe that that is one of the reasons why Dr. Rait and Mr. Goldschmidt confine themselves severely to a statement of constitution. The time is not yet ripe to come to grips with the full steelmaking problem, and their paper is a contribution to identification. The next stage will be to study the systems with temperature, and finally we should be in a better position to make an onslaught on the broad problem. I would associate myself with previous speakers in saying that the really important thing is the effect of slag constitution on the partition coefficient between the oxidising power of the slag and the oxide (or oxygen) in the metal, because that balance must determine the quality, inclusion type and a large number of other factors. All that I would plead is that we do not try to go too fast, because if we do we shall only have difficulty with conflicting view-points. We still have to rely very much on the skill and art of the steelmaker, although striving all the time to make steelmaking a better and more exact science.

Dr. J. H. Whiteley (Consett Iron Co., Ltd., Consett, Co. Durham) : I will confine my remarks to Dr. Zea's paper, into which it is quite obvious that he has put an enormous amount of work and thought. Dr. Zea has

done me the honour of referring to a paper that I gave on dephosphorisation over twenty years ago, which I believe was one of the earliest, if not the first, to deal with the matter from the standpoint of physical chemistry. I may say that the paper was in the nature of a preliminary investigation, and I fully intended to pursue the subject further, but was prevented from doing so. My method of attack was to obtain the molecular proportions CaO, MgO, FeO and MnO, and the two acids SiO_2 and P_2O_5 in the tapping slags and, after subtracting from the total for the bases three times the P_2O_5, to divide the remainder by the SiO_2. The molecular ratio of available bases to silica thus arrived at I termed the basicity index. I deducted three times the P_2O_5 because it seemed to me more likely that in the molten state the tribasic rather than the tetrabasic phosphate of lime would be present, but even if the latter form occurred the results would not have been seriously affected. Having obtained the basicity indices for a number of tapping slags I then plotted each one against the ratio of the amount of phosphorus in the slag to that in the pit sample, after allowing for the small increase due to the ferro-manganese addition, and found that a fairly linear relationship existed for the class of mild plate steel containing about 0·5% of manganese which was being made. For several reasons, however, I could not claim a high degree of accuracy and therefore made no attempt to ascertain the best straight line to be drawn through the points as Dr. Zea has done. For one thing, I was not sure that the SiO_2 determination was not on the low side, since fluorspar was always added to the bath during the boil and might cause some loss of SiO_2 as silicon fluoride in the estimation. Again, there was no certainty that the CaO was all in solution in the slag, although I took care to select casts in which no lumps of lime could be seen in the bath. Then, too, there was the Fe_2O_3 to consider. Like the SiO_2, that also might be on the low side, owing to some reduction by the H_2S liberated when dissolving the slag sample in hydrochloric acid. On the other hand, the amount might be increased by air oxidation as the sample cooled, although I could find no evidence of that. Further, there was, and still is, no means of ascertaining what proportion of the higher oxide existed as Mn_3O_4. But even if the actual Fe_2O_3 had not been in doubt I did not know whether to regard it as a base or an acid, and I do not think that even yet any one else does. Being in this quandary I decided to convert the total iron in the slag to FeO and to ignore the Fe_2O_3.

Notwithstanding these uncertainties, the curve indicated unmistakeably that the silica content of the slag was the most important factor. With all due respect to Dr. Jay, the results showed that dephosphorisation increased as the molecular ratio of the bases to the SiO_2 was increased; in fact, having once obtained the curve, I found again and again that I could arrive at the phosphorus content of the steel within close limits from the analysis of the tapping slag. Since that work was published three other methods of determining the phosphorus distribution from the slag analysis have been advanced, and Dr. Zea has recorded in Tables I., II., III. and VI. the results yielded by each one on a series of casts of varying composition. In these particular cases Schenck and Riess's method seems to give the closest agreement, but there is not much to choose between us, for in some instances we all appear to be rather wide of the mark.

As a result of his investigation Dr. Zea maintains that the degree of rephosphorisation in the ladle depends upon the ratio of free CaO to free FeO in the tapping slag as determined by Schenck's method. That method of estimation is, to say the least, of doubtful value, but, in any case, I fail to see why free CaO should be set against free FeO, since both these compounds aid in dephosphorisation and both attack the ladle

lining. In drawing the curve in Fig. 30 I am afraid that Dr. Zea has let his theory over-reach his judgment, for if the single point on the zero ordinate be ignored a quite different curve is obtained and the linear relationship vanishes.

I hold the view that the extent of rephosphorisation in the ladle depends primarily upon the degree to which the SiO_2 content of the slag is increased when in contact with the brick lining, and into that effect quite a number of factors enter. Among them I consider slag viscosity is of great importance, since the greater that is the slower will be the diffusion in the slag of the SiO_2 and Al_2O_3 taken up by its attack on the lining. Owing to that action and to their viscosity, basic slags are bound to vary in composition across the ladle as teeming proceeds, so much so in some cases that possibly both phosphorisation and dephosphorisation may actually proceed simultaneously at different places. This lack of uniformity of the ladle slags can be inferred from all those cases described by Dr. Zea. As examples I will take the fairly representative *R*.1 and *R*.2 slags. In the case of the *R*.1 tapping slag, the analysis total is 95%, whilst that of the ladle slag is only 92·71%. The difference of 2·29% I regard as due to the Al_2O_3 absorbed from the lining, and would note in justification that Dr. Zea says that the Al_2O_3 contents of his ladle slags were increased from 2·5% to 3·5% as a result of the attack. Then, assuming that at the end of teeming the ladle slag weighed 10 tons, a rise of 2·29% in the Al_2O_3 would mean that about 13 cwt. of brick had been dissolved. That amount is probably rather high, but even if there were only 5 tons of slag it would make very little difference to the argument, so that I will keep to 10 tons. Now, since the ladle bricks would contain about 60% of SiO_2 and 36% of Al_2O_3, an addition of the above amount of Al_2O_3 to the slag should be accompanied by an increase of the SiO_2 content from 7·78% at tapping to 11·3%, whereas it has actually risen to 26·08%. These figures, together with those likewise obtained for the CaO and MgO, are given in Table C, which also contains the results derived in the same way from the *R*.2 tapping and ladle slags. In making these calculations I have allowed an approximate value for the relatively small amount of SiO_2 which entered the slag as a result of the oxidation of the silicon in the steel.

TABLE C.—*Difference between the Actual and Calculated SiO_2, CaO and MgO Contents of two Ladle Slags, taking the Estimated Al_2O_3 rise as the Basis of Calculation.*

Cast.		Tapping Slag.	Ladle Slag, Calculated.	Ladle Slag, Actual.
R.1	SiO_2, %	7·78	11·3	26·08
	CaO, %	51·0	47·7	42·4
	MgO, %	5·55	5·2	5·58
	Al_2O_3 increase by diff., %			2·29
R.2	SiO_2, %	10·4	15·9	20·16
	CaO, %	51·7	46·0	43·60
	MgO, %	7·0	6·2	5·95
	Al_2O_3 increase by diff., %			4·01

The variability of the ladle slag composition which is to be inferred from the figures in Table C can be demonstrated in another way. If 26·08% represents the average SiO_2 content of the *R*.1 slag, then every 6·5 tons which entered the ladle dissolved 3·5 tons of brick and the CaO should have dropped to 33·15%. In the case of the *R*.2 cast 8 tons of tapping

slag would require 2·0 tons of brick to give the SiO_2 increase with a lowering of the CaO to 41·36%. As these amounts of brick are greatly in excess of the usual lining wastage per cast, it is again evident that the ladle slag analysis is not to be depended on in dealing with rephosphorisation. There appears to be no means of ascertaining precisely what changes take place in the ladle slag composition, for which reason I see little point in endeavouring to find the cause of rephosphorisation by examining the ratios of the bases, whether combined or free, in the tappng slags.

Mr. A. J. K. HONEYMAN (Messrs. Colvilles, Ltd., Motherwell): I would suggest that the composition of the final ladle slags is of the utmost importance when considering rephosphorisation. The data presented may be plotted in various ways, one of which, namely, the CaO/SiO_2 ratio against the P_2O_5/P ratio, shows that as the CaO/SiO_2 ratio increases to about 5, the P_2O_5/P ratio also increases, indicating efficient dephosphorisation. Beyond that, however, there does not appear to be any advantage in increasing the CaO/SiO_2 ratio. However, in the final ladle slags the CaO/SiO_2 ratio in every case drops to a very low value. This would surely indicate that from the very moment when the slag becomes intermixed with the metal as described by Dr. Jay its composition changes owing to reactions in the ladle. The progress of rephosphorisation in the ladle would, therefore, appear to depend chiefly upon the rate of attack on the ladle lining. Thus Dr. Jay's data would seem to confirm conclusion No. 3 in Dr. Zea's paper (p. 492 P).

Dr. J. WHITE (General Refractories, Ltd., Worksop): These papers cover a very wide field and it is impossible to discuss all their implications. I propose, therefore, to confine myself to discussing the contribution which they make to our knowledge of the constitution of oxide systems.

In my 1943 paper * to the Institute I tried to draw up a system of phase fields for basic open-hearth slags in the same manner as Dr. Rait and Mr. Goldschmidt have done. My chief difficulty then was the lack of adequate data at certain crucial points, of which the chief were perhaps the location in the phase groups of the spinels, gehlenite and the supposed "substituted" silico-phosphates described by Schneiderhöhn.†

Regarding the spinels, a study of the literature showed that they had not apparently been identified with certainty in basic slags in spite of their tendency to crystallise in characteristic forms. Further, the phase diagram of McCaffery and his co-workers for the system $CaO–MgO–Al_2O_3–SiO_2$ indicated that dicalcium silicate and spinel would not co-exist. I therefore concluded that the spinels would probably not appear until a fairly low basicity had been reached, and this to a certain extent dictated the sequence of phases in the phase fields from the appearance of dicalcium silicate onwards. By the time the paper was presented, however, Dr. Rait had already informed me of his finding that $2CaO.Fe_2O_3$ and $MgO.Fe_2O_3$ could co-exist, and he himself referred to this point, which threw doubt on my placing of the spinels, in the discussion on the paper. This doubt was confirmed by the phase/composition relations revealed by Mason's paper ‡ when it appeared. Now the phase assemblages which Dr. Rait and Mr. Goldschmidt give for the complete slag system and also for the system $CaO–MgO–Al_2O_3–SiO_2$ (the latter con-

* White, *Journal of The Iron and Steel Institute*, 1943, No. II., p. 579 P.

† Schneiderhöhn, *Mittellungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung*, 1931, vol. 13, p. 109.

‡ Mason, *Journal of The Iron and Steel Institute*, 1944, No. II., p. 69 P.

siderably modifying McCaffery's original diagram) clearly define the position of the spinels, and show that they appear much earlier than I had suggested. We have ourselves recently prepared and X-rayed a number of synthetic mixtures in the system $CaO–MgO–Al_2O_3–SiO_2$ in connection with work on basic refractories, and our results, as far as we have gone, are in complete agreement with theirs. In particular we find that in this system spinel co-exists with the orthosilicates, with akermanite and with gehlenite, that gehlenite and MgO do not co-exist and that so-called "madisonite" does not exist—at least we got a well-developed spinel pattern with this composition. We also confirmed some time ago that $2CaO.Fe_2O_3$ and $MgO.Fe_2O_3$ co-exist in the system $CaO–MgO–Fe_2O_3$. In one connection only have I felt the need to exercise some reserve with regard to these results, and that arises from the fact that, for reasons of expedience, we heated many of our mixtures to incipient fusion, as Dr. Rait and Mr. Goldschmidt have apparently done with some of theirs. In the presence of incongruency at the melting point, this might give rise to departures from equilibrium as regards the solid state. This might also apply to furnace slags cooled from the liquid state. Nevertheless, the fact that the same relationships have been shown to hold in the more complex furnace-slag systems goes far towards removing any doubts on this point. One of the most notable features of these oxide systems now emerging is what may be called the persistence of the spinels in the phase assemblages. They occupy an intermediate position between the highly basic and the highly acid compounds and co-exist with a surprisingly large number of different phases.

Regarding gehlenite, largely on the basis of McCaffery's diagram which showed it to co-exist with MgO, I showed it as occurring in the phase groups prior to the appearance of the substituted calcium orthosilicates (monticellite, &c.), though no definite identification of it in basic slags had apparently been made, this being attributable, I considered, to the low Al_2O_3 contents of normal slags. The new data show that this order should be reversed, at least at the normal range of Al_2O_3/MgO ratios. Only with values of this ratio greater than 102 : 40 should gehlenite occur before the substituted calcium orthosilicates.

Lastly, on the basis mainly of Schneiderhöhn's examination of Hoesch furnace slags of low basicity (*loc. cit.*), I considered it likely that substitution of CaO by MgO, &c., in the silico-phosphates would precede substitution in dicalcium silicate. The new data show that nagelschmidtite, at least, co-exists with the substituted orthosilicate merwinite. The evidence of irregularities in the nagelschmidtite pattern in certain of the open-hearth slags, and the results of the experiments detailed in Table XX. of the paper by Dr. Rait and Mr. Goldschmidt, however, seem to indicate that some substitution in this compound does occur at about the composition at which substitution in dicalcium silicate occurs, though whether this will result ultimately in the formation of a new phase is not yet clear. The doubt regarding the co-existence of silico-carnotite with CaO was mentioned in the Addendum to my own paper in the light of Barrett and McCaughey's diagram which had then just been published. On the whole, the data of Dr. Rait and Mr. Goldschmidt support Barrett and McGaughey's finding.

I should like to draw attention to one very important relationship revealed by the phase assemblages and the various Tables of phases published by Dr. Rait and Mr. Goldschmidt, namely, the significance of the phase group MgO(FeO,MnO in solid solution)–$2CaO.SiO_2$–spinel. This now emerges as a critical combination in these oxide systems. Briefly, in systems in which the RO/R_2O_3 ratio is less than unity molecularly this combination occurs when the CaO/SiO_2 ratio is 2 : 1 molecularly.

If the CaO/SiO_2 ratio is greater than 2 : 1 calcium ferrites, aluminates (including brownmillerite) and chromites appear and substituted calcium orthosilicates are absent. If it is less than 2 : 1 the calcium ferrites, aluminates and chromites disappear and the substituted calcium orthosilicates appear. One can illustrate this by reference to the three quaternary diagrams shown in Fig. C. The phase assemblages in the systems shown are inherent in the phase assemblages for the more complex slag system, though all have been studied to some extent separately, the data for (*a*) being given in the present paper by Dr. Rait and Mr. Goldschmidt, while (*b*) has been studied by Dr. Rait (private communication) and (*c*)

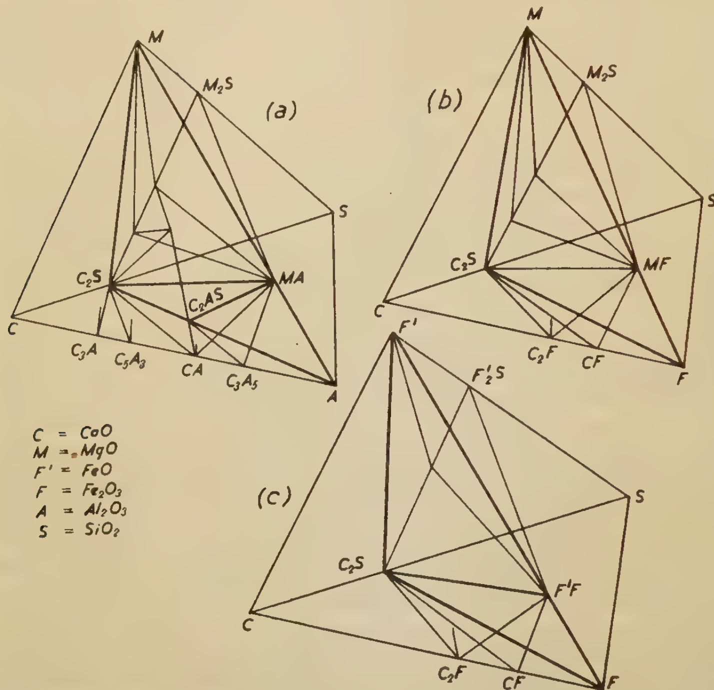

FIG. C.—Phase Assemblages in the Vicinity of the RO–$2CaO.SiO_2$–Spinel Composition Plane in the Systems (*a*) CaO–MgO–Al_2O_3–SiO_2, (*b*) CaO–MgO–Fe_2O_3–SiO_2, and (*c*) CaO–FeO–Fe_2O_3–SiO_2.

was proposed by myself in my 1943 paper as an amendment of an earlier diagram by Hay and White and is confirmed by the present data. In all three the quaternary phase space is divided into two distinct parts by the composition plane MgO(FeO)–$2CaO.SiO_2$–sesquioxide, which includes the spinel and also, in the case of (*a*), gehlenite. On the high-lime side of this plane the calcium aluminates ((*a*)) or the ferrites ((*b*) and (*c*)) occur, on the other side the substituted calcium orthosilicates occur, all co-existing with the spinel as shown by the tie lines. Diagram (*a*) also illustrates the relationships governing the occurrence of gehlenite in slags, to which reference has already been made. With the normal Al_2O_3 and MgO contents found in furnace slags the locus of the plot of composition with decreasing CaO/SiO_2 ratio must first pass through the

MgO–orthosilicate–spinel composition plane before it reaches a field in which gehlenite appears. When Al_2O_3 is high enough relative to MgO, however, gehlenite can appear in combination with $2CaO.SiO_2$, $CaO.Al_2O_3$ and spinel before any substitution in $2CaO.SiO_2$ occurs. The establishment of these surprisingly simple relationships is a very big step forward indeed and should be of immense help to all who are concerned with the constitution of either slags or refractories.

The discovery by Dr. Rait and Mr. Goldschmidt of the phase " *A* " silicate in some of their slags and the subsequent identification of it with their synthetic α′-dicalcium silicate is another contribution to our knowledge of the high-temperature forms of dicalcium silicate. I had meant

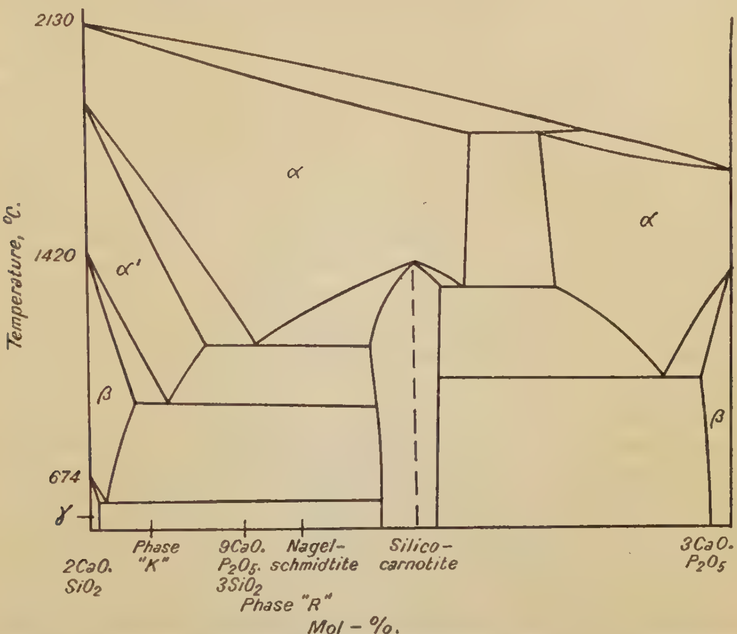

FIG. D.—Qualitative Diagram of the System $2CaO.SiO_2$–$3CaO.P_2O_5$ according to Bredig (*loc. cit.*).

to draw their attention to the very important paper by Bredig * to which Dr. Rait referred in presenting their paper. Bredig based his conclusions on a study of the published lattice-spacing data of the various silicophosphates reported and on a comparison with the behaviour of K_2SO_4 which has the same type formula (A_2XO_4) as $2CaO.SiO_2$, and exists in two forms, a high-temperature, hexagonal α form and a low-temperature, orthorhombic β form, both of which are end-members of solid solution series. He had previously shown that the lattice-spacing data of nagelschmidtite could be accounted for on the basis of a hexagonal structure similar to that of α-K_2SO_4. He now pointed out that certain other complexes of dicalcium silicate reported in the literature, *e.g.*, $Ca_{27}(SiO_4)_{12}(PO_4)_2$ and $Ca_{23}K_2(SiO_4)_{12}$, had the β-K_2SO_4 orthorhombic structure. He con-

* Bredig, *American Mineralogist*, 1943, vol. 28, p. 594.

cluded, therefore, from analogy with the solid solutions of K_2SO_4, that these complexes were really solid solutions in hexagonal and orthorhombic forms of $2CaO.SiO_2$ and predicted the existence of a stability range of the former, which he called α-$2CaO.SiO_2$, from some temperature over 1420° C. to the melting point, and of the latter, which he called α'-$2CaO.SiO_2$, from 1420° C. to the temperature of inversion to α. To show the relationships between the various forms of $2CaO.SiO_2$ and the silico-phosphate solid solutions he constructed the qualitative diagram shown in Fig. D, which accounts successfully for the various silico-phosphates reported in the literature. He also pointed out that the silico-phosphate phases, *K* and *R*, reported by Trömel, whose paper appeared while his own was in the press, could be fitted into this diagram, *K* being orthorhombic and a member of the α' series and *R*, which included nagelschmidtite and Körber and Trömel's phase $9CaO.P_2O_5.3SiO_2$ in its range, being hexagonal and a member of the α series. Further confirmation of these views has been provided by Greene * in a paper which appeared very shortly after Bredig's and was referred to by the latter. In studying the phase relations of soda in cement clinkers Greene found certain phases near to $2CaO.SiO_2$ in composition, but differing markedly from normal β-$2CaO.SiO_2$ in optical properties and X-ray pattern. In the light of Bredig's work he was able to show that they had a hexagonal structure, and were apparently solid solutions of small amounts of Na_2O and Fe_2O_3 and of Na_2O and Al_2O_3 in the hexagonal form of $2CaO.SiO_2$ proposed by Bredig. No phase corresponding to Bredig's α' form was observed, the only other form of $2CaO.SiO_2$ occurring being β. The question that arises then is to which of Bredig's proposed forms, α or α', does the α' of Dr. Rait and Mr. Goldschmidt belong. I suggest that it is of some importance that they should try to settle this point as a further contribution towards the elucidation of the somewhat complex relationships of dicalcium silicate.

Finally, there is the question of the free basic oxide phase or phases in these systems. Since my own paper was written our knowledge of the relationships between these oxides has been increased considerably, first by the investigations of Dr. Jay and Dr. Andrews on the solid solutions formed between FeO–MnO, FeO–MgO, CaO–MnO and MnO–MgO, and now by the data presented in the present papers. There are perhaps two main points that call for comment. First, there is the matter of the solid solutions between FeO and MnO. In our early work at Glasgow we had definite indications under the microscope of what appeared to be a two-phase region in this system. At the time when I wrote my own paper I had begun to doubt this, particularly in view of Whiteley's work on synthetically produced FeO–MnO inclusions in steel,† and I indicated in the paper that the appearance of a two-phase range might have been caused by a high Fe_2O_3 content in the FeO. In our work we, of course, melted the oxides, as we wished to determine the melting points in the system. We did this in molybdenum crucibles and I believe that the high Fe_2O_3 contents were mainly due to reaction with the molybdenum. In selecting metals in which to melt FeO (if iron is ruled out, as it was in this case, owing to the high temperatures involved) one has apparently the choice of two evils, as metals which form more stable oxides than FeO will generally reduce it, and most of those which form less stable oxides tend to abstract iron from the melt, thus in effect raising its oxygen or Fe_2O_3 content. Platinum, for instance, acts in the latter way, and so

* Greene, *National Bureau of Standards Journal of Research*, 1944, vol. 32, p. 1.

† Whiteley, "Seventh Report on the Heterogeneity of Steel Ingots," p. 23, *The Iron and Steel Institute*, 1937, *Special Report No.* 16.

apparently does molybdenum. We had indications of this at the time in that a certain amount of alloying at the crucible walls was apparent. Also, when we examined some of our 100% FeO specimens by X-rays (our first attempt at the use of this medium) we found that specimens that had previously been melted either in molybdenum or in platinum gave evidence of the presence of appreciable Fe_3O_4, which was either completely absent or barely detectable in the batches as prepared. At the time we considered this quite a satisfactory state of affairs, as it was such an obvious advance on previous attempts to melt iron oxide in refractory crucibles, and particularly as we intended to use the conventional approximation of treating FeO as a single component. It now appears, however, that the errors resulting were not always as inappreciable as was hoped; for example, we got low melting points in studying the system $FeO–SiO_2$, owing mainly, I think, to the presence of ternary eutectics containing Fe_3O_4.

The second point relates to the supposed incompatibility of CaO and FeO, which in the light of the present observations would appear to be in some doubt. Our finding in Glasgow that CaO and FeO did not co-exist as solid phases in the system $CaO–Fe–Fe_2O_3$, being transformed into $2CaO.Fe_2O_3$ and metallic iron, was in agreement with the findings of Kanz, Scheil and Schulz * and Martin and Vogel † and was consistent with certain features of the reduction isotherms of R. Schenck and his co-workers.‡ I therefore concluded that CaO probably would not occur as a phase in basic open-hearth slags, or at least not in the presence of the FeO-rich oxide phase, and this appeared to be supported by the fact that the identification of CaO as a phase in such slags had not been reported in the literature (except where undissolved lime occurred). It is quite evident now, however, that CaO-rich and FeO-rich oxide phases do co-exist in such slags, and one naturally wonders what has happened to the supposed incompatibility. Accepting for the present, in view of the consistency of the evidence from the different sources, that incompatibility between CaO and FeO does exist in the pure system $CaO–Fe–Fe_2O_3$, the most likely explanation would appear to be that its range of occurrence is somewhat restricted in the presence of the other basic oxides. As it happens, I drew attention to the possibility of such an effect in my own paper in discussing the system $CaO–MgO–Fe–Fe_2O_3$ and also in my reply to the discussion, when I pointed out that a certain range of MgO–FeO solid solutions might be capable of co-existing with CaO. Accepting the join $2CaO.Fe_2O_3–Fe$ in the system $CaO–Fe–Fe_2O_3$ and the complete solid solubility of FeO in MnO and of CaO in MnO as found by Dr. Jay and Dr. Andrews, it is possible to predict that a similar condition should exist in the system $CaO–MnO–Fe–Fe_2O_3$ (more correctly CaO–Mn–Fe–O, as the higher oxides of manganese will be involved at the higher oxygen contents) as shown in Fig. E (*a*), the main features of which are as follows: (1) The common solubility of MnO in CaO and FeO will cause a closing of the solubility gap between CaO and FeO with increasing MnO as shown by the curved solubility curve. The exact position of this curve is not known. (2) Assuming incompatibility to exist, there will be a composition Z in the series FeO–MnO (shown with some CaO in solution) which is in equilibrium with $2CaO.Fe_2O_3$, CaO and metallic iron with some manganese in solid solution (composition Y). The relationship between Z and Y will be determined by the equilibrium relations between

* Kanz, Scheil and Schulz, *Archiv für das Eisenhüttenwesen*, 1934–35, vol. 8, p. 67.

† Martin and Vogel, *Archiv für das Eisenhüttenwesen*, 1934-35, vol. 8, p. 249.

‡ R. Schenck, Laymann and Jenckel, *Zeitschrift für anorganische Chemie*, 1938, vol. 206, p. 313.

FeO–MnO and Fe–Mn solid solutions, always allowing for some modification due to CaO in solution. This equilibrium does not appear to have been determined for the solids, but is given for the liquid solutions by the slag-metal constant K_{Mn}. Hence, if the solid-phase equilibrium does not differ much from that between the liquid solutions at the freezing point, the conjugation lines between the two solid solutions will run approximately as shown by the dotted lines and the composition of Z will be conditioned by the concentration of manganese at which Fe–Mn

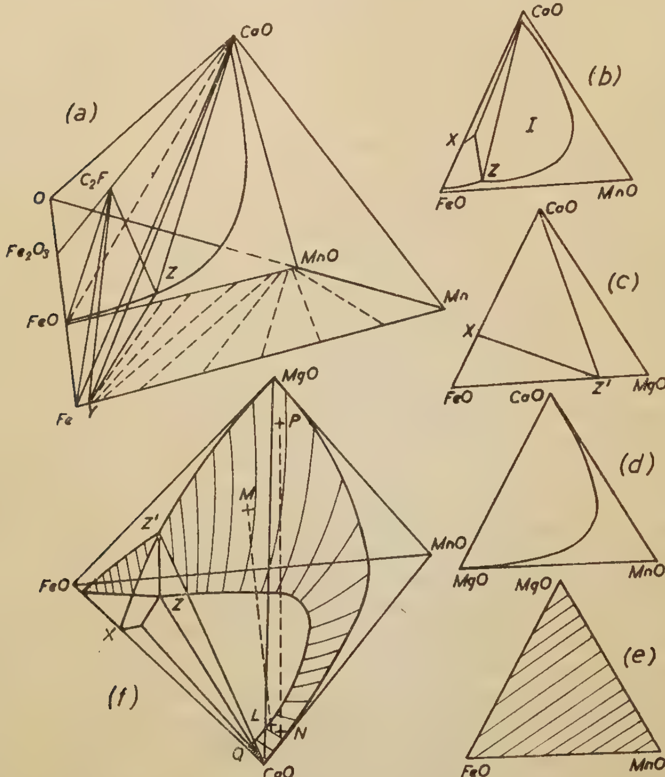

FIG. E.—Phase Relationships (Tentative) in the Systems (*a*) CaO–Mn–Fe–O, (*b*) CaO–FeO–MnO, (*c*) CaO–MgO–FeO, (*d*) CaO–MgO–MnO, (*e*) MgO–MnO–FeO, and (*f*) CaO–MgO–MnO–FeO.

alloys become reducing towards $2CaO.Fe_2O_3$. It seems likely that this concentration may be small, so that Z may lie fairly near to the FeO end of the FeO–MnO series. Fig. E(*b*) shows the phase relations in the plane CaO–FeO–MnO of this diagram with compatibility between CaO–MnO and FeO–MnO solid solutions occurring in field I. The three other ternary diagrams of the four oxides are given in (*c*), (*d*) and (*e*). In (*d*) the exact position of the curved solubility line is not known, but we recently found that an equimolecular mixture of CaO and MgO with 60% of MnO by weight was single-phase. On the basis of the four ternary

diagrams, the phase diagram of the system formed by the four oxides shown in (*f*) can be constructed. In it the effects of the incompatibility of CaO and FeO are manifested to the left of plane *ZZ′Q* only. To the right of this plane there is compatibility between pairs of conjugate solutions. According to the lattice constants of the CaO-rich and FeO-rich phases given in the present papers the compositions of these phases in slags will lie in somewhat restricted regions round *L* and *M*, the direction of the conjugation lines being roughly parallel to *LM*. In partly slagged dolomite bricks which we have recently studied the conjugate phases, according to their lattice dimensions, lie in the region of *N* and *P*. The exact locations of *Z* and *Z′* are, of course, unknown, but *ZZ′Q* must be to the left of *LM*. *Z* may be close to the FeO–CaO edge of the diagram and *Z′* nearer to the MgO corner. It must be understood, of course, that the proposed diagrams are tentative and are put forward mainly to indicate points for future investigation.

It will be noted that in constructing the diagrams I have indicated some degree of mutual solubility of CaO and FeO. Indications that there is some solubility of CaO in FeO have been given by Martin and Vogel (*loc. cit.*), and this can also be deduced from the work of R. Schenck (*loc. cit.*). I am not aware that any diagram showing solubility of FeO in CaO has been given, but Dr. Jay has been expressing this opinion for some time now (private communication). We have recently found indications that point to the same conclusion. This does not, of course, necessarily affect the question of the existence of the $2CaO.Fe_2O_3$–Fe join in the system CaO–Fe–Fe_2O_3.

JOINT CORRESPONDENCE.

Mr. J. E. Pluck (Messrs. Steel, Peech and Tozer, Sheffield) wrote: In the past, we have tried different types of basic bricks in ladle linings with a view to increasing lining life. The basic ladle to which Dr. Zea referred in his paper was the first in which we had used a full lining of stabilised dolomite bricks.

Apart altogether from the metallurgical results, we obtained an increased life of 100% over the normal firebrick lining. At the same time, the cost of ladle refractories per ton of steel produced was twice that of the normal practice.

The main trouble was the accumulation of slag and metal growth on the dolomite lining, thereby reducing the volume of the ladle. The removal of this growth damaged the bricks and required their replacement. If the ladle had been 10% larger than normally necessary, the growth need not have been removed and the life of the lining would have been considerably increased, the cost per ton being brought down proportionately.

Perhaps the use of insulation at the back of the basic-brick lining would lower the overall conductivity. There is little doubt that, if this growth can be avoided, this type of lining will show a vast improvement over firebrick.

Mr. C. W. J. Crawford (Messrs. John Summer & Sons, Ltd., Shotton, Chester) wrote: The authors are to be congratulated on their excellent and profound papers. I have used Schenck and Riess's method of phosphorus calculation and have been so impressed by results that I have felt it safe to use the theoretical assumptions underlying it as a basis for understanding slag–steel phenomena. I congratulate Dr. Zea on the thorough-

ness of his investigation and admit that I am pleased with his findings. However, the other papers dealing with slag constitution are equally commendable, and these appear to put a large question mark against some of the assumptions made by Schenck and embodied in his phosphorus calculation. So I should like with all deference to ask the authors with what degree of certainty their experience in the X-ray examination of slag structures would allow them to say that their finding that phosphorus is not present in solid normal basic open-hearth slag as either tricalcium phosphate or tetracalcium phosphate but chiefly as lime-silica phosphates or fluorapatite, is true of the same slag in its furnace condition ?

Mr. H. J. GOLDSCHMIDT (Messrs. William Jessop & Sons, Ltd., Sheffield) wrote : I wish to congratulate Dr. Jay on his excellent paper, and hope the following few remarks, which are mainly concerned with X-ray or structure points, will not be taken as in any way detracting from its great value. I was pleased to see that wherever our papers overlap,* namely, in regard to basic open-hearth slags, there is substantial agreement of results. This refers particularly to the prominence of nagelschmidtite and fluorapatite and the findings on the CaO- and FeO-rich solid solutions.

Regarding the evidence for fluorapatite taking silicon into solution, I wonder whether Dr. Jay considers his "silico-fluorapatite" to be definitely established. In our slags the measured lattice dimensions practically agreed with those of the pure compound. The ability of silicon to substitute phosphorus in apatite is, however, not denied. Indeed, McConnell † reported a wide isomorphous replacement of elements in the apatite structure ($Ca_{10}F_2P_2O_{24}$) as follows :

Ca by Mn, Mg, Na, K, Sr and C (?).
P by Si, S, V, C, As.
F by Cl, O, OH.
O by F, OH.

In basic slags, the most important substitutions are therefore those by manganese, magnesium, silicon and carbon, which are all conceivable; but until these possibilities have been explored it may perhaps be safest to refer to the phase as fluorapatite.

In this connection it is interesting that in his experiments of waiting 2, 8 and 20 sec. before quenching the spoon sample Dr. Jay finds for the CaO-rich phase a rise in spacing with time of waiting; this will be due to a loss of MnO from CaO solid solution, which, in the present light, might have entered apatite. It may be significant that the slowly cooled samples, which gave a parameter much nearer that of lime, actually showed this apatite.

Regarding the same experiment, Dr. Jay suggests that the decrease of nagelschmidtite is due to a reaction between this phase and the liquid. Is it not more probably due to the fact that the lower-melting fluorapatite is given an opportunity to crystallise out already in the spoon during cooling before quenching, though it may not yet be in sufficient amount to be visible ?

I was pleased to find such good general agreement between our parameter values in CaO- and FeO-rich solid solutions (*cf.* Dr. Jay's Tables IV. to IX. and our Table XIX. ‡). At the moment, however, it does not seem possible to give a practical interpretation to these values without

* Rait and Goldschmidt, *Journal of The Iron and Steel Institute*, 1945, No. I., p. 391 P (this volume).
† D. McConnell, *American Mineralogist*, 1938, vol. 23, p. 1
‡ Rait and Goldschmidt, *loc. cit.*

ambiguity, in view of the presence of MgO as well as MnO in the FeO solid solution. It is interesting that he also finds (as we do in some cases) the two distinct FeO-rich phases of spacings in the orders of 4·34 and 4·26 Å. Does he think this split is due to a slower rate of cooling, allowing the phases to separate on entering a low-temperature immiscibility region?

Dr. Jay proves convincingly by X-rays the correctness of the equations:

$$2FeO.SiO_2 + 2CaO \rightarrow 2CaO.SiO_2 + 2FeO$$
$$\text{and } 2MnO.SiO_2 + 2CaO \rightarrow 2CaO.SiO_2 + 2MnO,$$

but I do not think that this proof is quite relevant to basic slags, the compositions of which are so lime-rich that fayalite and tephroite do not enter into consideration in the part of the phase diagram concerned. Further, would Dr. Jay not agree that the fact that dicalcium silicate is formed in basic slags preferentially to ferrous or manganous silicate, around which his proof centres, has long been known and utilised in steelmaking control? The heats of formation are:

$$2FeO + SiO_2 = 2FeO.SiO_2 + 8100 \text{ cal.}$$
$$\text{and } 2CaO + SiO_2 = 2CaO.SiO_2 + 28{,}400 \text{ cal.},$$

and the melting point of $2CaO.SiO_2$ (2120° C.) is so much higher that it is bound to solidify first from the liquid, especially at the lime-rich compositions concerned. Nevertheless the proof is valuable, though I frankly think that it is only confirming a long-established fact.

In Appendix II., where Dr. Jay further discusses reactions between lime and ferrous and manganous silicates, he gives details of materials and conditions used in preparing the compounds, but the initial relative amounts of oxides are left quite open; thus, on the evidence given, the result of obtaining the olivine structure ($2RO.SiO_2$) seems rather fortuitous; pyroxenes ($RO.SiO_2$) might equally well have been formed, depending on the proportions of FeO(MnO) and SiO_2 used. Similarly, the initial proportions of lime to ferrous silicate would be important to know, as the final CaO/FeO ratio in the silicate produced arbitrarily depends on these amounts.

In the same connection Dr. Jay mentions that, after firing the lime and ferrous silicate, two "olivines" are observed, one being β-$2CaO.SiO_2$, the other $(CaO)_{1.3}(FeO)_{0.7}.SiO_2$. However, is it correct to call β-$2CaO.SiO_2$ an olivine? Its actual structure has not yet, I believe, been analysed, but it certainly differs from that of the olivine family.

I should be interested to know whether Dr. Jay actually finds a single-phase compound at the composition $(CaO)_{1.3}(FeO)_{0.7}.SiO_2$. Are the suffixes intended to be exact? Actually, in a vain attempt of ours to produce "ferrous merwinite," $3CaO.FeO.2SiO_2$, we produced a monticellite analogue, $CaO.FeO.SiO_2$. This co-existed with β-$2CaO.SiO_2$, and it would be valuable to find out whether Dr. Jay's compound perhaps corresponds to an actual ferrous merwinite.

Regarding the mixtures of lime and $2MnO.SiO_2$, it would be interesting if Dr. Jay could confirm our "manganese merwinite," $3CaO.MnO.2SiO_2$.

If I may raise a point of atomic structure, could Dr. Jay please elaborate his conception of CaO replacing FeO and MnO in their silicates "molecule by molecule?" As molecules are non-existent in silicates, I think the term "replacement ion by ion" might have been better, if this is meant. Further, true replacement is conceivable only in isomorphous substances, whilst actually the $2CaO.SiO_2$–$2FeO.SiO_2$ series forms compounds of quite different structures. Does Dr. Jay simply imply by "molecular replacement" that the type formula $2RO.SiO_2$ is being maintained, *i.e.*, that $2CaO.SiO_2$–$2FeO.SiO_2$ (or $2MnO.SiO_2$) forms in fact a binary system?

Reverting to nagelschmidtite, it is interesting that Dr. Jay notes a new phase between this and β-$2CaO.SiO_2$. This may be the same as our α'-$2CaO.SiO_2$, or Trömel's K phase. If this is so (which could be readily verified by comparing our patterns), the phase probably corresponds to a high-temperature solid solution in $2CaO.SiO_2$.

One point on which Dr. Jay's and our results do not seem to tally is that tetracalcium phosphate is not found in his slags, whilst the phase occurs in some of ours, though quite probably in a metastable condition.

As regards calcium ferrite ($CaO.Fe_2O_3$), Dr. Jay states that it is not present, but, in view of the strong basicity, is this at all a phase liable to occur? On the other hand, I think that $2CaO.Fe_2O_3$ might have been considered instead, as this is the more likely constituent. Dicalcium ferrite appeared in fact as a subsidiary phase in our basic open-hearth slags, though Dr. Jay's samples perhaps cooled too rapidly for it to be formed.

There are two final items on which I should appreciate his elucidation.

Regarding the slag quenched from the liquid (p. 528 P), a list of three constituents is given, including a non-crystalline phase, *i.e.*, a supercooled liquid, but the list excludes an FeO-rich phase, (Fe,Mn)O. Is it implied that the latter is not present as a crystalline phase and that in fact it is latent in the glassy material?

In experiment 7(*b*), p. 531 P, Dr. Jay asks the question whether ferrous or ferric iron is associated with lime-silica phosphates, but the X-ray result that lime and nagelschmidtite are formed does not seem to answer it, as neither phase contains iron, or is it implied that the ferric iron (appearing in the analysis) is dissolved in the nagelschmidtite phase?

I should much value Dr. Jay's comments on these points of query, which, I trust, will not hide the fact of good agreement between our investigations in all essentials. His paper forms a most useful contribution, not least because it furnishes another example for the successful application of X-ray analysis to an industrial problem.

Mr. S. W. Pearson (Messrs. Steel, Peech and Tozer, Sheffield) wrote: In these days the control of phosphorus in basic open-hearth steelmaking does not provide any real difficulty to the practical man. However, the economics of slag-making, especially where high-phosphorus metal has to be used, is always in the mind of the open-hearth manager. Furthermore, the retention in the slag of the phosphorus burden with the minimum degree of oxidation is considered to be a very important factor affecting the quality of fully killed steel. Therefore, any assistance from our more technical friends is both welcome and acceptable.

The papers by Dr. Zea and Dr. Jay on this subject are valuable contributions, although entirely different in the approach to the problem.

The painstaking and ingenious manner in which Dr. Zea has employed his data gives ample confirmation of the known factors that influence the phosphorus reactions and suggests a numerical value for each factor. It is suggested to the author that the approach to equilibrium of other phases such as gas/slag, slag/hearth, metal/hearth, slag-and-metal/ladle-lining is bound to influence the state of equilibrium existing between the slag and metal, so that it is difficult to envisage any numerical values of phosphorus in the metal for any set of conditions being anything but widely approximate.

Dr. Jay, on the other hand, stressing the lack of finality in his work, seems to be intentionally provocative in his treatment of the subject—no doubt in order to focus discussion and further thought on the matter. His inference that in order to minimise rephosphorisation one should increase the silica burden (within certain limits) will be strongly doubted

by many practical steelmakers, although certain evidence is produced in support. This does show the necessity for further work to be undertaken before these views can be entirely acceptable. The significance of Table X. in Dr. Jay's paper seems to be quite obscure, and it is seriously suggested that it is carbon and not manganese that is the influencing factor in this series of results.

It is the opinion of the writer that the part of Dr. Zea's paper dealing with the basic-lined ladle will prove to be most profitable to the steel industry in the future. It was interesting to take part in experiments designed to reproduce deliberately conditions known to promote rephosphorisation in ordinary circumstances, and find only a negligible amount taking place. The ceramics engineers should give further early attention to the problem of providing a brick that will fulfil these requirements with an improved physical life. One wonders how much higher the silica could be taken in the manufacture of this type of brick. A relatively lower refractoriness would be no disadvantage, the thermal conductivity might be improved, and at much higher levels the silica would remain non-reactive with basic slag.

Mr. W. Geary (Appleby-Frodingham Steel Co., Ltd., Scunthorpe) wrote : These three papers form a most interesting group, because, although they are so nearly allied in subject matter, they are dissimilar both in scope and in treatment.

First I should like to congratulate Dr. Rait and Mr. Goldschmidt on the effective way in which they have carried out such an extensive survey of slag constitution. One perceives in the present paper a correlation of information that has become available since the publication of White's paper in 1943,* as well as an offering of new work based on the use of the X-ray powder method. The proposal to use an X-ray method of slag control is interesting, and it is to be hoped that the authors will be able to give more information as their experimental work progresses.

In considering Dr. Zea's paper, I must express my admiration of the careful and painstaking investigation that it records. His conclusion that rephosphorisation is related to the working temperature, slag composition and fluxing of the ladle lining will doubtless be accepted on the majority of melting floors. Dr. Zea's experiments with basic-lined ladles, as described in Appendix B, show clearly that the influence of the ladle lining overrides the other factors, since, with the deliberate use of slags that would normally have permitted pronounced rephosphorisation, none was observed when basic ladle lining was used. This being the case, caution in the use of temperature, and care in the adjustment of slag composition must be merely palliatives, to be used until a satisfactory basic lining is established on a practical and commercial foundation. At the present time the inducements to a shop manager to continue to use fireclay bricks are many, and where the melting practice is conducive to rephosphorisation the operator must take advantage of such expedients as are open to him. Dr. Zea, in paragraph (4) of his Summary and Conclusions, states that reduction of rephosphorisation may be obtained (circumstances permitting) by increasing the lime content of the slag. I should be interested to know whether Dr. Zea suggests the addition of more lime to the slag in the furnace only, or whether he suggests that it should be added to the slag in the casting ladle.

Dr. Jay's paper, dealing as it does with both slag constitution and rephosphorisation of steel, provides the link which draws the three papers into a group. I venture to suggest that it may also prove to be an essential

* *Journal of The Iron and Steel Institute*, 1943, No. II., p. 579 P.

link with the future. Dr. Jay's X-ray work on basic slags (avowedly undertaken at the request of Dr. Zea) has led to the propounding of a revolutionary hypothesis concerning the constitution of molten basic slags, the hypothesis following the use of a new technique of examination.

The suggestion that molten slag contains two distinct solid phases suspended in a liquid phase may find neither ready nor universal acceptance. We have become used to conceiving the slag as a liquid in which we dissolve this or that reagent according to the reaction that we wish to promote; yet there have been puzzling features about the behaviour of the liquid. It seems to me that Dr. Jay's hypothesis that the " molten " slag is a suspension starts to provide a solution to some of the puzzles.

The melter for several decades has worked his charge through his perception of the slag. Modern methods of slag control, admittedly, have put upon the scientist the onus of achieving a desired slag composition; yet the melter, and indeed the shop manager, still has more than a passing interest in the physical condition of his slag. The desirable physical feature of most finishing slags is " creaminess." But what is creaminess? The viscosity of slags has been measured; but is it considered that they have a viscosity in the same way that a true liquid has? I suggest that their appearance in the furnace and when poured or otherwise manipulated agrees much more nearly with the assumption that they are suspensions than that they are liquids, and that increases in measurable viscosity are merely caused by increases in the amount of solid matter being carried by a given amount of liquid.

If we now consider the action of magnesia on molten basic slag we shall find that a small increase of MgO, expressed as a percentage of the slag composition, has a great influence in increasing the viscosity of the slag. The melter may refer to such a slag as being " leathery." It has hitherto been a mystery why the effect should be so pronounced, but if one accepted Dr. Jay's hypothesis the solution appears. The MgO dissolves in the liquid phase and increases the viscosity of the liquid portion of the slag upon which the fluidity of the whole depends.

It seems to me that Dr. Jay's conception will appeal strongly to practical steelmakers, who will await with interest the outcome of the further work which must inevitably follow.

Considering now the technique by which Dr. Jay's results were reached, I find myself in the position of having to accept one half while perhaps understanding the other. X-ray examination is such a special branch of science that the ordinary person is in the hands of the experts: but the quenching of samples to retain something of a state that exists at high temperatures is a technique with which one is familiar. It is this technique of quenching that has enabled Dr. Jay to put forward his theory, but the quenching technique has been so fully productive only because Dr. Jay had the new instrument to exploit it. Is there not an analogy between the great days of metallography and the present time? It seems to me that, just as the science of metals expanded under the developing technique of heat-treatment and the metallurgical microscope, so the science of slags may grow by means of heat-treatment (including all such processes as are applied to metals) and the use of X-ray apparatus.

Assuming that further work confirms Dr. Jay's present technique and conclusions, it is quite possible that to-day we are on the edge of the long-desired field of knowledge concerning the constitution of slags at steelmaking temperatures.

Mr. A. Preece (Leeds University) wrote: The papers under discussion represent a considerable step forward in our attempt to elucidate the constitution of open-hearth slags.

The all-important property would seem to be the oxidising character of the slag in so far as it governs the oxygen content of the metal or some fundamental property which is indicated by the oxygen content. Some recent experimental results have suggested that the overheating tendencies of a steel are in some way governed by the original oxygen content of the metal.

Dr. Zea deserves high praise for his very careful analysis of the information provided by a limited number of casts. From my contact with this work while Dr. Zea was working for his higher degree at Leeds University, I think it might be stressed that the paper does not indicate the vast amount of work involved.

Dr. Zea has chosen his index of the rephosphorisation by observation; it is therefore empirical and must be established for each set of steelmaking conditions. For these reasons its usefulness does not depend on the accuracy of Schenck's assumptions.

I would ask Dr. Jay and Dr. Rait whether they have considered the possibility of using the method of differential thermal analysis in elucidating the slag constitution. This method is now being used very successfully in determining the constitution of clay minerals.

AUTHORS' REPLIES.

Dr. Rait and Mr. Goldschmidt wrote in reply: Mr. Robertson has asked for a set of equations for the calculation of the slag–metal relationships. We will certainly give our attention to Mr. Robertson's request, but must disappoint him by pointing out that much fundamental work lies ahead. For many years the practical basic steel melter has realized the importance of the function of the slag in his process, and many attempts have been made to account for the slag–metal reactions. The practical steel melter is not so much concerned with the type of complex compounds present in basic slags as in having a simple direct method of slag control to ensure that the slag fulfils its refining function, and that there is no return of phosphorus to the steel. Unfortunately, there is no easy approach to the complete understanding of the slag–metal reactions, and accordingly the practical melter must be patient with his technical colleague who is striving to accumulate the necessary information on basic slags. Schenck made an outstanding attempt to account for these reactions and proposed certain equations based on the existing phase equilibria data on slag systems in the solid state, on certain assumptions regarding the nature of the compounds occurring in the liquid slag, and on experiments carried out on basic open-hearth furnaces. These equations which purport to account for slag–metal reactions are open to the following grave criticisms:

(*a*) The phase equilibria data employed by Schenck were incomplete and, in certain cases, inaccurate.

(*b*) There is little real evidence to substantiate the existence of the compounds assumed to be present in the liquid slag, and the important silico-phosphates which the work of Dr. Jay and of ourselves indicates to be extremely stable are overlooked.

(*c*) The results derived from the practical experiments carried out on basic furnaces are probably only applicable within the range of conditions encountered in Schenck's experiments, and extrapolation of the equations to conditions outside this range is extremely dangerous.

It is believed necessary to understand thoroughly the constitution of the slags in the solid state and to acquire more knowledge regarding the

stability of compounds in the liquid state before equations can be produced to account for the slag reactions under universal conditions. The recent comprehensive review on basic slags by White indicated the gaps in our knowledge of the constitution of slags in the solid state, and, although the present work has filled in many of these gaps, much information is still required. It is therefore not unexpected that as yet no really satisfactory correlation between the various equations proposed for the phosphorus reactions and actual practice has been found.

In conclusion we would like to express our thanks for the complimentary comments made about our investigation.

Dr. Zea wrote in reply: I am grateful to the many people who have contributed to the discussion either verbally or by correspondence.

Mr. Robinson's remarks are noted. The casts examined were taken at random from normal works practice, and in order to make certain that equilibrium between the slag and the metal was attained, only final refining and tapping samples were used in calculating the results shown in Tables I., II., III., and VI. In the fixed-furnace open-hearth process the phosphorus content of the metal towards the end of the refining period is usually of the order shown by the casts examined; I therefore take it that Mr. Robinson is referring to steel made in tilting furnaces, where there is less danger of ladle rephosphorization, and hence the phosphorus content of the steel leaving the furnace need not be reduced to the same extent as in fixed-furnace practice.

I thank Mr. Robertson for pointing out those instances where Schenck's phosphorus equation does not give satisfactory results. In so far as those melted samples are concerned, I share Mr. Robertson's views that the more viscous slag and the lack of sufficient agitation of the bath at the stage before it has attained an open boil both tend to hinder the attainment of complete equilibrium between the slag and the metal—a condition which is necessary for the physico-chemical laws to operate.

The slag and metal samples in the Hoesch process cited by Mr. Robertson had the following analyses:

Slag Analysis.				Metal Analysis.	
Fe	5·70%	P_2O_5	23·90%	Carbon	1·80%
CaO	43·00%	MgO	4·40%	Phosphorus	0·30%
SiO_2	12·80%	Al_2O_3	1·10%	Manganese	0·36%
MnO	6·31%				

The corresponding bath temperature was not given; but, calculating on the basis of the work of Roeser and Wensel,* a metal of the said composition would have an initial freezing point of 1393·5° C. Assuming a superheat of 80° C., the corresponding bath temperature would be in the neighbourhood of 1475° C., in which case, according to Schenck's phosphorus equation, the metal corresponding to the slag in question would at equilibrium contain 2·3% of phosphorus instead of 0·30% as shown by analysis. Even assuming the bath temperature to be 1400° C., below which the metal would commence freezing, the calculated phosphorus content would be no less than 0·63%. It is, however, noted that both the temperature and the P_2O_5 content of the slag in this example were very much outside the range of those which correspond to the casts on which Schenck based his equilibrium diagrams and his empirical formula for phosphorus calculations. In the absence of precise knowledge concerning the conditions under which the said samples were taken, it is difficult to assess to what extent, if any, the accuracy of Schenck's formula may be affected by the high P_2O_5 content of the slag.

* *Journal of Research of the National Bureau of Standards*, 1941, vol. 26, pp. 273–287.

With regard to Mr. Robertson's suggestion, I hope to extend this study to cover a wider range of slag and steel compositions in the future when opportunities are available.

I thank Mr. Manterfield for his statement testifying to the practical applicability of using the ratio of the free CaO to the free FeO to estimate ladle rephosphorization.

In reply to Mr. Oliver, the immersion temperature readings were used in all the calculations without correction for the probable error in Schenck's temperatures. In the early stages of this work, experiments were carried out to compare the immersion temperature with optical-pyrometer readings. An instrument of the disappearing-filament type was used and the procedure described by Schenck and Brüggemann * was followed—that is, the " dark point " of the metal was taken with the instrument focused on the lip of the spoon while pouring. The measured temperature was corrected by adding 25° C. for cooling in the spoon, and the figure thus obtained was further corrected by an emissivity factor of 0·5 to black-body conditions. The results of these experiments showed the optical-pyrometer readings to be higher by 35° C. on the average. This figure agrees closely with the value reckoned by Mr. Oliver, although he did not state which method gave the higher readings. The introduction of this temperature correction in the application of Schenck's phosphorus equation would result in a slight increase in the calculated values for phosphorus. As in many instances the calculated phosphorus contents are lower than those given by chemical analysis, a correction for temperature as such would bring the calculated figures into better agreement with the analytical results.

It will be seen in Table D that the omission of such a temperature correction had not affected—at least, not to any appreciable amount—the general results of the samples dealt with in the paper. In my opinion, Schenck's equation is only an empirical formula, and the accuracy of optical temperature measurements depends largely on the technique and experience of the operator; I therefore do not consider that there is sufficient justification to introduce such a temperature correction in these calculations without further investigation.

I thank Dr. Whiteley for his constructive criticism. He is quite correct in saying that both free CaO and free FeO aid in dephosphorization, but that does not necessarily mean that these compounds should not be set against each other in the treatment of rephosphorization in the ladle, where conditions are quite different from those prevailing in the furnace. It is true that both lime and iron oxide attack the firebrick lining of the ladle, but, other things being equal, the severity of attack would depend upon the relative amounts of these two compounds present. Furthermore, the relative amounts of lime and iron oxide in the slag, among other factors, would have an effect in determining its viscosity, the importance of which on rephosphorization has been emphasized by Dr. Whiteley in his discussion. With reference to the curve in Fig. 30, its position is drawn in such a way that due importance is given to the points corresponding to appreciable rephosphorization. A rephosphorization of 0·002% is indeed negligible, bearing in mind that the figure represents the difference in the phosphorus contents of two samples, and is therefore well within the permissible errors in chemical analysis. Since dephosphorization of the metal does not take place in the ladle, it is reasonable to draw the curve to pass through the two points, on the left-hand side, which represent the minimum values of the ratio free-CaO/free-FeO that correspond to zero or negligible rephosphorization, whereas the three points which do not fall on the curve may be considered to include some safety factor in regard to rephosphorization.

* *Archiv für das Eisenhüttenwesen*, 1936, vol. 9, pp. 543–553.

Table D.—*Deviation in Phosphorus Content of Metal as Determined by Chemical Analysis and as Calculated from Schenck and Riess's Work,* (a) *without Temperature Correction and* (b) *with a Temperature Correction of* 35° C. *added to the Immersion Readings.*

Cast No.	[P] Analysed, %.	(*a*) Without Temperature Correction.		(*b*) 35° C. Added to Immersion Temperature.	
		[P] Cal., %.	Discrepancy, % P.	[P] Cal., %.	Discrepancy, % P.
		Sampled during Refining.			
34/2238	0·020	0·015$_5$	0·004$_5$	0·020$_5$	0·000$_5$
30/4600	0·025	0·015	0·010	0·017	0·008
31/501	0·018	0·020$_5$	0·002$_5$	0·028$_5$	0·010$_5$
31/508	0·012	0·015	0·003	0·020$_5$	0·008$_5$
35/7148	0·024	0·019	0·005	0·022$_5$	0·001$_5$
35/7156	0·030	0·012	0·018	0·014	0·016
31/599	0·023	0·012	0·011	0·016$_5$	0·006$_5$
31/572	0·024	0·020	0·004	0·025	0·001
30/4591	0·031	0·028	0·003	0·034	0·003
25/1518	0·023	0·021	0·002	0·028	0·005
25/1521	0·016	0·017	0·001	0·021	0·005
23/3736	0·021	0·020	0·001	0·026$_5$	0·005$_5$
25/1527	0·029	0·036	0·007	0·049	0·020
25/1528	0·018	0·024	0·006	0·028	0·010
22/2309	0·022	0·016	0·006	0·021	0·001
		Sampled on Tapping.			
34/2238	0·018	0·014	0·004	0·017$_5$	0·000$_5$
30/4600	0·020	0·018	0·002	0·020	Nil
31/501	0·021	0·020	0·001	0·025$_5$	0·004$_5$
31/508	0·016	0·015	0·001	0·018	0·002
35/7148	0·027	0·019	0·008	0·025	0·002
35/7156	0·016	0·010	0·006	0·010	0·006
31/599	0·011	0·012	0·001	0·013	0·002
31/572	0·022	0·018	0·004	0·022$_5$	0·000$_5$
25/1518	0·028	0·020	0·008	0·025$_5$	0·002$_5$
25/1521	0·018	0·017	0·001	0·021	0·003
23/3736	0·022	0·020	0·002	0·025$_5$	0·003$_5$
25/1527	0·039	0·038	0·001	0·054	0·015
25/1528	0·018	0·025	0·007	0·029	0·011
22/2309	0·027	0·018	0·009	0·021	0·006

	Calculated Phosphorus Content, %.	
	(*a*) Uncorrected.	(*b*) Corrected.
Discrepancy : Maximum	0·018	0·020
Minimum	0·001	Nil
Mean	0·004$_8$	0·005$_5$
Standard deviation	0·003$_8$	0·005

In calculating the amount of ladle bricks eroded per cast, Dr. Whiteley had assumed a slag weight of 10 tons. This figure actually varied from 1 to 3 tons in the casts examined. Also, Dr. Whiteley had not taken into account in his calculations the effect of the preferential attack by the slag on the free silica in the firebrick. Whilst this is difficult to assess quantitatively, it is evident that preferential attack does occur, as indicated in the experiments referred to on p. 489 P of this paper. This may well account for the difference between the SiO_2 content of the final ladle slag as calculated by Dr. Whiteley and the analytical figure.

In reply to Mr. Geary, the lime must be properly fluxed and taken into solution by the slag, and would therefore have to be added in the furnace.

I thank Mr. Pluck and Mr. Pearson for their suggestions concerning methods to overcome the present difficulties in the successful use of basic-lined ladles. It would be extremely interesting if full-scale trials could be made along these lines.

I would also like to thank Mr. Preece for his appreciative remarks.

Dr. Jay wrote in reply : Dr. Desch states that the mineralogical constitution of a spoon-cooled or a slowly cooled slag can give only indirect information of the slag condition in the furnace. As far as X-ray work is concerned, I agree with him, and for that reason I examined water-quenched slags. It is possible that microscopic examination of spoon-cooled slag samples will provide additional information.

I am pleased to have Mr. Robinson's comments on the influence of the variables phosphorus pentoxide, silica, slag iron, and manganese oxide. The citric-acid-solubility tests are very interesting, and one hopes for a test to separate for subsequent examination the solid and liquid components existing in furnace slag. My opinion regarding " activity " of the slag is that this is governed by the nature, fluidity, and chemical analysis of the liquid phase. The solid phases, fluxed lime and a lime silico-phosphate, are potentially active—that is to say, they can and will take part in reactions when a change takes place in the slag, owing to variation of temperature, additions made to the charge in the furnace, and corrosion of the refractory lining of the ladle. With regard to Mr. Robinson's last remarks, my opinion is that for the steelmaking practice studied in this paper the silica content of a slag is important in providing a " well fluxed slag."

Replying to Dr. Fairley, may I make a considered statement on the problem of rephosphorization of pit-metal samples. The final ladle slag, being the drainings from the ladle nozzle after the last metal has drained away, cannot be taken as representative of the bulk of the slag on the metal during the pouring of the ingots. If any correlation is desired between this slag and metal, then the metal must be the last metal. This last metal is not under examination in this study of rephosphorization, for, if Dr. Fairley will note Fig. 9, the " condition of the tapping slag " is only related to the period " furnace to fourth pit." It is, of course, true that the marked rephosphorization of the last ingot or last metal is brought about by the loss of basicity of the slag on the metal in the ladle. Dr. Fairley's last remark regarding freedom from serious rephosphorization when the basicity and oxidizing power are adequate is true. The paper might be described as a study of rephosphorization of pit samples when the furnace slag conditions vary from the ideal.

Mr. Robertson is anxious about the use of the terms " free lime " and " free iron oxide." My use of the term " free lime " implies uncombined lime in a finely dispersed form. Thus, lump lime is not considered as part of the slag. This free lime is easily identified on the X-ray photograph. Its precise nature is not that of pure lime but rather of lime in

which other oxides, such as manganous oxide and ferrous oxide, have dissolved without changing the crystalline nature of the lime. An analogy in metallurgy would be an iron-chromium (10% chromium) alloy where the chromium had dissolved in the iron and the resultant alloy was isomorphous with iron. An X-ray photograph cannot determine the presence of free iron oxide in the furnace slag; the iron-oxide (FeO) phase only makes its appearance by crystallization in a slowly cooled slag.

Mr. Manterfield's remark about the effect of turbulence during the filling of the ladle prompts me to say that this turbulence—mixing of some slag with some metal—provides a condition which was not operative in the furnace before tapping and accelerates any change which is consistent with late additions to the bath. I was already aware of the behaviour of the casts examined by Mr. Manterfield. Those data enabled me to discuss slag behaviour under conditions which are, in my opinion, rather abnormal. The metal analyses were 0·17% carbon, 0·7% manganese, for cast 23/4395 and 0·4–0·6% carbon, 0·7–0·8% manganese for the remaining casts. The low-carbon cast behaved as expected; the medium-carbon casts behaved irregularly and apparently in conflict with the views expressed in this paper. A more detailed study of the latter casts shows that the slag samples covered a wide range of composition and illustrates many of the points discussed in the text. I have made use of a triangular plot of CaO–SiO_2–P_2O_5 for a fuller explanation of the slag–metal behaviour than is possible by the simple treatment given in the paper. This triangular plot makes possible a better correlation with X-ray data *in respect of the amount of free lime present, viz.*: Cast 22/3250, lime, trace; Cast 22/3279, lime, small amount (less than usual); and Cast 22/3249, lime, large amount (greatly in excess of usual practice). The results on the seven casts are summarized in Table E. These casts fully illustrate the view that there is

TABLE E.—*Characteristics of Seven Slags.*

Cast No.	Free Lime Amount.	Basicity.	SiO_2/P_2O_5 Ratio.	Remarks.	
				Dephosphorization.	Rephosphorization.
22/3250	Trace.	Too low.	1·1 low.	Poor.	Yes, as expected, due to early loss of basicity in ladle.
22/3279	Small.	Still less than normal.	2·3 medium/high.	Good in spite of low basicity.	
23/4596	Medium.	Normal.	$1{\cdot}3_5$ low/medium.	Good.	Low, as expected.
23/4632	Medium.	Normal.	2·09 medium.	Good.	
23/4622	Medium/large.	More than usual.	1·32 low/medium.	Not so good.	0·035–0·051%, in spite of high basicity.
25/2616	Medium/large.	More than usual.	2·33 high.	Not so good.	0·035–0·060%, as above.
22/3249	Large/medium.	More than usual.	1·77 medium.	Poor.	0·040–0·055%, in spite of very high basicity.

an optimum slag condition, especially the amount of the free-lime phase (depending on the steel quality), and that the use of excess lime can be equally as serious as the use of insufficient lime, if not more so.

Mr. Pearson's remarks on his citric-acid-solubility tests are worthy of note in that he finds such a marked difference in behaviour among slags taken during the refining period. I do not think that there is any doubt but that a correlation will be found between the analysis and constitution of the slowly cooled slag and the solubility value. Mr. Pearson discusses the possibility outlined by Dr. Rait and Mr. Goldschmidt of using X-ray methods for slag control. May I say that I have taken slag samples to this

end. The problem is not one of merely determining the presence and amount of the lime and iron-oxide phases in a cooled slag, for in addition one must determine the nature of the lime silico-phosphate, *i.e.*, whether the SiO_2/P_2O_5 ratio is high or low. A further complication arises from varying magnesia content. This oxide is dissolved in the ferrous-oxide phase in the slowly cooled slag, thus making an apparent increase in the FeO solid-solution phase. This is an unfortunate condition, because an increase in the magnesia content can cause poor dephosphorization. At the moment I do not visualize the use in the works laboratory of X-ray methods for the control of open-hearth furnace slags. Replying to Mr. Pearson's next question, a number of slags, with 15–20% of SiO_2, 15–20% of P_2O_5, and 27–30% of CaO, have been examined in the water-quenched condition and found to contain this lime silico-phosphate. Mr. Pearson is quite right in stating that convection currents in the metal contribute to rephosphorization during the pouring of the ingots.

Mr. Honeyman states that increasing the CaO/SiO_2 ratio up to a certain point, depending presumably on the method of working the charge, assists dephosphorization, but that beyond this point there is no advantage in having a high basicity value. I naturally agree with this statement; in fact, there is an optimum slag condition beyond which (higher basicity) the slag becomes less efficient in its reactions with the metal.

Dr. White's contribution is very welcome. The co-existence of lime and ferrous-oxide solid solutions was determined in our laboratory some time ago. The solution of ferrous oxide in lime offers an explanation why a large amount of uncombined lime in a furnace slag is bad, in that the ferrous oxide normally concentrated in the liquid, active portion of the slag is absorbed by the solid lime phase with loss of slag oxidizing power.

Mr. Crawford is anxious about the X-ray findings. I can assure him that I have not found any evidence of tricalcium or tetracalcium phosphates in open-hearth slags.

Mr. Goldschmidt with Dr. Rait has concentrated mainly on definite oxide systems and has covered a large field of research. His comments and questions are therefore very pertinent to the problem of slag constitution. The phase which I described as silico-fluorapatite had an X-ray pattern characteristic of the apatites with lattice dimensions appreciably greater than those of fluorapatite. After consideration of the effect of solid solution of other oxides I decided that these increased dimensions could only be attributed to a replacement of phosphorus by silicon. The increase of the spacing of the lime phase during cooling of the slag sample is due to a loss of MnO and FeO. In many cases this results in two cubic phases, (*a*) FeO–MgO-rich phase, and (*b*) MnO–FeO-rich phase. Regarding the next question, I can only repeat that I believe that the reaction during the cooling of these slags is of the form:

$$\text{Lime silico-phosphate} + \text{liquid} \longrightarrow \text{lime silicate } (2CaO.SiO_2) + \text{silico-fluorapatite}.$$

The problem of the two cubic phases has been referred to above; the values quoted by Mr. Goldschmidt, 4·26 Å. and 4·34 Å., will therefore correspond to the FeO–MgO phase crystallized from the liquid phase and the FeO–MnO phase caused by a lowering of the solubility of FeO in CaO during cooling. In some cases it appears that during cooling these two phases have reacted to form one phase of nearly uniform composition. The examination of the products of the reactions $2FeO.SiO_2 + 2CaO$ and $2MnO.SiO_2 + 2CaO$ was carried out in view of Dr. Zea's use of Schenck's hypothesis regarding the nature of slags. I can assure Mr. Goldschmidt that his reasoning was also mine—nevertheless for the reason stated the work was done. The iron and manganese silicates were prepared to the compositions given by the formulæ in Appendix II., Section (2). In studying the reaction of lime

and iron silicate quantities were chosen to give a lime–iron silicate of $(CaO)_{1\cdot 6}.(FeO)_{0\cdot 4}.SiO_2$ (Section II., Results, third paragraph). In reply to Mr. Goldschmidt's next query, the use of the word "olivine" (third paragraph) was meant to imply a phase of the $2RO.SiO_2$ type. Mr. Goldschmidt wishes to know if the phase described as $(CaO)_{1\cdot 3}.(FeO)_{0\cdot 7}.SiO_2$ was single-phased; yes, it was, and the pattern agreed with that of another sample made to this composition. This phase is not a merwinite, a phase which has been found by Mr. Goldschmidt. I have no personal experience of a manganese merwinite.

Mr. Goldschmidt apparently prefers the use of the word "ion" to "molecule" in describing the reaction of lime with iron and manganese silicates. I do not understand his reason for such a statement. I was not aware that atomic, molecular or ionic replacements were confined to systems where the end members had the same atomic structure. The phase (or phases) found between $2CaO.SiO_2$ and the lime silico-phosphate nagelschmidtite may have a connection with Trömel's *K* phase, but this has not been definitely established. Tetracalcium phosphate was not found in slags where the SiO_2/P_2O_5 ratio was greater than unity. The occurrence of the phase requires a low SiO_2/P_2O_5 ratio. The phase $2CaO.Fe_2O_3$ was clearly one to be considered in the first place, but it was equally clear that in the slags quoted in the paper this phase was not present. Our work has shown that $2CaO.Fe_2O_3$ is present only in slags with abnormally high slag iron and basicity value. The constitution of a truly basic slag is considered to be a dispersion of particles of lime (not pure) and lime silico-phosphate in a liquid. There is no FeO-rich phase at furnace temperature; this phase crystallizes during cooling. Mr. Goldschmidt's last query is in regard to the mode of occurrence of Fe_2O_3 in mixtures of lime and nagelschmidtite (experiment 7(*b*)); there is evidence from this experiment and from quickly cooled furnace slags for the belief that some ferric oxide dissolves in the lime silico-phosphate phase.

Mr. Pearson questions the implication in Table X. that the need for a higher oxidizing power of the slags in steels 2 over steels 1 arises from the difference in final manganese contents. Whatever the significance of the data, is this an example of a fortuitous coincidence that the ratio of the final manganese contents or the ratio of the manganese additions should agree with the ratio of the slag oxidizing powers required to perform the same task —reduce rephosphorization to a negligible amount for the two types of steel?

Mr. Geary sums up the position very concisely and I am grateful for his comments.

Replying to Mr. Preece, I may say that the thermal analysis of basic open-hearth slags has not, as yet, been given a high priority.

In conclusion, I thank all who have contributed to the discussion.

SECTION II.

A SURVEY OF LITERATURE ON THE MANUFACTURE AND PROPERTIES OF IRON AND STEEL, AND KINDRED SUBJECTS.

CONTENTS.

Abstractors and translators: Mr. R. A. Ronnebeck and Dr. D. R. Steiner.

REFRACTORY MATERIALS

Chrome Ore and Chrome Magnesite. J. H. Chesters. (Iron Age, 1943, vol. 152, Nov. 11, pp. 68–71; Nov. 18, pp. 52–55, 166). The distribution and properties of chromium-bearing ores and the manufacture, properties and applications of chrome and chrome-magnesite bricks are discussed.

The Effect of Hydrocarbon Gases on Refractory Materials. Part VI. A Further Study of the Action of Methane on Certain Firebricks. E. Rowden. (Transactions of the British Ceramic Society, 1944, vol. 43, July, pp. 105–116). Certain firebrick specimens when subjected to the action of methane at 800° C. were disintegrated after 15–40 hr. The iron spots in the brick were the foci of the attack, carbon being deposited at the iron spots, this carbon deposition and accumulation apparently causing the disrupture of the bricks. Analyses of the gases leaving the furnace tube showed that: (1) Pronounced decomposition of the methane into carbon and hydrogen to the extent of approximately 25% occurred only when the specimens were disintegrating. (2) Small percentages of the order of 0·5–2·0% of carbon monoxide were present; this may have been produced by reduction of the oxides present in the brick, such as the iron oxide. (3) Traces of unsaturated hydrocarbons were present, but usually the amounts were practically negligible. The rate of disintegration of the specimens bore no relation to the total iron content of the brick, although it must be related in some manner to the iron present. Silica specimens were found to resist the disruption by methane at 700° C., 800° C., and 900° C.

The Effect of Hydrocarbon Gases on Refractory Materials. Part VII. The Change in the Composition of Coal Gas when Passed under Different Conditions over Refractory Materials at 800° C. E. Rowden. (Transactions of the British Ceramic Society, 1944, vol. 43, July pp. 116–125). *See* Journ. I. and S.I., 1940, No. I., p. 118 A).

The Effect of Hydrocarbon Gases on Refractory Materials. Part VIII. Some Aspects of the Action of Methane on Refractory Materials. E. Rowden. (Transactions of the British Ceramic Society, 1944, vol. 43, July, pp. 125–130). The results of further experiments on the action of methane on refractory materials are reported. These show that: (1) The minimum temperature of attack of purified dry methane on a number of firebrick specimens was between 600° and 700° C.; (2) the black deposit formed among the disintegrated firebrick specimens contained 70% of carbon, the remainder consisting of brick substance and some form of iron compound; (3) methane saturated with water vapour at 55° C. did not attack the specimens maintained at a temperature of 800° C. in a period of 210 hr., although similar specimens exposed to the dry gas were disintegrated in from 15 to 44 hr.; (4) specimens of a firebrick, fired in either a reducing or oxidising atmosphere to 1400° C. before subjecting them to the action of methane, resisted attack by the gas at 800° C. better than the ordinary specimens obtained direct from the brick; those specimens fired in a reducing atmosphere were the more resistant to the gas; (5) specimens of a firebrick fired in either a reducing or oxidising atmosphere to 1300° C. and held at this temperature for 1½ hr. did not resist the action of the methane at 800° C. any better than the untreated specimens; and (6) specimens of a firebrick fired in a reducing atmosphere to 1300° C. and held at this temperature for 4 hr. resisted the attack of methane at 800° C. (for 200 hr.), whereas specimens fired similarly in an oxidising atmosphere were disintegrated in the same time as the untreated specimens.

FUEL

The Industrial Gas Burner. B. Schmidt. (Giesserei, 1944, vol. 31, Feb., pp. 25–29). The fundamentals of the design of gas burners for industrial purposes are outlined and the difference in the length and type of flame produced by premixing the gas and air and by mixing them at the burner orifice are discussed. Some gas burners for heating foundry ladles are described.

Factors Influencing the Moisture Content of Furnace Atmospheres. F. Gilbert and E. Scheuer. (Metallurgia, 1944, vol. 30, Aug., pp. 187–190). Data on the effects of moisture and hydrogen in fuel and of moisture in the air on the amount of moisture in the products of combustion are presented and discussed The most important factor is the hydrogen content of the fuel; if more than 5% is present no changes in the moisture in either the air or the fuel will have any appreciable effect on the final moisture content of the combustion gases.

German Steelworks Power Plant. K. Schröder. (Engineering and Boiler House Review, 1944, vol. 58, Sept., pp. 222–229, 248). This is an abridged translation of a paper on steam-driven power stations for steelworks which appeared in Stahl und Eisen, 1944, vol. 64, Jan. 6, pp. 2–10; Jan. 13, pp. 24–29. (*See* Journ. I. and S.I., 1944, No. II., p. 118 A).

The Fuel Systems at Great Lakes Steel Corporation F. C. Frye. (Iron and Steel Engineer, 1944, vol. 21, Aug., pp. 78–81). At the Ecorse plant of the Great Lakes Steel Corporation, built in 1929, the open-hearth and reheating furnaces were designed for burning fuel oil. In 1938 blast-furnaces and coke-ovens were erected and large quantities of mixed blast-furnace gas and coke-oven gas became available. The modifications in the fuel supply and control systems which were made to the furnaces to permit of the two fuels being used are described.

Application of Metallic Recuperators. J. H. Loux. (Steel, 1944, vol. 115, Aug. 21, pp. 118–120). Some applications of metal recuperators to reheating furnaces for forges and rolling mills are discussed. In cases where the flue-gases are too hot for the recuperator to withstand, equipment for diluting the gases with air is provided so that the temperature is brought down to 1700° F. If this temperature is not exceeded the recuperator maintenance is negligible.

Raw Material Mining and Beneficiation at Kaiser Company, Inc., Iron and Steel Division, Fontana. T. M. Hart. (Blast Furnace and Steel Plant, 1944, vol. 32, July, pp. 798–802). An account is given of the mining, washing, stocking, crushing and grading of coal used in the Fontana coke-ovens. The mining and subsequent treatment of the iron ore smelted by the plant are also described.

Coke Plants for the Sinews of War. (Kopper's News: Blast Furnace and Steel Plant, 1944, vol. 32, July, pp. 811–814). A general description is given of the construction of a modern Koppers coke-oven plant. Photographs are included of the new coke-ovens at Fontana, California, during and after completion.

Coals for Carbonisation. J. Roberts. (Coke and Smokeless-Fuel Age, 1944, vol. 6, Sept., pp. 173–175, 177). A general survey and classification is given of those coals and related carbonaceous deposits which may be utilised for gas and coke production. An account is also given of the British Standards Institution's crucible swelling test for coals and the interpretation of swelling numbers is explained.

Sulphur Removal and Recovery from Coke-Oven Gas. N. G. Farquhar. (Chemical and Metallurgical Engineering, 1944, vol. 51, July, pp. 94–96).

The corrosive action of sulphur compounds in coke-oven gas is a problem of increasing severity. The Thylox system of sulphur recovery, recently introduced at the Ford Motor Company's River Rouge plant, is described.

The Products of Carbonisation. J. W. Cobb. (Fuel in Science and Practice, 1944, vol. 23, Sept.–Oct., pp. 121–127). A variety of substances were subjected to prolonged fractional carbonisation at temperatures advancing by 100° stages to 1200° C., securing as far as possible completeness of reaction at each stage. The quantity of the gaseous products was measured and the solid products were examined by X-rays. The substances carbonised were cellulose, bakelite, glycine, petroleum-coke, anthracite, and New Hucknall and Sharlston Wallsend coals. An orderly sequence in the emission of primary gaseous products resulted, which was practically the same for all the substances after passing the 500° C. stage. X-ray examination clearly disclosed carbon atoms in hexagonal networks at 500° C., a structure in which the increasing thermal shocks of higher temperatures only led to further growth and stabilisation. A causal connection seems to be established between the oxygen surviving to 400–500° C. in carbonisation, the micro-capillarity and the reactivity.

Producer Gas for Small Heat-Treatment Furnaces and Other Processes. (Metallurgia, 1944, vol. 30, Sept., pp. 269–271). **Small Producer-Gas Unit.** (Sheet Metal Industries, 1944, vol. 20 Oct., pp. 1727–1728). **Small Producer-Gas Heat-Treatment Furnaces.** (Metal Treatment, 1944, vol. 11, Autumn Issue, pp. 157–160). **Small Producer-Gas Furnaces.** (Iron and Coal Trades Review, 1944, vol. 149, Sept. 22, pp. 413–415). A description is given of the new Wellman-Cura 7½-therm gas producer and scrubber. The producer, with its ancillary heat exchanger, stands on its own feet on any level surface. On full load it consumes not more than 1 cwt. of fuel per hr. It can be operated for 8 hr. at 7½ therms per hr. without shutting down for removing ash.

PRODUCTION OF IRON

Modern Blast-Furnace Design and Operation. J. Dale. (Journal of the West of Scotland Iron and Steel Institute, 1943–44, vol. 51, pp. 45–106). A comprehensive review of blast-furnace design and operation, particularly with respect to the American type of furnace is presented. Particular attention is paid to modern types of blast-furnace charging gear and to the problem of the distribution of the charge at the stock line.

Standardisation of Blast-Furnaces. G. Bulle. (Stahl und Eisen, 1944, vol. 64, May 4, pp. 285–290). The German pig-iron producers and blast-furnace builders have agreed upon a standard form of blast-furnace, details of which are presented and discussed. The design incorporates the latest developments accruing from experience and represents a considerable saving in constructional materials. The furnace has a volume of 825 cu. m. and a hearth diameter of 6 m. For instance, the shell plating weighs 200 tons less than that of earlier 850-cu. m. furnaces and only about 1000 tons of refractories instead of 1360 tons are required. (An English translation of this paper is available as No. 199 in the Translation Series of The Iron and Steel Institute).

The Smelting of Krivoi Rog Lump and Fine Ores in Burdens for Basic Iron. H. Reinfeld. (Stahl und Eisen, 1944, vol. 64, Apr., 6, pp. 217–222). During the German occupation of the Ukraine quantities of Krivoi Rog ore became available to the blast-furnaces in eastern Germany. This ore varied greatly in its physical properties and there was a high proportion of fines in it. An account is given of extensive trials in the smelting of this

ore which were carried out with a small blast-furnace in 1943. Burdens containing different proportions of Krivoi Rog ore were tried and it was found quite satisfactory to use 350–400 kg. of this ore per 1000 kg. of iron produced, even when it contained a high percentage of fines. A considerable improvement was obtained by crushing the large lumps and sintering the fines.

Studies on Solubility of Carbon in Silico-Manganese. N. N. Kurnakov. (Comptes Rendus (Doklady) de l'Académie des Sciences de l'URSS, 1943, vol. 39, No. 8, pp. 314–316). The results of tests to determine the solubility of carbon in silico-manganese in relation to the silicon content are reported. The carbon was introduced into the melts in the form of 77·92% ferro-manganese containing 4% of carbon. In the range of 10–23·4% of silicon, 0·8% of carbon was in solution and any excess of carbon was precipitated as double carbides of iron and manganese and as graphite.

Sponge Iron—Its Possibilities and Limitations. C. F. Ramseyer. (Iron and Steel Engineer, 1944, vol. 21, July, pp. 35–44). The processes of producing sponge iron are reviewed. The Wiberg process is the only one in successful operation at present, but it has three disadvantages. It cannot use the fine concentrates just as they come from the magnetic separator, which means the additional cost of sintering; secondly, the operation has to be carried out at such a high temperature that sticking difficulties are encountered; and thirdly, the regeneration of carbon dioxide to carbon monoxide is not very satisfactory. The Hoganäs and Anderson processes can use finely divided ore, but the heat transmission to the ore is inefficient. The use of hydrogen under pressure is advocated as a reducing agent and its cost and efficiency are compared with those of carbon. The reduction of magnetite to iron with hydrogen should not be carried out at above 1100° F., otherwise ferrous oxide will be formed which is difficult to reduce. These considerations have led to the development of the Brassert process, in which coke-oven gas is used as the source of hydrogen. A plant for this process is being built at Warren, Ohio. In this the coke-oven gas is desulphurised, preheated to 1200–1300° F. and passed into the bottom of a vertical Herreshof furnace into which the ore is fed at the top. In this furnace the finely divided ore passes down over a series of disc plates to the slotted grate at the bottom, through which the hydrogen passes upwards. By bubbling the gas through the fine ore the latter is made to flow along channels and from one plate to another as if it were liquid. To prevent reoxidation of the hot reduced fines they are passed directly to a briquetting press, which is designed to make corrugated slabs of compressed sponge iron measuring $7\frac{1}{2} \times 5\frac{1}{2} \times \frac{3}{4}$ in. with an expected density of between 5 and 6. Design data for this process are compared with blast-furnace operating data.

The Manufacture of Electric-Furnace Pig Iron. C. Hart. (American Iron and Steel Institute: Steel, 1944, vol. 115, Aug. 7, pp. 111–114, 140). The conditions for the production of electric-furnace pig-iron in Scandinavia are compared with those prevailing in North America. Whilst production is economic in some parts of Europe where electricity is cheap, there is no prospect of manufacture on a large scale in the United States.

Processes for Making Metal-Powder Products. E. V. Crane and A. G. Bureau. (Electrochemical Society: Steel, 1944, vol. 115, June 24, pp. 90–94, 128–133). *See* Journ I. and S.I., 1944, No. II., p. 72A.

Sword Moulds *v.* Currency Bars. E. Wyndham Hulme. (British Steelmaker, 1944, vol. 10, Aug., pp. 342–347). The origin of iron sword moulds (from 21 to 34 in. long, weighing between 436 and 745 g.) found mainly in Gloucestershire, is discussed and the theories of Thomas Wright and of the Department of British and Mediaeval Antiquities (the " Museum theory ") are put forward. Wright maintained that these moulds were

for making swords and that they belonged to the late Celtic period (250–75 B.C.). The Museum theory is that the moulds were a form of Celtic money for currency purposes only.

FOUNDRY PRACTICE

The Desulphurisation of Cupola Iron with Special Reference to the Process in the Bessemer Shop. R. Stolle. (Giesserei, 1944, vol. 31, Feb., pp. 21–25). After a review of the literature on various methods of reducing the amount of sulphur in cupola iron, some of the author's investigations on factors affecting desulphurisation with soda ash are described. Tests with soda in an inverted crucible resting on the bottom of a ladle were carried out in such a way that when the iron was poured in the reaction took place under the pressure set up by the gases generated; the iron tapped from the bottom of the ladle contained only 0·002–0·005% of sulphur, whereas a spoon sample from the top of the ladle contained 0·110%. Other tests showed that additions of 25, 30 and 35 kg. of soda to 3500 kg. of iron in the ladle reduced the sulphur content by 45·8%, 51% and 57·2% respectively. Every shop must experiment to determine the optimum quantity of soda to use per ton of iron, but it will generally work out at about 10 kg.

Introduction of the Production of Anti-Friction Cast Iron. N. I. Feygin. (Vestnik Metallopromyshlennosti, 1940, No. 11–12, pp. 15–22). (In Russian). The first part of the paper records experience made in applications of anti-friction cast iron, and the second deals with the production and testing of the material. In practice the life of anti-friction cast-iron bearings has proved to be equal to that of bronze bearings. The anti-friction cast iron developed contained total carbon 3·2–3·6%, silicon 2·2–2·4%, manganese 0·6–0·9%, phosphorus 0·15–0·25%, sulphur 0·12% max., chromium 0·2–0·35%, nickel 0·3–0·4%, and copper 0·2–0·3%; 0·1–0·15% of aluminium may be added. In the cupola charge the use of naturally alloyed Ural charcoal iron is recommended. Cupola melting is preferred to electric-furnace melting, as it assists in securing the required structure, which is that of a pearlite matrix, a phosphide eutectic in the form of small inclusions, ferrite (not exceeding 15%) and lamellar graphite.

The Development and Production of Inoculated Cast Iron. H. P. Hughes and W. Spenceley. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 73, Aug. 31, pp. 349–354; vol. 74, Sept. 7, pp. 3–7; Sept. 14, pp. 31–35, 30). Certain castings (*e.g.*, band saw tables 30 in. square × ⅝ in. thick) were made of iron containing a high proportion of hematite iron and were weathered for a long time before machining. As hematite has become very scarce, experiments were made with inoculated irons in order to improve the machining properties and reduce distortion. The ferro-silicon feeder installed over the cupola spout is described. The results of mechanical tests and microstructure investigations of irons with and without inoculation with ferro-silicon are presented. It is suggested that the graphite in random formation produced by inoculation is the result of graphitisation taking place in the melt previous to eutectic temperature, in other words, proeutectic graphite is produced. The improvements in the properties of ordinary cupola metal obtained by inoculation were: Impact value 63%, transverse strength 30%, deflection 28%, tensile strength 23% and Brinell hardness 2%.

Manufacture of Homogeneous Armor, Both Cast and Rolled. D. O. Davis. (Metal Progress, 1944, vol. 45, June, pp. 1081–1087). A brief account is given of the changes in peace-time operations necessary to pro-

duce gun-barrels, armour-plate and other heavy ordnance materials at the Dominion Foundries and Steel, Limited, Hamilton, Ontario.

Casting Alloy Steel for the Oil-Field Industry. (Steel, 1944, vol. 115, Aug. 21, pp. 102–108, 154). An illustrated account is given of the casting operations adopted by the Oklahoma Steel Casting Company for the manufacture of alloy-steel valves and other parts for oil refineries.

Some Useful Wartime Developments in Whiteheart Malleable Iron. G. R. Webster. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 74, Oct. 5, pp. 87–91; Oct. 12, pp. 109–114; Oct. 19, pp. 139–142). A detailed description is given of the manufacture of tank-track links of whiteheart malleable cast iron. The molten metal from the cupola contains total carbon 2·9–3·1% and silicon 0·55–0·7% and is made from a charge containing about 68% of returned scrap, 10% of hematite or refined iron, 21% of steel scrap and 1% of 13% ferrosilicon. All patterns and core-boxes are completely machined to within ±0·004 in. The malleablising is carried out in a town-gas-fired tunnel furnace at 980° C. taking 80–100 hr. After the required machining, the castings are hardened by heating to 780° C. and quenching in a mineral oil. This is followed by double tempering. Data on the tensile properties of the hardened and tempered castings are presented. Experiments to ascertain the causes of "peeling" are reported. The peeling could be produced artificially with sulphurous gases. The sulphur content of the metal appeared to have little effect on the peeling, which was probably caused by the sulphurous gases generated from pyritic pieces in the ore in which the castings were packed for malleablising.

Some Principles of Production Control and Their Application to Steel Foundry Work. A. B. Lloyd. (Journal of the Institution of Production Engineers, 1944, vol. 23, Aug., pp. 279–296). The first part of the paper is devoted to an explanation of the basic conditions which are necessary in the organisation of a works in order that production control can be successfully applied. A description is then given of the principles involved and how they have been applied to facilitate the manufacture of steel castings.

Tests Temperature Attained within Specimens of Molding and Core Sands. D. C. Williams. (American Foundrymen, 1944, vol. 6, July, pp. 5–6). A testing procedure is described for determining the rate of heat transmission through rammed moulding and core sands. The tests were made with a thermocouple attached to a steel pin $1\frac{5}{8}$ in. long × $\frac{1}{4}$ in. in dia. at the centre of a cylinder of sand 2 in. long × $1\frac{1}{8}$ in. in dia.

The Conversion of Cupola Slag into Core Sand. A. Fischer. (Giesserei, 1944, vol. 31, Jan., pp. 5–7). A method of granulating cupola slag and grinding the granules into sand suitable for making cores is described. Grain-size analyses of this sand were made and the results compared with those for natural sands. Two years' experience with cores made of the new kind of sand has proved them to be very satisfactory even for very large cores. The cost of making the sand is very low, and the process solves the problem of what to do with an otherwise useless material.

A Foundryman's Notebook. I. Fitness for Purpose of Patterns and Cores. (Iron and Steel, 1944, vol. 17, Aug., pp. 564–565). Examples are given which show how the production of good castings was facilitated by changes in pattern design and the use of cores.

Graphite Rods Applied to Open Risers. F. J. Vosburgh and H. L. Larson. (Foundry, 1944, vol. 72, July, pp. 74–75, 180–186). The use of graphite rods in the risers of steel castings to reduce the shrinkage cavity has already been described (*see* Journ. I. and S.I., 1944, No. II., p. 40 A). Additional experience with this practice applied to open risers is now reported. The general conclusions are: (1) A smaller riser can be used when a graphite rod is placed across it; (2) the rod should be placed 2–3 in.

down from the top of the cope section of the mould; (3) the level of the rod above the junction of the riser and casting should be at least 2–3 in., to prevent carbon from being taken up by the casting; (4) a pointed rod pushed into the sides of the riser cavity is better than a flat-ended one resting at the bottom of two grooves; (5) with a pointed rod the riser becomes sealed over with a layer of solidified metal, below which the gas pressure becomes greater than atmospheric pressure; this ensures good feeding; and (6) the shrinkage cavity with a pointed rod is wider and shallower than with a flat-ended one.

The Influence of Centrifugal Casting upon the Structure and Properties of Steel. L. Northcott and D. McLean. (Iron and Steel Institute, 1945, this Journal, Section I.). Thick cylinders of a nickel-chromium-molybdenum steel were cast by the centrifugal process using chill moulds rotating about a horizontal axis without a central core. The casting conditions were varied with respect to mould speed, casting temperature and rate of pouring, and each casting was examined to determine the influence of these factors upon its structure. At low mould speeds the delay in pick-up, *i.e.*, the interval required for the metal to be rotated with the mould, resulted in the liquid tumbling about inside the mould, thus giving rise to the coarse circumferential zoning referred to as "type 1 structures," the explanation of which was confirmed by the preparation, at low mould speeds, of a series of castings in which the mould was rapidly stopped before the completion of solidification. Above these mould speeds there was an optimum range which gave uniform structures, referred to as "type 2 structures," free from segregation. At the highest mould speeds the vibration developed resulted in circumferential bands of segregate referred to as "type 3 structures," and the association of such segregation with vibration was confirmed by tests on vibrated unrotated ingots. High casting temperatures favour the formation of the type 1 structure, reduce the type 3 banding at high mould speeds and promote the growth of large primary crystals. Rapid pouring was found to favour segregation and, at high speeds, to lead to radial cracking; circumferential lapping was only observed under conditions of very slow pouring and at low casting temperatures. The mode of solidification of castings showing the different structures is discussed.

The Development of Methods of Chill and Pressure Die Casting. S. E. Rosenfeld. (Vestnik Metallopromyshlennosti, 1940, No. 7, pp. 49–54). (In Russian). Factors affecting the choice of chill casting or pressure die casting are discussed. Pressure die casting of ferrous metals in Russia is in the experimental stage. In the machine described, the metal is melted by heat generated by the passage of an electric current through the enclosed graphite crucible containing the metal. Injection into the die is done by compressed air through an opening in the bottom of the crucible opened and closed by a needle valve.

Centrifugal Lining of Cylinder Liners with Boron-Bearing Cast Iron. I. S. Brokhin and A. A. Davydov. (Vestnik Metallopromyshlennosti, 1940, No. 6, pp. 66–67). (In Russian). Steel cylinder liners of mud pumps used in drilling oil wells have been lined with a cast iron containing 0·7–1·2% of boron to improve the wear resistance. Ferro-boron (containing 10–15% of boron) was produced by an aluminothermic reaction, and the boron-bearing iron by melting in an electric furnace. The requisite quantity of the alloy was placed in the liner, which was closed by end plates welded or screwed on. The alloy was then melted in the liner by heating to 1200° C., after which the coating was formed by rotating the liner at 800 r.p.m. for 4–6 min. The inside of the liner was then ground to size. The alloys had a hardness of Rockwell C. 67–70. A liner coated as described lasted longer than nine plain steel liners.

Competitive Characteristics of Centrifugal Castings. J. Putchinski. (Iron Age, 1944, vol. 153, May 4, pp. 48–51). An outline is given of the types of castings that can be produced centrifugally. Factors determining the cost of parts manufactured by this means are discussed and some of the limitations of centrifugal casting are mentioned.

It Can be Cast Centrifugally. J. Putchinski. (Foundry, 1944, vol. 72, Aug., pp. 84, 190–194). The types of casting which lend themselves to production by the centrifugal method are described, with notes on factors affecting the economics of the process.

Speed of Rotation in the Centrifugal Casting Process. J. E. Hurst. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 74, Oct. 5, pp. 95–99; Oct. 12, pp. 117–120). The conditions of rotational speed adopted in the various types of centrifugal casting processes proposed or adopted commercially are reviewed. In each type of process there are differences in the speed conditions adopted by individual operators, but at least two notions of principle governing the speed conditions exist. The first is that of the choice of speed to obtain an approximately constant condition of tangential or centrifugal force during rotation. Different formulæ for the predetermined conditions of tangential force have been developed from practical experience; some of these are presented and discussed. An alternative notion is that of the choice of a speed of rotation to conform to a uniform predetermined peripheral velocity. The rotational speed depends very frequently on the method and conditions of pouring, the nature of the mould and the characteristics of the liquid metal.

Automobile Castings. (Automobile Engineer, 1944, vol. 34, Aug., pp. 315–323). The layout and organisation of the North Works of the Midland Motor Cylinder Co., Ltd., are described. Three special cast irons, Chromidium, Monikrom and Cromol, are produced for making cylinder blocks and heads, crankcases and brake-drums, camshafts and crankshafts. A high degree of mechanisation has been achieved at this foundry.

Automobile Castings. (Automobile Engineer, 1944, vol. 34, Sept., pp. 352–359). A new plant of the Midland Motor Cylinder Co., Ltd., erected for the large-scale production of alloy cast-iron cylinder blocks, cylinder heads and brake-drums is described and illustrated. The casting of automobile components requires very careful quality control; the means of ensuring high metallurgical and dimensional standards at this plant are described.

New Gray Iron Foundry Has Interesting Features. E. Bremer. (Foundry, 1944, vol. 72, July, pp. 70–73, 201–206). An illustrated description is given of the grey iron foundry recently completed for Deere and Co., Moline, Illinois, in which a high degree of mechanisation has been achieved, especially with regard to sand handling.

Specifications, Quality Control and Inspection. J. F. Kayser. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 74, Sept. 28, pp. 67–70). It is deplored that so many parts are made from drop forgings instead of castings. Indications are given of what a buyer should and should not include in specifications, and the use of frequency diagrams for control purposes and of X-ray apparatus for inspection is strongly recommended.

Handling Materials in the Crane Steel Foundry. E. F. Ross. (Foundry, 1944, vol. 72, Aug., pp. 75–77, 182–188). **Improved Production Efficiency.** E. F. Ross. (Steel, 1944, vol. 115, July 31, pp. 68–71, 110–113). A description is given of the foundry operations at the works of Crane Co., Chicago, where cast-steel valves are made.

Machine Shop Practice and Foundry Technique. C. W. Marshall. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 74, Sept. 7, pp. 11–12, 14). An appeal is made for increased co-operation

between the machine shop and the foundry, so that castings will be designed with the requisite cores, bosses or pads which will facilitate clamping and setting up in modern machine tools.

The Problem of Ventilation in Foundry Buildings. (Giesserei, 1944, vol. 31, Jan., pp. 2–4). The importance of the temperature, moisture content and circulation of the air in foundries in relation to the physical well-being of the workers is pointed out and recommendations for maintaining a high standard are made.

PRODUCTION OF STEEL

The Reorganisation of the English Iron and Steel Industry. J. W. Reichert. (Stahl und Eisen, 1944, vol. 64, June 29, pp. 413–419). Statistics on the number of plants and the production of pig-iron, steel and rolled products in Great Britain are presented and discussed. The organisation of the industry is criticised and compared with that of the German iron and steel industry. (An English translation of this paper is available as No. 195 in the Translation Series of The Iron and Steel Institute).

The Supply of Water for Iron and Steel Works. K. Guthmann. (Stahl und Eisen, 1944, vol. 64, May 11, pp. 301–310). Not only do iron and steel works consume large quantities of water, but the power consumed in supplying it at the required points at the necessary pressure amounts to 10–20% of the total electricity consumption of the works. In the general desire for economy and reduced costs this field has been neglected. The present report on water supply and circulation is made on the basis of data from the literature, answers to a questionnaire and the author's own experience. Water-circulation diagrams and consumption data relating to coke-ovens, blast-furnaces, steel furnaces, rolling-mills and reheating furnaces are presented and discussed.

The Water Supply for an Integrated Iron and Steel Works Situated Beside a Natural Water Course. H. F. Strähuber. (Stahl und Eisen, 1944, vol. 64, June 8, pp. 373–377). In an integrated iron and steel works the electricity consumption for supplying water to all departments may be as much as 20% of the total consumption of the works, and considerable economies can often be effected in this sphere. The water distribution, consumption and pumping arrangements at a German works drawing its supply from a river are described in detail with circulation diagrams. Methods of metering the consumption of different departments and of constructing flow diagrams are discussed and used as a basis for the study of possible economies.

The Water Supply of an Integrated Iron and Steel Works Not Situated Beside a Water Course. W. Güldner. (Stahl und Eisen, 1944, vol. 64, Aug. 3, pp. 497–503). A detailed account is given of the water-supply system of an iron and steel works which has to obtain its water from a town's water undertaking. The consumption of water per ton of steel ingots is 170 tons, of which 8 tons is make-up water from the town's mains, the remainder being in circulation within the works system. The method of drawing up a water-supply balance sheet is explained with an example. A rotating-drum type of water-filter for removing mill-scale, blast-furnace dust and other solids from the water is described. The consumption of the separate departments is discussed with a view to showing where the greatest economies can be effected.

The Metallurgy of the Side-Blown Converter Process. H. Wentrup and O. Reif. (Stahl und Eisen, 1944, vol. 64, June 1, pp. 349–358). The reactions in a 1-ton side-blown converter with an acid lining were studied by analysing

samples of steel and slag taken at 3-min. intervals during the blowing period. Contrary to the large bottom-blown converter practice, the reactions in the side-blown converter did not attain equilibrium until the end of the blow; this was due to the blast keeping particles of oxidised iron in suspension in the slag, thus causing the slag to be too high in ferrous oxide. If the slag were thick, this surplus oxide was only got rid of slowly. The addition of agents which liquefy the slag would therefore help the reactions to attain equilibrium more rapidly. Some characteristics of basic-lined side-blown converter practice are also discussed and compared with large converter practice.

Supplementary Heating of the Checker Chambers in the Operation of Open-Hearth Furnaces Fired with Cold Gas. A. Mund. (Stahl und Eisen, 1944, vol. 64, July 27, pp. 481–485). German steelworks using brown-coal dust for carburising in firing open-hearth furnaces have experienced difficulty with the rapid accumulation of ash in the slag pockets and checker chambers. A method of cleaning the checkers while the furnace is in operation is described. This is accomplished by fitting burners for coke-oven gas on the exit side of the checker chambers. This " melts out " the deposits and also gives additional preheating to the air. The effect of this procedure on the furnace temperature and production is discussed.

Operation of the Fontana Plant of the Kaiser Company, Inc., Iron and Steel Division. T. M. Price. (Engineers' Society of West Pennsylvania: Blast Furnace and Steel Plant, 1944, vol. 32, June, pp. 675–682, 688). Various aspects of the operation of the new large steel plant at Fontana, California, are described.

The Rate of Decarburisation and the Oxygen Content of Steel in the Basic Open-Hearth Furnace. H.-J. Krabiell. (Stahl und Eisen, 1944, vol. 64, June 22, pp. 399–404). An investigation of the relation between the oxygen and carbon contents of steel during decarburisation in the basic open-hearth furnace is reported. The results confirmed Schenck's findings that the oxygen content throughout the heat is governed mainly by the carbon content of the steel and the rate of decarburisation. Samples taken from various depths of the bath indicated that the rate of migration of oxygen from the slag into the steel is not so high as expected by Schenck, and that if no additions are made the development of carbon monoxide takes place mainly at the hearth. The rate at which the reactions proceed is greatly affected by the basicity of the slag, whilst the effects of its fluidity and temperature are less pronounced. (An English translation of this paper has been prepared, *see* Iron and Steel Institute, 1944, Translation Series, No. 197).

Postwar Prospects for Basic Electric Furnace Steel. F. T. Sisco. (Iron Age, 1944, vol. 154, July 27, pp. 34–35, 130–137). The economic and technical factors affecting the production of open-hearth and electric-furnace steels in the United States are considered and a cost study by A. G. McKee and Co. is critically examined. Experience with some of the newer electric-furnace installations indicates that in many localities electric-furnace steel will sooner or later be directly competitive with open-hearth steel where cold metal charging is the practice.

Getting the Most out of Electric Steel Electrode Furnaces. R. L. Baldwin. (Metals and Alloys, 1944, vol. 20, July, pp. 62–67). Various precautions and practices are outlined which make for the most efficient and economical use of electrodes in the production of electric-furnace steel. Handling methods, joint-assembly techniques and the electrical conditions most conducive to the long life of carbon and graphite electrodes are described.

Electrometallurgical Treatment of Ores. C. Hart. (American Iron and Steel Institute: Blast Furnace and Steel Plant, 1944, vol. 32, June,

pp. 673–674, 691–694). Electric furnaces as used for the production of pig iron, ferro-alloys and steel are described, together with a brief account of the different types of electrode at present available in iron-producing countries throughout the world. The manufacture of ferro-chromium and the recarburising of steel with molten iron in an electric furnace are described and a brief account is given of a high-shaft Swedish furnace and a Siemens-Halske furnace.

"Scrapalurgy." H. Brearley. (British Steelmaker, 1944, vol. 10, Sept., pp. 376–384). The problem of the most efficient use of scrap by the steelmaker is discussed with special reference to alloy steel turnings. The price delivered the blast-furnaces for mixed turnings exerts a dominating influence on the price at which alloy steel turnings are bought. It is suggested that it would be worth while for a steelmaker to use a 30–40 ton open-hearth furnace with picked furnacemen to find out how to make good ingots from scrap with the highest recovery of alloying elements, not just to produce the maximum tonnage of ingots.

Utilisation and Processing of Steel Turnings. O. S. Petrenko. (Vestnik Metallopromyshlennosti, 1940, No. 7, pp. 36–45). (In Russian). The cleaning, baling and briquetting of steel turnings are described and data on the performance of imported presses and equipment are given. Of interest is the electric briquetting press of Russian design in which the chips, after being compressed, are heated by passing an electric current through them. Briquettes having a density of 90–95% of the solid metal are obtained. The necessity of cleaning and drying the chips, the relatively low output and the high electricity consumption are drawbacks.

The Identification and Handling of Ferrous and Non-Ferrous Scrap. D. G. P. Paterson. (Sheet Metal Industries, 1944, vol. 20, Oct., pp. 1729–1731). Some methods of identifying scrap metals are reviewed; these include spot chemical tests, magnetic tests and spark tests.

The Relation of Open-Hearth Practice to Segregation in Rimmed Steel. J. W. Halley and G. L. Plimpton, jun. (American Institute of Mining and Metallurgical Engineers: Blast Furnace and Steel Plant, 1944, vol. 32, May, pp. 539–546, 550; June, pp. 683–688). The following factors affecting segregation in rimming steel ingots are discussed: (1) Differential solidification; (2) the removal of carbon, oxygen and manganese; (3) the efficiency of film removal; and (4) the ratio of rim zone to core zone. The desirable type of segregation in rimming steel ingots is promoted when: (*a*) The content of those elements which are not got rid of as reaction products is kept down; (*b*) the metal bath is in vigorous movement due to a high rate of gas evolution; (*c*) the rate of freezing is rapid; (*d*) the pouring temperature is low; and (*e*) the rimming action is checked early by "capping."

Factors Affecting Ingot Product Surface. H. F. Lesso. (Iron and Steel Engineer, 1944, vol. 21, Aug., pp. 63–72). Statistics relating to factors affecting the frequency with which cracks are found in the surface of ingots and billets are presented and discussed. In all the curves the ordinates represent the frequency with which serious surface cracks are observed, and the abscissae represent factors such as the proportions of light, heavy and alloy steel scrap in the charge; the silicon, manganese and phosphorus contents; the final manganese/carbon ratio; teeming practice; and time between teeming and stripping. Separate curves for coated and uncoated ingot moulds are plotted for each factor, and in nearly all cases the coated moulds result in less surface defects. The best results are obtained with the mould at about 250° F. when a pitch dressing is applied.

REHEATING FURNACES

The Rate of Heating Steel. W. Trinks. (Industrial Heating, 1944, vol. 11, July, pp. 1049–1056). Calculations are given which enable the rate of heating of steel ingots to be determined. Different methods of heating most suitable for the different shapes and sizes of ingot are described.

Control of Atmospheres in Open-Fired Furnaces. F. A. Locke. (Iron Age, 1944, vol. 154, July 27, pp. 36–38, 126–130). Factors affecting the decarburising and oxidising tendencies of atmospheres in gas-fired reheating furnaces of the type in which the charge is in direct contact with the products of combustion are discussed.

Experience with a Top and Bottom Heated Pusher Ingot Furnace with Three Zones. F. Fechter. (Stahl und Eisen, 1944, vol. 64, July 13, pp. 449–453). A detailed description is given of a producer-gas-fired reheating furnace in which the ingots lie horizontally across the furnace and are supported on a pair of heat-resisting steel rails passing under the ends. The furnace has three zones, and the burners are arranged so as to project the flames along the furnace above and below the ingots, *i.e.*, the flames pass across the ingot sides. The furnace is 16,380 mm. long × 1700 mm. wide, and it has a throughput of 10 tons per hr. There is no water-cooling. Particulars of the refractories used are given.

Unit Furnaces with Interchangeable Firing for Rolling Mills. W. Völkel. (Stahl und Eisen, 1944, vol. 64, May 18, pp. 321–325). In an effort to standardise reheating furnaces for ingots and to make it easier to alter their size to meet production requirements, a system of unit or cell construction has been developed by which the length of pusher furnaces can be altered by removing or adding units in the centre portion. The application of this principle to a number of furnaces fired by a mixture of blast-furnace and coke-oven gas is described and illustrated with several diagrams. The alterations necessary to change from producer gas to mixed gas firing are also considered.

FORGING, STAMPING AND DRAWING

Canadian Shell Forging Plant Uses Alloy Cast Iron Punch Tips. (Nickel Cast Iron News: Industrial Heating, 1944, vol. 11, July, pp. 1084–1086). A forging press in use at the Dominion Engineering Works, Ltd., in which the punches are tipped with a nickel-alloy cast iron is described. The use of nickel-chromium and nickel-molybdenum cast irons for the same purpose is also mentioned.

The Control of High-Speed Hydraulic Presses by Adjustable-Pressure Rotary Oil-Pumps. K. Schwencke. (Zeitschrift des Vereines Deutscher Ingenieure, 1943, vol. 87, June 26, pp. 391–398). Descriptions are given of the application of Pittler-Thoma rotary oil-pumps for the control of hydraulic presses of various types.

Drawing 105-mm. Steel Cartridge Cases. T. E. Lloyd. (Iron Age, 1943, vol. 152, Nov. 25, pp. 46–51). The sequence of processes in the manufacture of steel cartridge cases from steel blanks 7·385 in. in dia. × 0·515 in. thick is described.

Manufacturing 105-mm. High-Explosive Shell. T. E. Lloyd. (Iron Age, 1943, vol. 152, Dec. 2, pp. 60–63). The sequence of forging, machining and inspection processes in the manufacture of 105-mm. high-explosive shells from steel bars 3 in. in dia × 19 in. long is described.

Contoured-Cavity Shell Forging. (Metals and Alloys, 1944, vol. 19, June, pp. 1402–1404). *See* Journ. I. and S.I., 1944, No. II., p. 124 A.

Evaluating Steel Forgeability. C. L. Clark. (Iron Age, 1944, vol. 153, Mar. 16, pp. 52–55). The apparatus and procedure used in the hot-twist test are described, data from such tests are given and the results of tests on various steels interpreted. It is shown that the results of hot-twist tests may be used as an index of forgeability, since they furnish information on: (1) The temperature which must not be exceeded in the forging operation; (2) the change in ductility of the steel with varying temperature; and (3) the change with temperature of the force required to deform the steel.

Forging Die Design. J. Mueller. (Steel Processing, 1944, vol. 30, July, pp. 425–426, 439). The design of dies for forging rings and hubs is discussed.

Induction Heating and Automatic Spinner Aid in Shaping H.E. Bomb Casings. (Industrial Heating, 1943, vol. 10, July, pp. 972–974). A considerable reduction in the time required to shape bomb casings has been effected at an ordnance plant in the United States. A reorganisation of the production line is described in which induction heating and automatic spinning of the case are employed, the number of operations necessary for shaping being cut down from ten to two.

The Technique of Deep Drawing as Applied to Automobile Pressings. E. R. Salmon. (Sheet Metal Industries, 1944, vol. 20, Sept., pp. 1582–1589, 1601; Oct., pp. 1772–1774). Taking the production of a mudguard as an example of the large pressings dealt with in automobile manufacture, the technique of deep-drawing is dealt with in detail. The relative merits of the wedge-draw and cup-draw tests for determining the drawability of sheet metal are compared and the requirements of a good drawing material are enumerated.

Device for Calculating Wire-Mill Problems. E. H. Webb. (Wire and Wire Products, 1944, vol. 19, May, pp. 291–293). A calculator of the revolving-disc type for solving problems which arise in the wire mill is described. The calculator is based on a wire-gauge system in which each advancing gauge number represents a reduction in area of the wire section of 20·568% with No. 1 gauge equal to 0·2834 in. in dia.

Modern Wire-Drawing Practice. (Metals and Alloys, 1944, vol. 20, July, pp. 73–77). A profusely illustrated description of the new wire mill at the South Works of the American Steel and Wire Co., Worcester, Massachusetts, is given.

Strong, Fine Wire, Non-Magnetic. H. H. Uhlig and J. J. Naughton. (Metal Progress, 1944, vol. 46, July, pp. 82–83). A procedure for drawing and carburising fine wire of high-manganese steel to produce a non-magnetic wire is described. Ingots were cast from a heat of electrolytic iron and electrolytic manganese; this steel contained 14·4% of manganese and 0·03% of carbon. Wire 0·02 in in. dia. was drawn from this and then carburised in a mixture of hydrogen and xylene. To do this the wire was heated to 1000° C. in a pyrex tube containing the gases; the heat was supplied by passing a 10·1-amp. current through the wire; interrupting the current acted as a quench. Heating for 1 min. caused the wire to take up 1% of carbon at the surface. Further heating for 3 min. at 1000° C. in dry hydrogen caused some of the carbon to diffuse towards the core. Drawing the 0·02-in. wire down to 0·01 in. considerably increased the tensile strength, but made the wire slightly magnetic; this could be prevented by pickling the wire before the drawing operation.

The Use of Nitrate Baths for the Continuous Patenting of Steel Wire for Highly Stressed Springs. H. Wüstl and F. Schwertner. (Stahl und Eisen, 1944, vol. 64, July 27, pp. 485–490). The use of salt baths instead of lead baths for patenting steel wire has already been reported (*see* Journ. I. and S.I., 1942, No. II., p. 210 A). In this paper tests are described in which

the properties of wire patented by the two methods are compared with special reference to surface defects and suitability for coiling to make spiral springs. In most cases the wire patented in a sodium-nitrate bath was found to be equal to the lead-patented wire. In some instances with thick wire, local surface defects were encountered. The cause and methods of preventing them are discussed.

Polishing Suitable for Sintered-Carbide Drawing Dies. W. Trurnit. (Stahl und Eisen, 1944, vol. 64, Aug. 3, pp. 503–506). When a worn die for wire-drawing is reground and polished to draw wires of larger diameter, the length of the cylindrical drawing surface is increased much more than the diameter; this length has therefore to be reduced in order that the length/diameter ratio remains unaltered. This is done by conical polishing tools the shape of which maintains the required angles of entry and exit. A polishing procedure suitable for sintered-carbide dies is described by means of which the travel of the tools can be pre-set to give the required length to the cylindrical portion and the required length/diameter ratio.

ROLLING-MILL PRACTICE

Mill Design. A. E. Lendl. (Iron and Steel, 1944, vol. 17, Aug., pp. 560–563). The formula by T. Ekelund has proved to be the most reliable one for calculating spread in rolling problems. Examples of the application of this formula are given, with explanations of the influence of temperature, roll diameter, initial width and degree of reduction.

Bearing Friction and Boundary Layer Phenomena. S. Kiesskalt. (Zeitschrift des Vereines Deutscher Ingenieure, 1943, vol. 87, May 29, pp. 321–324). The properties of lubricants which affect the seizing tendency of bearings are discussed with special reference to the relationship between the pressure and viscosity of the oil.

The Dimensions and Design of Plain Bearings. A. Klemencic. (Zeitschrift des Vereines Deutscher Ingenieure, 1943, vol. 87, July 10, pp. 409–418). The design of plain bearings is discussed with special reference to reducing the size without lowering the capacity, so as to reduce the amount of non-ferrous alloy required for their construction.

The Emulsion Lubrication of Large Gas Engines. H. Hungsberg. (Stahl und Eisen, 1944, vol. 64, July 20, pp. 465–469). Tests are described the object of which was to determine an economical method of preparing an emulsion suitable for lubricating large gas engines at an iron and steel works power-station. An efficient emulsion mixer is described. It was found that the gas-engine lubrication system could be modified to use emulsions with a saving of 45% in lubricating-oil consumption.

The Rolling of Semi-Finished Steel. R. E. Beynon. (Iron and Steel Engineer, 1944, vol. 21, Aug., pp. 37–54). Some mills in the United States for rolling blooms, billets, slabs and sheet bars are described, with details of the rolling practice, roll design and the casting of rolls.

Designing Cogging and Billet Mill Trains for Alloy Steel. F. Anke. (Stahl und Eisen, 1944, vol. 64, June 15, pp. 381–387). The increase in the demand for billets of alloy steel in Germany has changed the proportions of the structural and the alloy steel produced by the rolling mills. To cope with this requires changes in the design and layout of the soaking-pits, mill-trains and the ingot and billet-cleaning department. These aspects are discussed at length and diagrams of proposed layouts are presented.

Glass Processing Wire and Steel Products. J. J. Caugherty. (Wire and Wire Products, 1944, vol. 19, May, pp. 299–303). *See* Journ. I. and S.I., 1944, No. II., p. 149 A.

Modern Metal-Rolling Practice. C. E. Davis and L. R. Underwood. (Institute of Metals: Metal Industry, 1944, vol. 63, May 19, pp. 306–309; May 26, pp. 322–324). The design principles and the theory involved in rolling metals are discussed with special reference to the rolling of strip. The effects of strip tension, roll diameter and strip lubrication on the power requirements and rolling speed are explained.

Cold Rolling of Narrow Strip and Its Equipment Requirements. R. J. Wean. (Wire and Wire Products, 1944, vol. 19, May, pp. 295–298, 317). Brief descriptions are given of mill-stands, scale-breakers, shears and other modern equipment for producing narrow cold-rolled strip.

Cold Roll Forming. C. M. Yoder. (Iron and Steel Engineer, 1944, vol. 21, July, pp. 64–72). Modern slitting mills and cold-forming mills for sheet and strip are described and illustrated.

Tubes and Tube Making—Development of an Industry. B. P. Planner. (Iron Age, 1943, vol. 152, Dec. 16, pp. 48–53; Dec. 23, pp. 60–66; Dec. 30, pp. 54–56). Early forms of piercing mills, the principles of the Mannesmann process and the development of the Pilger mill are described in the first two parts of this paper. In the concluding part the cold-reduction of tubes through tungsten-carbide dies is discussed.

Tube Mills for Russia. (Steel, 1944, vol. 115, Aug. 14, pp. 104, 160). The salient features of one of two mills, designed to produce 1½–6-in. resistance-welded steel tubing, just completed in the United States for shipment to the U.S.S.R., are described. Each mill is fitted with a 750-kVA. alternating-current welding machine carrying wheel-type electrodes.

The Perfected "Metadyne." Some Applications to Rolling Mills. (Iron and Coal Trades Review, 1944, vol. 149, Oct. 6, pp. 499–501). A brief account is given of the principles and applications in rolling-mills of the Metadyne D.C. generator. In many applications the Metadyne acts as a form of exciter and is used for controlling another machine, circuit or mechanism.

Improved Control for Run-out Tables. B. Starie. (Iron and Steel Engineer, 1944, vol. 21, July, pp. 73–75, 79). The principles and advantages of the amplidyne control system are explained. The amplidyne is a motor-driven exciter of unique design, characterised by its extremely low control wattage (a ten-thousandth part of its output) and its response, which is fifty times faster than that of a conventional exciter. Its application to the control of a rolling-mill run-out table is described.

The New Lost-Time Concept of Refa. H. Euler. (Stahl und Eisen, 1944, vol. 64, July 6, pp. 438–442). A committee set up by the German Government for studying labour problems, known as "Refa" (*Reichsausschuss für Arbeitsstudien*), has published books from time to time. The present paper is an extract of a recent Refa publication in which a new concept for lost time in production is set out and methods of analysing lost time are explained, taking rolling-mill production as an example.

HEAT TREATMENT

Principles of Heat Treating of Steels. T. F. O'Brien. (Steel Processing, 1944, vol. 30, Apr., pp. 225–227, 250; May, pp. 297–298, 326; June, pp. 361–363, 383; July, pp. 436–439). The theory of the heat treatment of steel is explained with simple analogies, and a comprehensive table is presented giving the effects of a number of alloying elements on the carbide-forming tendency, depth of hardness and tempering of steel. In the second part the precipitation of carbides, recrystallisation and grain growth are discussed. In the third part the hardening and annealing of steel are

explained, with notes on furnace atmospheres. In the concluding part liquid heating media, pyrometry and quenching media are dealt with.

Wartime Developments in Heat Treatment of Steel and Their Effect on Design. H. W. McQuaid. (Society of Automotive Engineers: Steel, 1944, vol. 115, Aug. 21, pp. 123, 164–165). Some ways in which designers could take advantage of war-time developments in the heat treatment of steel are indicated. In isothermal heat treatment, transformation takes place at about the same time throughout the part, thus preventing high stresses; the designer can therefore assume that in many cases the grinding of surfaces to be subjected to high loads can be done before heat treatment or eliminated altogether.

Application of S-Curve Data to Heat-Treating Problems. R. L. Cunningham and A. Dube. (Canadian Metals and Metallurgical Industries, 1944, vol. 7, July, pp. 22–26). The paper contains information on the isothermal transformation of austenite and on the interpretation and uses of S-curves, gathered by the authors from the literature which has appeared on this subject since the paper by Davenport and Bain in 1930.

An Anlaysis of a Typical Carburizing Gradient. F. E. Harris. (Metal Progress, 1944, vol. 45, June, pp. 1111–1117). In connection with the problem of carburising steel, a typical concentration/depth gradient is analysed from the point of view of the range of concentrations from surface carbon to core carbon; the effect of this range on the character of the gradient is demonstrated quantitatively by carburising a test bar of steel S.A.E. 1020. The equations on which the mathematical theory of this paper is based were given in an earlier paper by F. E. Harris (*see* Journ. I. and S.I., 1944, No. I., p. 17 A).

Measurement of Case Depth by Hardness Gradient. T. C. Fazio. (Metal Progress, 1944, vol. 46, July, pp. 89–90). A method of determining the case depth of carburised steels is described. The method is most useful for investigations on a large number of specimens of the same dimensions. A gradual taper is ground on a specimen hardened in the standardised manner, and the Rockwell C hardness numbers are determined at short intervals along the surface. The readings thus give the hardness at different depths from the original surface.

Case-Hardening. L. Kolar and K. Jüthner. (Korrosion und Metallschutz, 1943, vol. 19, Dec., pp. 309–312). The theory of case-hardening is discussed, with detailed explanations of all the factors involved.

Case-Hardening. V. P. Sacchi. (Korrosion und Metallschutz, 1943, vol. 19, Dec., pp. 312–319). The theory of case-hardening is explained with notes on the effect of alloying elements, and descriptions are given of case-hardening processes using solid, liquid and gaseous carburising media.

The Penetration of Nitrogen and Carbon in Salt Baths. A. Kozelli. (Korrosion und Metallschutz, 1943, vol. 19, Dec., pp. 320–322). The case-hardening of high-speed steel and of carbon steel in baths of a German proprietary brand of cyanide salt is described, and diagrams are presented showing the depth and hardness of the case in relation to the immersion time and bath temperature.

The Mechanism of Carburising with Solid Carburising Materials. V. P. Sacchi. (Korrosion und Metallschutz, 1943, vol. 19, Dec., pp. 323–332). The mechanism of case-hardening is discussed with special reference to the carburising capacity of carbon monoxide alone, in conjunction with charcoal, and in conjunction with charcoal and an activating agent. Carbon monoxide is dissociated in the case-hardening box in accordance with the equation $2CO \rightleftharpoons CO_2 + C$ with the formation of a very small amount of elementary gaseous carbon at the outer surface of the case, and, because of its vapour pressure, this carbon dissolves in the γ-iron; the carbon in solution in the austenite also has a vapour pressure, and it is therefore

fed by the carbon formed in accordance with the above equation. The decomposition of the carbon monoxide and the solution of the carbon in the γ-iron continue until the vapour pressure of the carbon in the austenite at the surface of the metal balances that of the elementary carbon. Whether a given CO–CO_2 mixture at a given temperature has a carburising or decarburising effect depends on the amount of carbon in the austenite at the surface of the metal. The literature on the function of charcoal in carburising compounds is reviewed.

Springs of Case-Hardened Mild Steel. A. M. Borzdika. (Iron Age, 1944, vol. 154, July 27, pp. 49–51). A translation of a paper which appeared in the Russian journal "Stal" is presented. Difficulties were experienced in the supply of high-carbon spring steel, and a method is described for the case-hardening and subsequent heat treatment of low-carbon steel in which solid carburising compounds containing crushed crab shells and fish-scales were used.

Liquid Carburizing of Transmission Gears. W. A. Silliman. (Metals and Alloys, 1944, vol. 20, July, pp. 58–61). Owing to distortion during carburising and heat-treating tractor gears at the Cleveland Tractor Co., Cleveland, Ohio, rejections could not be kept at an economically low level. The problem was solved by treating the parts in electric salt-bath furnaces; the operation of one such furnace is described.

Gas Case-Hardening. A. D. Assonov. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 72–76). (In Russian). The process developed by the Central Scientific Research Institute for Machine Construction employs a gas mixture obtained by the pyrolysis and cracking of paraffin. The plant used for the purpose and the gas-fired muffle furnace in which the case-hardening is done are described. The paraffin is pyrolysed at 830° C. and, after scrubbing, a portion of the products are cracked at 970° C. in the presence of water vapour. The products of the two processes are then mixed to produce an atmosphere consisting of: C_nH_{2n} 2–5%, carbon monoxide 12–16%, methane 10–15%, oxygen 0–0·5%, carbon dioxide 0·2–0·5%, hydrogen 66–70% and the remainder nitrogen. The case-hardening furnace is operated at a temperature of 930° C. at the input end and 900° C. at the unloading end. It has an output of up to 300 kg. per hr. The case-hardening capacity is double that obtainable with solid compounds.

Electronic Control for Resistance Furnaces. H. J. Hague. (Steel, 1944, vol. 115, Aug. 14, pp. 106–108, 162–170). A method developed for the accurate control of the temperature of electric-resistance heat-treatment furnaces is described in detail. The equipment involved includes a saturable core reactor, an electronic regulator and a pyrometer with a motor-operated slide-wire rheostat.

Surface Hardening with Oxy-Acetylene Burners. B. M. Davidson and S. D. Rekhter. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 85–92). (In Russian). Particulars are given of the equipment used at a Russian works for the oxy-acetylene hardening of gear-wheel teeth and other parts. The control of the process, and experience and results obtained in its operation are described.

High-Frequency Heating. A. G. Robiette. (Iron and Coal Trades Review, 1944, vol. 149, Oct. 6, pp. 491–496). The fundamental theoretical and practical aspects of the high-frequency electric heating of metals are reviewed, together with the metallurgical effects of such heating, and brief consideration is given to the economics of the process.

Megacycle Induction Heat for Thin Cases. V. W. Sherman. (Metals and Alloys, 1944, vol. 19, June, pp. 1409–1414). The application of currents of very high frequency (over a million cycles per sec.) for case-hardening to a depth of about 0·01 in. is discussed. The method is particu-

larly useful for hardening instrument parts. No distortion or scaling occurs, and the properties of the core of the part remain unchanged.

Mass Hardening of Small Parts by the Induction Method. V. P. Vologdin and V. L. Sergeevich. (Vestnik Metallopromyshlennosti, 1940, No. 11–12, pp. 57–59). (In Russian). An apparatus for the individual high-frequency hardening of sewing-machine needles is described. The fall of each needle through a tube surrounded by the high-frequency induction coil is arrested by a permanent magnet, the needle being held in position until its temperature reaches the Curie point of about 720° C. It is then released and, while falling through the remaining portion of the induction coil tube, it is heated another 60–80° C. to the quenching temperature. It then falls into the quenching oil. The whole process takes not more than 0·3 sec.

Practical Aspects of Induction Heating. W. M. Roberds. (Iron Age, 1944, vol. 154, Aug. 24, pp. 50–55). The theory and practice of the induction heating of the surface of metal parts are discussed. A formula for calculating the depth of penetration in relation to the current density is presented, and the design of inductor coils and leads for heating parts of unusual shape is dealt with.

Problems in the Heat Treatment Shop. A. C. Schrapes. (Australian Institute of Metals: Australasian Engineer, 1944, vol. 44, July 7, pp. 29–58). Three specific heat-treatment problems are dealt with in detail. They are: (1) The heat treatment of carburised nickel-molybdenum steel; (2) nitriding hardened high-speed steel; and (3) the heat treatment of chromium ball-race steel. The high-speed steel was nitrided by immersion in a cyanide salt bath, and it was shown that no distortion took place, provided that the tool was fully tempered and properly hardened. To ensure complete stability of the steel multiple tempering was necessary; a single treatment, even if prolonged, resulted in instability of the product.

New Annealing Machine Developed for Steel Cartridge Cases. (Industrial Heating, 1943, vol. 10, July, pp. 966–970). Three annealing machines developed by the Morrison Engineering Company, Cleveland, Ohio, for annealing 37-mm., 40-mm., 57-mm., 75-mm., 90-mm., and 105-mm. steel cartridge cases are briefly described.

Localised Heat-Treating Machine for Mouth-End Annealing Steel Cartridge Cases by Radiant Gas Heat. (Industrial Heating, 1943, vol. 10, Aug., pp. 1126–1128). A machine for the continuous annealing of steel cartridge cases is described. The case is revolved as it passes through the machine and the mouth-end is annealed by means of a series of ceramic-cup radiant gas burners which are automatically lowered and raised as the cartridge cases pass beneath them.

Bright Annealing of Steel Strip and Parts. I. S. Khinsky and I. S. Vinner. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 77–84). (In Russian). A detailed description is given of the plant put into operation in 1939 at a bicycle works in Moscow. It comprises two coal-gas-fired pit-type furnaces and five cooling pits arranged in line. The protective atmosphere, consisting of nitrogen 90% and hydrogen 10%, is obtained by dissociating and then partly burning ammonia. A safety feature of the equipment is the low pressure (600 mm. of water) of the ammonia in the evaporator for the liquid ammonia.

Turret Ring Line Features Unique Heat-Treating Method for Circular Parts. G. W. Birdsall. (Steel, 1944, vol. 115, July 24, pp. 82–88). Descriptions and illustrations are given of special machines for forming, butt-welding and heat-treating tank-turret rings which are 54 in. in dia. with a section $3\frac{7}{16} \times 1\frac{5}{16}$ in. For hardening, the ring is mounted in a vertical plane on two small rollers. The descending side the revolving ring gasses a preheating burner and then through a box containing six gas-

burners arranged to impinge on all four faces ; leaving the box, the ring passes between two spray heads which provide the quench, and it then moves past two pairs of burners for stress relieving. The ring moves at 2¼ in. per min. There is no distortion after this form of treatment. Machines of this type can be easily adjusted to suit rings of different sizes.

Heat-Treating Machine-Gun Links. J. Ade. (Metals and Alloys, 1944, vol. 18, Dec., pp. 1339–1341). An illustrated description is given of a continuous heat-treatment plant for machine-gun links. This consists of four lines, each comprising a muffle-furnace in which the links are heated to 1550–1600° F., an oil-quenching bath and a tempering furnace for reheating the parts to 660–690° F. The capacity of each line is 1100 lb. per hr.

Interrupted Quench in Salt. (Steel Processing, 1944, vol. 30, June, pp. 379–382). Three methods of heat treatment are described, namely, austempering, isothermal quenching and " martempering." The first two have frequently been described in the literature. In martempering, the parts are heated above the critical point and quenched in a salt bath at about 400° F., in which they are kept sufficiently long for them to cool to the bath temperature at the centre. They are then cooled in air.

Importance of Design and Mass in Water Quenching Steel Castings. W. J. Phillips. (Foundry, 1944, vol. 72, Aug., pp. 89, 201–204). This was one of the papers presented in a symposium on the water-quenching of steel castings held recently by the Steel Founders' Society of America (*see* Journ. I. and S.I., 1944, No. II., p. 153 A).

The Cold-Treatment of High-Speed Cutting Tools. F. W. Whitcomb. (Machinery, 1944, vol. 65, Sept. 28, pp. 350–351). A brief review is given of the three principal methods of cold-treating high-speed steels with notes on the results of such treatment on milling cutters.

Shrink Fits. C. S. Lucas. (Steel, 1944, vol. 115, July 17, pp. 110–112, 160–162). The application of low-frequency induction heating to expand aero-engine parts so as to shrink them on to bushes or spindles is described. The current is 60-cycle A.C. and the heating time required is less than 5 min. Data on the coil design, current consumption, temperature attained and time required are presented.

WELDING AND CUTTING

The Metallurgy of Arc Welding. H. Allsop. (Institute of Welding : Shipbuilder and Marine Engine-Builder, 1944, vol. 51, Sept., pp. 319–327). The factors affecting the deposition of sound weld metal and the influence of welding conditions on the properties of various steels are discussed, and the similarity between the behaviour of steel poured into an ingot mould and weld metal deposited on steel is dealt with. The properties of the weld metal may be modified to some extent by welding technique, but the mode of crystallisation cannot be altered, and the radiating type of structure is inseparable from single-run welds.

Hard-Surfacing by Welding. C. R. Whittemore. (Canadian Welding Society : Canadian Metals and Metallurgical Industries, 1944, vol. 7, June, pp. 42–46, 61 ; July, pp. 39–42). The oxy-acetylene and the electric-welding process of depositing hard wear-resisting alloys on the surface of worn parts are described in detail, and the specification established by the Deloro Smelting and Refining Co., Ltd., for the manufacture of cast welding rods is given.

Hard-Surfacing. Applications and Techniques. (S.A.E. Journal, 1944, vol. 52, June, pp. 25–32). The equipment and procedure for hard-sur-

facing by welding a hard, wear-resisting alloy on to parts subjected to very severe conditions of service, are described. The applications for which details are given include valve faces, the teeth of tractor sprockets, the lips of power-driven shovels and the shoes of caterpillar-driven machinery.

Spot Welding High-Tensile Steels with Automatic Post-Heat Treatment. W. S. Simmie and A. J. Hipperson. (Sheet Metal Industries, 1944, vol. 20, Sept., pp. 1609–1619, 1624). The object of the investigation here reported was to determine the extent to which spot welds in certain high-tensile steels could be made sufficiently free from brittleness for structural purposes by passing a post-heat-treatment current of short duration after the actual welding cycle. A U-test was developed to determine the strength of the welds; in this, two U-shaped pieces were placed base to base and spot-welded together and then pulled apart in a tensile-testing machine. The properties of all the spot welds were greatly improved by the automatic tempering treatment. The U-test was the quickest and best quantitative test for discriminating between brittle and ductile spot welds. In the case of a nickel-chromium-molybdenum steel the maximum value obtained in the U-test after tempering was four times that without the treatment. Considerable latitude in the cooling time is possible as long as it is above the minimum time required for the weld to cool to martensite-forming temperatures. Some suggestions for applying the technique for production work are made.

Weldable Steel Castings. W. J. Crook. (Welding Journal, 1944, vol. 23, June, pp. 257-S–271-S). An investigation at Rock Island Arsenal is described, the object of which was to establish a specification for a cast steel with a yield strength of 65,000–70,000 lb. per sq. in. which could be incorporated in welded plate structures of similar strength under conditions which prevented preheating the welds. The weldability test applied was to deposit a short single bead of metal down the centre of cast plates measuring $6 \times 3 \times \frac{1}{2}$ in. and to make hardness traverses across the plate; if the maximum hardness at any point exceeded a certain value, the steel was rejected.

The Dynamic Properties of Flash-Welded Tubing. R. B. Bland and P. E. Sandorff. (Welding Journal, 1944, vol. 23, June, pp. 280-S–302-S). An investigation of the strength of flash welds in tubes of chromium-molybdenum steel is described. The tubes were $1\frac{1}{2}$ in. in inside dia., with a 0·065-in. wall. Tensile impact tests were applied by a falling weight, and recordings of the elongation and fracture of the specimens were obtained with a high-speed cinema-camera. Flash welds in normalised tubes of this steel were satisfactory in both static and dynamic tensile tests. Notches and abrupt changes in section were very detrimental. The total energy absorbed in fracturing the tubes was greater in dynamic loading than in static loading. Details of the test apparatus, the derivation of energy formulæ and particulars of some fatigue tests on tubes of this steel are given in appendices.

Resistance Welding—No. 2: Operating Practice for Various Alloys. K. Rose. (Metals and Alloys, 1944, vol. 19, Feb., pp. 357–363). Continuation of a series of articles (*see* Journ. I. and S.I., 1944, No. I., p. 90 A). Recommendations are made together with data on current strength, mechanical pressure and welding speed for spot-welding low-carbon steel, 18/8 stainless steel and aluminium and magnesium alloys.

Mechanical Characteristics of Resistance Welds in Plain Carbon Steels. W. Spraragen and M. A. Cordovi. (Welding Journal, 1944, vol. 23, July, pp. 305-S–343-S). A comprehensive review is presented of the American literature up to Jan. 1, 1944, and that of other countries up to Jan. 1, 1941, on the mechanical properties of spot welds, seam welds,

projection welds, resistance butt welds and flash butt welds. A bibliography containing 88 references is appended.

Metallic Arc-Welding Electrodes. H. Lawrence. (Welding Journal, 1944, vol. 23, July, pp. 610–621). *See* Journ. I. and S.I., 1944, No. II., p. 157 A.

Characteristics of the Chief Makes of Soviet Welding Electrodes with Thick Coatings. A. S. Ogievetskiy. (Vestnik Metallopromyshlennosti, 1940, No. 6, pp. 45–50). (In Russian). The properties of eight different makes of coated electrodes for welding low-carbon steels are discussed. Particulars are given of the compositions of the coatings and the mechanical properties of the weld metal.

Deviations in Predicted Weldability by Lehigh and Rensselaer Systems. D. Rosenthal. (Metal Progress, 1944, vol. 46, July, pp. 67–68). In an earlier paper M. Seyt described two methods of assessing the weldability of steel (*see* Journ. I. and S.I., 1944, No. II., p. 49 A). In this paper it is shown that differences in the weldability predictions by the two methods are likely to occur at slow cooling rates.

Hot Cracks *versus* Cold Cracks—To Distinguish Them in Welded Joints. G. H. Boss. (Metal Progress, 1944, vol. 46, July, pp. 68–69). Two kinds of cracks occur in welded steel. These are "cold cracks" or "hard cracks" and "hot cracks." The susceptibility of a steel to the former depends on its hardenability and rate of cooling after welding; the cold cracks occur at below 600° F., and the cracked surfaces are always bright, and they may or may not pass into the weld metal. The surfaces of hot cracks are always covered with blue-black magnetic oxide, and the cracks probably occur at above 1800° F. They are in the heat-affected zone, and never enter the weld metal.

Weldability—A More Complete Definition. O. I. Temper. (Metal Progress, 1944, vol. 46, July, pp. 69–73). The difficulty in assigning any one cause for cracks in welds and in relating weldability to any one property is discussed. Tables are presented in which welding conditions and properties of the parent metal and the weld metal are related to the sensitivity of the welds to cracking.

Testing of Alloy Steels for Tendency to Develop Cracks in Welding. S. K. Zvegintsev. (Vestnik Metallopromyshlennosti, 1940, No. 7, pp. 67–71). (In Russian). Short single beads were deposited on steel plates under different conditions of cooling and rigidity and were then examined for cracks. The rate of cooling was varied by using plates of different thickness and by water-cooling. Rigidity was obtained by machining a circular depression in the metal plate, the unreduced thickness of the metal around the depression ensuring rigidity. The bead was deposited on the bottom of the depression. In another test the weld metal was deposited in a V-shaped slot, 50 mm. long, cut in the middle of a steel plate. Observations made in such tests on low-alloy and carbon steels are described.

The Stress-Free Annealing of Electric-Welded Beams of Structural Steel St 52. C. M. Bohny and H. Busch. (Stahl und Eisen, 1944, vol. 64, June 8, pp. 365–372; June 15, pp. 387–391). The failure of beams used in bridges and other structural work has been found to be due to multiaxial stresses set up by electric welding. An account is given of tests carried out on two electric-welded plate girders of high-tensile steel, 6 m. long × 1 m. deep, with flange plates 35 mm. thick, one of which was annealed. Deflection measurements were made with increasing loads. The annealed girder buckled after considerable deformation. The unannealed one broke with no permanent deformation. Supplementary tests were made on specimens cut from the girders so as to determine the magnitude of the welding stresses before and after annealing, the peak hardness in the heat-affected zone and the nature of the cracks. High-tensile

basic-Bessemer rimming steel can be used in the construction of plate girders to a limited extent.

Methods of Stress Relieving. W. E. Ryan. (Welding Journal, 1944, vol. 23, June, p. 536). The methods of relieving stresses in welds are described. These include peening, stressing the metal to a little above the yield point, heating to 1100–1250° F., cooling slowly and full annealing at about 50° F. above the critical temperature.

Repair and Maintenance of Tools. T. B. Jefferson. (Welding Journal, 1944, vol. 23, June, pp. 506–511). The technique for silver-soldering, brazing and atomic hydrogen welding for repairing tools is described.

The Brazing of Low-Alloy Tool Tips Using Powdered Brazing Alloys. N. A. Minkevich, N. A. Bukhman and Yu. A. Geller. (Vestnik Metallopromyshlennosti, 1940, No. 8–9, pp. 45–57). (In Russian). The quenching temperatures of low-alloy chromium-tungsten-molybdenum-vanadium tool steels range from 1175° to 1240° C., and are thus below the melting points of the materials generally used for brazing on tungsten high-speed steel tips. General requirements which brazing alloys for this particular purpose have to satisfy are considered. Ferro-manganese (75%) was selected as a brazing material, its melting point being controlled between 1100° and 1280° C. by additions of up to 40% of copper. Two copper-nickel alloys were also tested. Furnace brazing and electrical resistance brazing were employed. Shear tests and heat-treatment and cutting tests were made on the brazed-on tips. The best results were given by a ferro-manganese brazing alloy containing 20–25% of copper. The copper-nickel alloys gave a higher shear strength, but their behaviour during subsequent heat treatment of the tools was unsatisfactory. Attempts to quench tools directly from the brazing temperature resulted in reduced cutting efficiency.

The Tipping of Cutting Tools with Hard Materials. N. A. Minkevich, Yu. A. Geller and E. I. Belikova. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 38–52). (In Russian). A number of alloy compositions for the hard-tipping of cutting tools and different methods of applying these materials were studied in respect of the quality and structure of the tips and their cutting properties. Cutting tips produced by welding contained about 4% of carbon, 15·5% of chromium, 6% of molybdenum, 6% of vanadium, 1–2% of silicon and 0·75% of manganese. Either soft iron electrodes with a ferro-alloy coating or cast alloy electrodes with a chalk/water-glass coating were employed. The structure of the deposited material was that of a ledeburite eutectic with large primary carbides. Another alloy of the same type, but containing less carbon and more vanadium, gave a more uniform and finer structure. Welded tips were difficult to grind to shape as, for this purpose, their hardness could not be appreciably reduced by annealing. Chill-cast tips of the above alloys and of less highly alloyed compositions were produced for attachment to the tool shanks by brazing. These tips tended to have casting faults. Cutting tests on the welded and cast tips showed them to be, on the whole, inferior to high-speed tungsten steel. Preference is to be given to forged tips of low-alloy steel with carbon 0·9–1·0%, chromium 4·5–5·3%, molybdenum 3·5–4·0%, vanadium 2·0–2·5%, silicon 0·4% max. and manganese 0·4% max., attached to the tool shank by mechanical means. These tips proved to be equivalent, and in some respects even superior, to tungsten high-speed steels.

Underwater Cutting of Metal. M. F. Rodman. (Welding Journal, 1944, vol. 23, July, pp. 603–609). A comprehensive survey dealing with the difficulties, basic principles and application of the oxy-acetylene torch for cutting metals under water is presented.

MACHINING

Shaved Aircraft Gears. I. E. Calicoat. (Iron Age, 1943, vol. 152, Dec. 16, pp. 44–47). To prevent distortion in the heat treatment of gears for aero-engines, quenching in a press has been resorted to. A machining procedure is described by which the gear teeth are shaved before heat treatment to dimensions which allow for the amount of distortion which will occur. The shaving operation is much faster than grinding.

The Use of Super-High Cutting Speeds in the Machining of Metals. G. M. Yakovlev. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 114–116). (In Russian). Machining tests at Krupps have shown that as the cutting speed is increased the temperature of the cutting tool reaches a maximum and then falls off. In some milling tests reported by the author, aluminium, bronze, steel and cast iron were cut at speeds of 105 and 140 m. per sec. With bronze and aluminium these speeds appeared to be beyond the critical speed. The temperature rose only slightly, and there was no observable wear of the cutting edges of the tool. With steel and cast iron excessive heating took place with rapid wear ; even melting of the cutting teeth was observed.

The Possibility of the Super-High-Speed Cutting of Metals. V. D. Kuznetsov. (Vestnik Metallopromyshlennosti, 1940, No. 7, pp. 27–35). (In Russian). Theoretical considerations confirmed by experimental results indicate that the energy consumption in cutting is primarily made up of the energy consumed in plastic and elastic deformation. With increasing speed a metal tends to behave more and more as a brittle material, with plastic deformation becoming less and less, and consequently the total energy consumption in the cutting process decreases. Turning tests on steel at 1500 m. per min. showed that little heat was generated. At 2200 m. per min. the life of the carbide tool was practically zero, for the mild steel cylinder ground the tip away without itself suffering any change. When machining cast iron and aluminium with a milling cutter at a peripheral speed of 1700–2300 m. per min. these metals behaved like brittle materials, whilst carbon steel and copper did not.

The Cutting Properties of Different Types of High-Speed Steels. A. P. Gulyaev and S. I. Krestnikov. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 93–96). (In Russian). The results of machining tests with the following types of tool steels are reported : (*a*) High-tungsten steels ; (*b*) steels with 9–13% of chromium, 1–2% of vanadium and about 1% of silicon with and without additions of 2–5% of tungsten ; and (*c*) steels with 2–5% each of molybdenum, vanadium and chromium with and without additions of about 3% of tungsten. On the whole the high-chromium steels proved to be inferior to the low-alloy compositions, whilst the latter were equal to tungsten high-speed steel, for which they may be regarded as fully equivalent substitutes.

Experimental Work on the Superfinishing of Bearing Races. S. M. Kedrov. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 53–67). (In Russian). Different Russian abrasives used in finishing the surfaces of bearings were tested with a view to selecting the most suitable material and determining the optimum operating conditions and lubricants, the best way of preparing the surface, and the dimensional changes involved in superfinishing.

Tool Steel Identification Systems. (Iron Age, 1944, vol. 153, Jan. 13, pp. 67, 139). The three largest automobile manufacturers in Detroit have agreed upon a code system for identifying tungsten and molybdenum high-speed steels. Each code is made up of letters and numbers which indicate the analysis of the steel and the manufacturer's name.

Dust-Collecting Systems. F. C. Morey. (Iron and Steel Engineer, 1944, vol. 21, July, pp. 45–51). The design of equipment for removing dust in such operations as grinding, polishing, sand-blasting, crushing and mixing is reviewed.

CLEANING AND PICKLING OF METALS

Pickling. F. Taylor. (Iron and Steel, 1944, vol. 17, July, pp. 525–527). The composition of oxide scale and the theory of pickling are discussed. The advantages and disadvantages of sulphuric acid and of hydrochloric acid solutions are pointed out, with explanations of the theory of inorganic and organic inhibitors. Brief details of the following processes are given: (1) De Lattré's process, which depends upon the reversible reaction

$$FeCl_2 + H_2SO_4 \rightleftharpoons FeSO_4 + 2HCl;$$

(2) a process developed by Hanson Van Winkle Munning Co., in which the material is first passed through a sulphuric acid solution and electrolysed cathodically, then into another sulphuric acid solution where it is electrolysed anodically, and finally neutralised in weak sodium carbonate solution; and (3) the Bullard-Dunn process.

The Pickling of Steels—A Survey of the General Principles. T. P. Hoar. (Sheet Metal Industries, 1944, vol. 20, Sept., pp. 1533–1539; Oct., pp. 1715–1717). An attempt is made to give an adequate treatment of the physico-chemical principles involved in ferrous pickling, based on a survey of the specialised pickling literature for the last twenty years and on modern general theory.

Pickling Plants Giving a Swinging and Rolling Motion to the Material. W. Freudenthal. (Stahl und Eisen, 1944, vol. 64, Apr. 13, pp. 233–240). Illustrated descriptions are given of modern pickling plants, particular attention being drawn to the mechanism for moving the material while it is immersed in the bath. Where adequate movement is provided the pickling time is only half that required when there is no movement. In one type a rocking beam is pivoted between two baths, the charges being suspended from the ends of the beam; this gives the charges an up-and-down movement. Another arrangement of cranks provides a movement in a horizontal plane, and in a third type the suspended baskets are swung like a pendulum.

Pickling Plant Developments and the Problem of Planning Lay-Outs. (Wire Industry, 1944, vol. 11, Oct., pp. 499–501). The planning of installations for pickling coils of wire is discussed. The advantages and disadvantages of pickling with hydrochloric acid and sulphuric acid are compared, and the lay-out of a plant for a high rate of production is described.

Light-Weight Long Life Pickling Equipment from Welded Wrought Materials. H. R. Strohecker. (Steel Processing, 1944, vol. 30, June, pp. 358–360). Pickling baskets and racks made of welded Monel metal and stainless steel, which are lighter than cast baskets, are described and illustrated.

Protective Chemical Coatings on Metals. J. R. Ewing. (Metal Progress, 1944, vol. 45, May, pp. 894–895). A mixture of organic chemicals, distillates and solvents has been developed which, in one operation, removes grease, oil and other foreign matter from metal surfaces and leaves them immune from atmospheric corrosion for a short period.

COATING OF METALS

Precision Selective Plating. C. T. Wells. (Metal Finishing: Metal Industry, 1944, vol. 65, Aug. 18, pp. 106–108). The development of the technique of "stopping off" to prevent particular surfaces of parts from being plated is reviewed, with descriptions of several examples.

Chromate Passivation. F. Taylor. (Metal Industry, 1944, vol. 65, Sept. 8, pp. 149–150). The corrosion resistance of zinc and cadmium coatings on steel can be increased by passivation, which is done by immersion in a mixture of sodium dichromate and sulphuric acid. Three processes—namely, the Cronak, Passival and Rushtol Passivation "A" processes—are briefly described.

Engineered "Erosion." G. W. Birdsall. (Steel, 1944, vol. 114, June 19, pp. 82–83, 156–160). A description is given of the Van der Horst process of producing porous chromium-plated surfaces. Such surfaces will hold lubricating oil, and thus reduce the friction on cylinder liners (*see* Journ. I. and S.I., 1944, No. II., p. 84 A).

The Wear Resistance of Chromium Electrodeposits. V. I. Arkharov, A. M. Zagrubskiy and S. A. Nemnonov. (Vestnik Metallopromyshlennosti, 1940, No. 10, pp. 13–15). (In Russian). Tests were made to determine the best method of producing good wear-resisting chromium plating. The best wear-resistance was obtained when an electrolyte containing 150 g. of CrO_3 and 1·5 g. of H_2SO_4 per litre of water was used at 50° C. with a current density of 40 amp. per sq. dm. The wear-resistance increased with improvement in the degree of perfection of the octahedral texture of the deposit. Thus, X-ray examination provides a non-destructive means of checking the wear-resistance of chromium coatings.

Dimensionally Accurate Hard Chromium Plating of Machine Parts. D. V. Pletnev. (Vestnik Metallopromyshlennosti, 1940, No. 6, pp. 29–38). (In Russian). The hard chromium plating of internal or external cylindrical surfaces of parts suspended vertically in a chromium plating bath is considered. In order to obtain a deposit of uniform thickness over the whole of the surface, the anode must be closer to the cathode at the bottom of the bath than at the top. A formula for calculating anode-to-cathode distances is derived, and its use is demonstrated in several examples. The economic aspects of the process are considered.

The Production of Chromium Coatings by Diffusion. H. Bennek, W. Koch and W. Tofaute. (Stahl und Eisen, 1944, vol. 64, Apr. 27, pp. 265–270). The theory and practice of chromium diffusion by liquid and gas processes are discussed. In the liquid or salt-bath process the bath concentration can be controlled at any desired value by adding alkaline-earth chlorides which do not interfere with the chromium-iron exchange reactions. The amount of chromium which diffuses into the iron can be adjusted as desired either by changing the composition of the bath, or the $CrCl_3/CrCl_2$ ratio, or by adding vanadium chloride. In the gas process the difficulty has been to find a carrier which would mix properly with the heavy chromium chloride gas. A suitable carrier has been found. The pressure at which the gas process is carried out depends on the sum of the partial pressures of the miscible components of the atmosphere. This sum is, however, always small, and lies between 20 and 200 mm. of mercury; the pressure is therefore below atmospheric pressure, so that a completely enclosed muffle of austenitic stainless steel capable of withstanding a reaction temperature of 1000° C. must be used. In the gas process the chromium penetrates to a uniform depth even on the inside of small holes. The results of corrosion tests in acids were very good. Any welding should

be carried out if possible before chromium diffusion, as only thin sheets of chromium-impregnated steel can be satisfactorily welded. Welds in thick material do not maintain their corrosion resistance.

Salvage and Recovery of Over-Machined Parts. (Iron and Coal Trades Review, 1944, vol. 149, Aug. 11, pp. 185–187). Information is given from a Memorandum issued by the Ministry of Supply on the building up of over-machined steel surfaces by the electro-deposition of heavy coatings of nickel or chromium.

The Effect of the Carbon Content of Steel Sheets on Their Galvanising Properties. W. Püngel. (Stahl und Eisen, 1944, vol. 64, Feb. 17, pp. 101–105). An investigation of the effect of the carbon content of steel sheets on the uniformity, thickness, adhesion and pliability of zinc coatings applied by the hot-dip process is reported. When the steel was cleaned by sand-blasting, the carbon content had no effect, but when the sheets were hot-dipped after cleaning by pickling only, non-uniform coatings showing dark patches were produced, the quality deteriorating with increasing carbon content. This was probably due to non-uniform attack by the pickling solution. The carbon content had no marked effect on the coating thickness. The pliability of the coating in folding and deep-drawing tests was slightly less with high-carbon steel sheets. Increasing the carbon increased the resistance to repeated bending, and cleaning by sand-blasting also improved this property. Generally speaking, the galvanising properties are not impaired by a high carbon content in the steel, provided that the surface is thoroughly cleaned before hot-dipping.

The Formation of Hard Zinc in the Lead-Zinc Galvanising Process. H. Bablik. (Stahl und Eisen, 1944. vol. 64, Feb. 24, pp. 120–122). In the lead-zinc galvanising process the bath is filled with molten lead; a portion of its surface is walled off with refractories, and this portion is covered with molten zinc; the material to be galvanised is dipped in the lead and withdrawn through the molten zinc. It was claimed that this process prevented the formation of the iron-zinc compounds which accumulate at the bottom of zinc baths, causing an appreciable loss of zinc. It is now shown that this claim is not valid, and that the iron-zinc compounds are formed in the lead-zinc process just as in the all-zinc bath, unless the proper precautions are taken.

Copper in Galvanizing Zincs and Galvanizing Baths. W. G. Imhoff. (Wire and Wire Products, 1944, vol. 19, June, pp. 353–357, 390). The effects on the quality of zinc coatings of the presence of various amounts of metal impurities are considered from the practical point of view. Special attention is paid to the effects of copper which finds its way into the galvanising zinc accidentally, *e.g.*, from remelted zinc or from a hydrochloric acid bath previously used for pickling copper coils. It is pointed out that the quality of the zinc coating may be improved when not more than 0·25% of copper is present. Large amounts of impurities are harmful, and in certain types of galvanised product, *e.g.*, strip, small additions cause the zinc spangles to drop off.

Current Zinc Electroplating Practice. A. Bregman. (Iron Age, 1943, vol. 152, Nov. 11, pp. 46–53; Nov. 18, pp. 64–66). The following methods of electrogalvanising are described: (*a*) Acid zinc plating; (*b*) rapid zinc plating with zinc-chloride/zinc-acetate solutions; (*c*) cyanide zinc plating, including bright and semi-bright processes; and (*d*) the Hubbell-Weisberg process. The rapid process (*b*) has been developed by the Hanson-Van Winkle-Munning Co.; the solution consists of zinc chloride and zinc acetate without addition agents, and with zinc-aluminium alloy anodes no corrosion takes place when plating is not in progress. Steel sheets 3–6 ft. wide can be electrogalvanised at a current density of 400–500 amp. per sq. ft. by the rapid process. The advantages of the cyanide bath for zinc

plating are the high throwing power, the ability to obtain semi-bright or mirror-bright deposits, the relatively low cost of the equipment, and that plain steel auxiliary anodes can be used for plating in holes and recesses; the disadvantages are that the rate of deposition is lower than with acid baths, and that both the initial and the maintenance costs of cyanide solutions are higher. Nine formulæ for cyanide baths are presented.

Re-Galvanizing of Welded Joints. (Iron Age, 1944, vol. 153, Mar. 16, pp. 66–67). A method is described whereby metal surfaces may be quickly and conveniently regalvanised at places where the original galvanising coating has been burned off by welding. A stick of low-melting-point galvanising alloy is applied to the welded area while it is still hot. A number of further applications of this "Galv-Weld" alloy are mentioned.

Development of the American Tin-Plate Industry. H. A. Knox. (Iron Age, 1943, vol. 152, Nov. 18, pp. 44–47; Nov. 25, pp. 60–65; Dec. 2, pp. 64–70). The author describes the discouraging beginning of the American tinplate industry, its rapid growth in the latter part of the nineteenth century and its present status.

Metallic Substitutes for Hot-Dipped Tinplate. R. H. Lueck and K. W. Brighton. (Industrial and Engineering Chemistry, Industrial Edition, 1944, vol. 36, June, pp. 532–540). The electrolytic tinning and the Bonderising of steel strip are described, and the results of tests with various foods on the durability of cans protected by these methods are discussed. Bonderised steel has performed satisfactorily on the ends of cans for relatively non-corrosive products, such as lima beans, peas, corn, meats and some kinds of fish. Electrolytic tinplate has not fully met the expectations as to its suitability for the mildly corrosive fruits and vegetables.

Continuous Electroplating of Steel Strip. D. A. Swalheim. (Iron and Steel Engineer, 1944, vol. 21, Aug., pp. 55–60). An illustrated description is given of a continuous electro-tinning plant for tinning steel strip up to 38 in. wide at speeds up to 1350 ft. per min. A halogen electrolyte developed by E. I. Du Pont de Nemours and Co., is used.

High-Frequency Generators for Fusion of Tin Plate. H. G. Frostick. (Steel, 1944, vol. 115, July 31, pp. 90–94). To ensure the uniformity of electrolytically deposited tin coatings on steel sheet and strip, the coating is heated momentarily to fusion temperature by passing the material through an induction heating coil. The essential features of an induction heater for this purpose are described. Several items of the equipment came from a dismantled radio transmission station. The adjustment and maintenance of this type of equipment are discussed.

Chemical Coatings on Steel. V. M. Darsey. (Metal Progress, 1944, vol. 45, May, pp. 895–897). Phosphate and oxide coatings on steel are discussed. A treatment called "Parco Lubrizing" has been developed for producing a wear-resisting coating on bearing surfaces; it consists mainly of a mixture of iron and manganese phosphates. If steel is given a very light coating of zinc by electrolysis and this is phosphatised, a high resistance to corrosion is obtained.

New Process for Corrosion-Proofing Steel Sheets. O. E. Brown. (Steel, 1944, vol. 115, July 3, pp. 90–92). Results of salt-spray and other corrosion tests on steels treated with "Permyron," phosphate and oxide coatings are reported. Permyron is an inorganic material giving a black, lustrous appearance to steel. The surface to which it is applied must be chemically clean. Its resistance to salt spray and to dilute and concentrated acids is far greater than that of oxide, chromate and phosphate coatings.

Protective Coatings for Aircraft Parts. (Iron Age, 1944, vol. 154, Aug. 17, pp. 72–73, 184). Some details are given for the Parkerising and

Bonderising of steels and the anodising and chromodising of aluminium and aluminium alloys.

The Wire Pistol Method of Metal Spraying. B. G. Delves. (Engineering Inspection, 1944, vol. 9, Summer Issue, pp. 35–38, 46). A description is given of the metal-spraying process, with details of the latest types of equipment employed.

Production Metallizing. H. E. Linsley. (Steel, 1944, vol. 114, June 19, pp. 92, 132–134). An account is given of the development by the Wright Aeronautical Corporation of automatic equipment for spraying steel aero-engine cylinders with aluminium at the rate of one every 75 sec.

PROPERTIES AND TESTS

The Elastic Modulus and Damping Capacity of Irreversible Iron-Nickel Alloys. E. Scheil and G. Reinacher. (Zeitschrift für Metallkunde, 1944, vol. 36, Mar., pp. 63–69). An investigation of the $\gamma \rightarrow \alpha$ transformation in some irreversible iron-nickel alloys containing 10–29% of nickel was carried out by making simultaneous measurements of the elastic modulus, which is dependent on the microstructure, and of the damping capacity, which is not dependent on the microstructure. With pure iron, with alloys containing a small percentage of nickel and with those containing 15–29% of nickel the $\gamma \rightarrow \alpha$ transformation is associated with a reduction in the elastic modulus; with alloys containing 10–15% of nickel there is an increase in the elastic modulus. Raising the nickel content lowers the elastic moduli of both α-iron and γ-iron. The $\gamma \rightarrow \alpha$ transformation is accompanied by an increase in the damping capacity and this corresponds to a similar increase caused by plastic deformation.

The Significance of Tensile and Other Mechanical Test Properties of Metals. H. O'Neill. (Proceedings of the Institution of Mechanical Engineers, 1944, vol. 151, pp. 116–130). Critical consideration of the conventional quantities obtained from the tensile test reveals their limitations for design purposes, especially where notch-fatigue effects predominate. Against such notch-fatigue the importance of high work-hardening capacity in a metal is emphasised and its relation to tensile-elongation behaviour outlined. It is suggested that even conventional tensile records may yield some rough measure of work-hardening capacity prior to cracking by means of quantities which the author calls "plasticity ratio" and "plasticity value." The deformation and strain-hardening which take place up to the point of "critical plasticity" (*i.e.*, "necking" in the tensile test) appear to differ in internal crystallographic mechanism from that developing beyond this point. It is found that several physical properties reach limiting values at the "necking stress." Wear remains a complex problem, but examples are given showing that relatively good wear resistance has corresponded with relatively high "plasticity values."

Selecting Steels on the Basis of Carbon Content. A. S. Jameson. (Steel, 1944, vol. 115, Aug. 14, pp. 94–100). Tensile-test data on steels and the results of microscopic examination are presented in a manner which shows their dependency on the carbon content. The carbon content is the most important factor in the selection of a steel for a particular application. The effect of alloying elements on the hardenability and tensile properties of several National Emergency steels is also discussed.

A Precision Extensometer and Its Use for the Measurement of Poisson's Ratio. W. Lethersick. (Journal of Scientific Instruments, 1944, vol. 21, Oct., pp. 180–183). A simple optical arrangement is described which is suitable for measuring Poisson's ratio under various conditions of stress

and strain. This improved apparatus requires only one image, and is consequently simpler to use, and is adapted to rapid testing.

Variables Affecting the Results of Notched-Bar Impact Tests on Steels. C. E. Jackson, M. A. Pugacz and F. S. McKenna. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1668: Metals Technology, 1944, vol. 11, Aug.). The stress system imposed upon a notched bar undergoing an impact test is so complicated and unpredictable that no simple evaluation of the test has been possible. As a result of a study of this type of test, the authors are able to contribute new data. A comparison of the results of tests on keyhole and V-notch specimens is included and the effect of the testing temperature considered.

On the Nature of Plastic Deformation. H. Hencky. (Iron and Steel Institute, 1944, Translation Series, No. 196). An English translation is presented of a paper which appeared in Zeitschrift des Vereines Deutscher Ingenieure, 1925, vol. 69, May 16, pp. 695–696 and Sept. 26, pp. 1253–1254. By resolving the state of stress into two independent parts, namely, into a state of homogeneous pressure and tension, as obtains in fluids, and into a state which does not cause any change in volume, the author derives a system of equations which is valid for the state of elastic-plastic equilibrium in the neighbourhood of the yield point. These equations enable the deformation to be calculated in a simple manner.

Crystallite-Deformation at the Surface of Parts containing Residual Stresses Subjected to Static Tensile and Compressive Loads. A. Schaal. (Zeitschrift für Metallkunde, 1944, vol. 36, Mar., pp. 70–72). Specimens of a chromium-molybdenum steel containing residual stresses as a result of machining and heat treatment were examined by an X-ray technique and the stress determined at increasing depth from the surface. A residual compressive stress of 55 kg. per sq. mm. was determined in the surface layer to a depth of 0·15 mm. On superimposing a tensile stress, the stress in the surface layer was found to be much lower than that in the deeper layers; this resulted in the metal beginning to flow below the surface. When a compressive stress is superimposed, the residual and applied stresses are additive with the result that the highest loading is at the surface where the metal soon begins to flow.

The Metallurgical Notch as a Factor in Fatigue Failure. W. H. Bruckner. (Metal Progress, 1944, vol. 45, June, pp. 1102–1103). The term "metallurgical notch" is defined as any narrow region within a piece of metal where considerable changes in hardness occur. Three cases of failure due to this defect are described and microstructures in the notch regions during stages of re-hardening are illustrated.

The Fatigue Strength of Nickel-Chromium-Cobalt and Iron-Chromium-Cobalt Alloys. A. Krisch. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1944, vol. 27, No. 1, pp. 1–12). An investigation of the fatigue strengths of a nickel alloy and high-cobalt steels at temperatures in the 600–800° C. range is reported. The nickel alloy contained nickel 40%, chromium 12%, cobalt 21%, carbon 0·08% and 3·5–4% each of molybdenum, tungsten and titanium. The four steels tested contained cobalt 15–24%, chromium 13%, molybdenum 5–6% and carbon 0·1%. The results were evaluated by the method defined in the German Standard A 117/18 and are presented in tables and by curves. Micrographs of the structures before and after testing are reproduced.

Increasing the Fatigue Strength of Gear Wheels by Special Shaping, Hardening and Machining of the Tooth Base. M. Ulrich. (Iron and Steel Institute, 1944, Translation Series, No. 198). An English translation is presented of a paper which appeared in Luftwissen, 1942, vol. 9, Nov., pp. 311–312. Equipment for determining the fatigue strength of gear wheels in aero-engines is described and particulars are given of the measures taken

to increase the strength of the teeth of already finished wheels. These measures included enlargement of the chamfer at the shoulder of the tooth, alteration to the thickness of the rim and removal of the grooves in the tooth gullet caused by grinding and polishing.

Hardenability of Cast Steel—A Symposium. (Metals and Alloys, 1944, vol. 19, June, pp. 1415–1424). After a brief introduction by C. W. Briggs the following papers are reproduced :

> **Hardenability of Cast Steel and Wrought.** J. B. Caine. The manner in which the hardenability of cast and wrought steel is affected by the composition and grain size is discussed. Fine-grained cast steels are usually higher in silicon and aluminium than fine-grained wrought steels because, for castings, the deoxidation procedure must be carried much further. With steels of the same initial analysis the hardenability of cast steel is therefore slightly higher than that of wrought steel.
>
> **Use of the End-Quench Test for Cast Steel.** E. J. Wellauer. A table is presented giving the cooling rates in the end-quench test from the core to the surface of steel bars 1, 2, 3 and 4 in. in dia. and it is shown how to use the data in conjunction with curves to predict the hardness variation across the section of a bar at different distances from the end.
>
> **Hardenability of Some Cast Steels.** F. Kiper. A variation of the end-quench hardenability test, developed by the Ohio Steel Foundry Company, is described and the results of this test on specimens of four manganese-molybdenum cast steels are given.

Hardenability of 4150 Steel. H. L. Walker, E. J. Eckel, J. Hino and F. H. Mueller. (Metals and Alloys, 1944, vol. 19, Feb., pp. 346–350). Some tests on the hardenability of steel S.A.E. 4150 are recorded and the development of the correct heat treatment to produce a hardness of Rockwell C 54–55 at the centre of $2\frac{1}{4}$-in. dia. bars is described.

Dynamic Hardness Testing at Elevated Temperatures. E. Fetz. (Iron Age, 1943, vol. 152, Dec. 23, pp. 42–53 ; Dec. 30, pp. 40–52). The literature on the dynamic hardness testing of steel and non-ferrous metals at elevated temperatures is critically reviewed and the results of many series of hardness tests are presented in tables and by graphs.

End-Quench Test Permits Rapid Evaluation of Steels. C. M. Parker. (Machinist, 1944, vol. 88, Oct. 7, pp. 91–94). The fact that the properties of steels of the same chemical composition may differ considerably makes individual testing a necessity. The preparation of test specimens from each heat is a laborious process ; an assessment can be rapidly made, however, by means of the standard end-quench hardenability test. In this paper the relation between tensile properties, hardness as determined by the end-quench test and microstructure is considered.

Magnetic Method of Controlling the Heat-Treatment of Tractor Parts. M. N. Mikheev. (Vestnik Metallopromyshlennosti, 1940, No. 8–9, pp. 87–89). (In Russian). A hardness-testing instrument based on the measurement of the coercive force or the demagnetising current is described. The instrument is calibrated using specimens of known hardness and with a known microstructure. This test takes very much less time than the Brinell test and, unlike the latter, is unaffected by surface roughness, scale, or superficial decarburisation.

Ball Wear in Cylindrical Mills. E. W. Davis. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1736 ; Mining Technology, 1944, vol. 8, July). A critical review of a paper by T. K. Prentice on the wear of steel balls used for grinding ores is presented. (*See* Journ. I. and S.I., 1943, No. II., p. 55 A).

The Friction of Cast Iron Against Chromium-Plated Steel. V. I. Arkharov. (Vestnik Metallopromyshlennosti, 1940, No. 10, pp. 10–12). (In Russian). The coefficient of friction was determined for cast iron against quenched and unquenched steel unplated and with three types of chromium plating, namely, mat deposits plated at 20° C., bright deposits at 50° C. and milky deposits at 80° C. The coefficients of friction against quenched and unquenched steel were approximately the same. Against plated steel the smallest friction was obtained with the bright deposits formed at 50° C.

Crack Detection. (Machinery, 1944, vol. 65, Sept. 7, pp. 253–256). A description is given of the Reynolds electro-magnetic equipment for the inspection of welded tubular structures. This equipment is for applying current to tubular structures for magnetic powder tests to detect cracks. Special clamps are provided for conveying the current to the tubes for longitudinal tests, and a single-loop induction ring is supplied for making transverse tests.

Equipment Designed by VIAM (the Soviet Union Institute for Aircraft Materials) for the Control of Parts by the Magnetic Powder Method. A. A. Kiselev. (Vestnik Metallopromyshlennosti, 1940, No. 11–12, pp. 78–87). (In Russian). A range of magnetic-powder testing machines has been designed with a view to their production for widespread use in industry. The various types are illustrated and described with reference to their fields of application, and the electrical circuits used.

A Testing Stand for Magnetic Powder. R. Berthold. (Zeitschrift des Vereines Deutscher Ingenieure, 1943, vol. 87, June 26, pp. 399–401). In magnetic powder testing the results obtained depend greatly on the properties of the powder or suspension used. In order to establish uniformity and to provide data for specifying these powders the Reichs-Röntgenstelle has developed an apparatus for determining their properties. A detailed description of the apparatus and some results achieved with it are given.

A Note on the Physical Properties of an Austenitic Weld-Metal and Its Structural Transformation on Straining. K. Winterton. (Iron and Steel Institute, 1945, this Journal, Section I.). Mechanical tests at elevated temperatures on composite 18/8 weld-plate tensile specimens showed that the tensile strength, yield strength and hardness declined with increasing testing temperatures. The elongation was at a maximum when testing at 250° C. The effect of prior heat treatment at 850° C. in causing increased tensile strength and decreased yield strength decreased with testing temperature and was not apparent above 150° C. Its effect in reducing elongation persisted for testing temperatures up to 500° C. Microscopic examination showed a breakdown of dendritic regions to a light-etching α-constituent, and the formation of lines and blocks of a deep-etching α-constituent, probably due to uneven straining. After treatment at 850° C. the deep-etching α-constituent was found at the grain boundaries, particularly after testing in the range of 200–400° C.; this reduced the elongation values.

Tensile Properties of Medium-Carbon Low-Alloy Cast Steels. H. A. Schwartz and W. K. Bock. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1719: Metals Technology, 1944, vol. 11, Aug.). It is shown that when the tensile strength of a given steel in various states of heat treatment is plotted against its elongation or Brinell hardness number a straight line is obtained, the equation for which can be computed with considerable accuracy from the chemical composition of the steel. The logarithm of the slope of the line is shown to be proportional to the cumulative effect of the total addition of alloying elements.

The American National Emergency Steels. R. F. Mather. (Metallurgia, 1944, vol. 30, Aug., pp. 197–201). A comprehensive and detailed

account is given of the development of the American National Emergency steels. The revised list of compositions of National Emergency Alloy Steels issued by the American Iron and Steel Institute in May, 1944, is reproduced.

Index of Uses for National Emergency Steels. (Metal Progress, 1944, vol. 46, July, pp. 91–94). A catalogue compiled by the Bethlehem Steel Company is presented in which the National Emergency steels are given in numerical order each one being followed by a list of its uses.

How United States Designers Shift to Use of N.E. (National Emergency) Steels. O. L. Mesenhimer. (Steel, 1944, vol. 115, Aug. 7, pp. 98–101, 160). The chemical and physical properties of a number of war-time substitute steels used for highly-stressed parts such as diesel engine studs are tabulated and discussed.

Development and Application of Military and Special Steels for Ordnance Purposes. J. H. Frye. (American Iron and Steel Institute: Blast Furnace and Steel Plant, 1944, vol. 32, June, pp. 695–697). The effects of the changing war situation upon the American steel industry, particularly in connection with specialised production, is described.

Development of Special Steels for Naval Use. E. G. Touceda. (American Iron and Steel Institute: Blast Furnace and Steel Plant, 1944, vol. 32, July, pp. 807–810). The procedure followed by the Research and Standards Branch of the Bureau of Ships of the United States Navy Department in establishing criteria for the acceptability of steel for certain specifications is explained. The type of information on steel plate still being sought by the Bureau is discussed, and the research methods proposed to achieve this end are outlined.

Materials for the Valves of Internal Combustion Engines. H. Cornelius. (Stahl und Eisen, 1944, vol. 64, July 6, pp. 433–438; July 13, pp. 453–458). The requirements and properties of the heat-resisting alloy steels for the valves, valve seats and valve guides of internal combustion engines are discussed with special reference to their resistance to fuels containing lead compounds.

The Influence of High-Frequency Surface-Hardening on the Mechanical Properties of Structural Steels. E. V. Shleyer and I. A. Oding. (Vestnik Metallopromyshlennosti, 1940, No. 7, pp. 7–17). (In Russian). Test-pieces of plain carbon steel, and of low-chromium and chromium-nickel steels, all containing 0·36–0·37% of carbon, were tested in the normalised condition, after quenching and after high-frequency surface hardening with and without subsequent heat-treatment at 400° and 550° C. Whilst raising the tensile strength and the elastic limit, surface-hardening reduced the ductility (as measured by the elongation, reduction of area and notched-bar impact strength, all of which could be restored by subsequent heat treatment). Irrespective of the effect of the surface-hardening on the yield point, the fatigue strength of the plain carbon steel was either raised or lowered depending on the conditions of treatment. In the case of the alloy steels, hardening by any form of heat treatment reduced the fatigue strength as compared with that of the normalised material.

The Magnitude and Nature of Residual Tangential Stresses Resulting from Different Heat Treatments. E. S. Tovpenets. (Vestnik Metallopromyshlennosti, 1940, No. 10, pp. 41–44). (In Russian). Experiments were made on chromium-nickel-molybdenum steel tubes which were subjected to quenching, stepped quenching and different tempering treatments. The thermal and structural stresses were calculated. Internal stresses up to 40 kg. per sq. mm. were observed in some cases after rapid quenching in water.

Metallurgical Laboratory Has Many New Features. (Steel, 1944, vol. 115, Aug. 14, pp. 116–118, 174). **Jessop Steel Company Opens New**

Research and Metallurgical Laboratory. (Steel Processing, 1944, vol. 30, July, pp. 440–443 452, 460). An illustrated description is given of the laboratories recently completed at the works of the Jessop Steel Company. Both routine testing and research work are carried out and considerable space is allotted to heat-treatment research.

Purchase of Steel on Performance Rather than Analysis. G. C. Riegel. (Metal Progress, 1944, vol. 45, June, pp. 1090–1092). The problem of correlating laboratory test results with the performance in service of machine parts is discussed, emphasis being laid on dependability as the manufacturer's chief criterion.

Some Aspects of Commercial Production of Alloy Steels to Hardenability Requirements. W. G. Bischoff. (American Iron and Steel Institute: Blast Furnace and Steel Plant, 1944, vol. 32, July, pp. 803–806). One of the chief difficulties of the producer of constructional alloy steels is that of meeting the consumer's demand for a guarantee of hardenability. Such a guarantee generally requires the performance of numerous tests of all kinds. The Technical Committee on Alloy Steel of the American Iron and Steel Institute has therefore devised a testing procedure which covers such elaborate specifications in a simple and rapid manner. This procedure is described.

A System of Abbreviations for Metallic Materials. E. Siebel and F. Fischer. (Stahl und Eisen, 1944, vol. 64, May 18, pp. 317–321). Proposals are made for a system of abbreviations for metals and alloys. The system is based on a short syllable derived from the chemical formula of the element combined with a number indicating the percentage present and letters and figures indicating the method of production, state (*e.g.*, type of heat treatment) and tensile strength.

METALLOGRAPHY AND CONSTITUTION

Radiography of Non-Uniform Sections. R. Taylor. (Metals and Alloys, 1944, vol. 19, June, pp. 1393–1395). A technique is described for obtaining radiographs of castings having varying thicknesses of metal. Two films with a lead screen between are mounted in the same casette; the films have different speeds, the faster one being selected to suit the thicker part of the casting and the slower one for the thinner part. Thus only a single exposure is necessary.

X-Ray Unit Can Penetrate Steel Plate 8 in. Thick. J. L. Bach. (Blast Furnace and Steel Plant, 1944, vol. 32, June, pp. 689–690). A new million-volt X-ray inspection unit developed by the General Electric X-ray Corporation of America is described. The X-rays produced by this machine are capable of penetrating 8 in. of steel plate and by its use the time normally spent in examining components has been considerably reduced.

The X-Ray Determination of the Stress Distribution in Tensile Tests on Notched Flat Specimens. H. Neerfeld. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1944, vol. 27, No. 2, pp. 13–27). The stress distribution in notched flat specimens of a chromium-molybdenum steel was determined from the lattice constants obtained by numerous X-ray back-reflection pictures, the stress being calculated from the lattice constant by Neerfeld's formula. The differences in the stresses arising from V, U and trapezoid notches was observed. With V and U notches the fall in the value of the longitudinal stress from the notch surface towards the core was very steep; calculated stress values for flat bars of infinite width with a hyperbolic notch decreased much more gradually

and these cannot therefore be applied to notches found in practice. With trapezoid notches the maximum value for the longitudinal stress is not necessarily at the base of the notch. Stress determinations made by micro-elongation measurements on bars with similar notches were in good agreement with those obtained by the X-ray method described.

The Effect of Overheating on the Transformation Characteristics of a Nickel-Chromium-Molybdenum Steel. K. Winterton. (Iron and Steel Institute, 1945, this Journal, Section I.). The S-curve for a nickel-chromium-molybdenum steel has been determined by a dilatometric method, using an initial treatment of 10 min. at 850° C. By the same means the S-curve was redetermined after 10 min. treatment at 1200° C. (before treatment at 850° C.). The isothermal characteristics were also found for a prior treatment of only 1 min. at 1200° C. With treatment at 850° C., transformation was sluggish and incomplete. Prior treatment at 1200° C. accelerated the intermediate transformation and retarded transformation to pearlite. Prior treatment at 1200° C. had no effect on the transformation to martensite by continuous cooling.

Tensile Properties of Unstable Austenite and its Low-Temperature Decomposition Products. A. H. Cottrell. (Iron and Steel Institute, 1945, this Journal, Section I.). The mechanical properties and transformation behaviour of an air-quenched nickel-chromium-molybdenum steel were studied at various stages before, during and after the austenite-martensite change. Part of the S-curve was determined as a preliminary to the investigation. Tensile tests on specimens in the unstable austenitic condition show that (*a*) the metal has a low elastic limit and is ductile, (*b*) transformation is induced by plastic strain, the extent of the effect depending upon the sluggishness of the $\gamma \rightarrow \alpha$ change, as indicated by the S-curve, and (*c*) the products of transformation (bainite and martensite) are more elastic and less ductile than the austenite from which they are derived. The tensile properties of martensite obtained by air-quenching were examined during cooling and show that the material possesses high strength and appreciable ductility. Minimum ductility in the metal is obtained immediately after the completion of the change to martensite. These results are discussed briefly in relation to the problem of cracking encountered in this type of steel after welding.

Some Observations on the Austempering and Isothermal Transformation of Steels, with Special Reference to the Production of Martensite. F. C. Thompson and L. R. Stanton. (Iron and Steel Institute, 1945, this Journal, Section I.). Employing the sensitive recording dilatometer already described (Journ. I. and S.I., 1943, No. I., p. 95 P) in conjunction with hardness measurements and microscopic examination, the transformation of austenite at temperatures in the martensitic zone was investigated. A theoretical explanation based on the rate of recrystallisation is offered to account for the shape of the typical S-curve representing the isothermal transformation of austenite; from the results of experimental work it is suggested that the retention of austenite and the subsequent "period of induction" depend on the relaxation of stresses set up in the material at the time of quenching. The experimental work covered the effect of the mass of the specimen, the nature of the steel, its pretreatment and the effect of initial quenching temperature and of variation in that temperature, together with the implications arising from the latter two points with relation to grain size. The variation in behaviour on isothermal transformation of apparently identical specimens was considered, and it is suggested that this may be due to some variation in the material not taken into account in ordinary chemical analysis. Various possibilities accountable for the shape of the austenite transformation curve are reviewed, and, whilst the nucleation theory is not ruled out, it does not seem possible that the

transformation proceeds according to any one specific physico-chemical reaction.

The Activities of Components in Austenite. J. Gerasimoff. (Acta Physicochimica U.R.S.S., 1943, vol. 18, No. 4, pp. 275–299). A method of calculating the activities and free energies of a heterogeneous system composed of a gas mixture and unsaturated solutions is developed and applied to the determination of the activities of carbon and of carbide in austenite at 800–1100° C., using the equilibrium constants for the system austenite plus CO_2 and CO. Austenite may be considered as a cubic modification of carbide in γ-iron, the free energy of its formation from cementite being small and positive. The new modification of carbide is less stable than cementite.

Study of the Processes of Transformation in High-Speed Steel under Pressure. V. I. Prosvirin. (Vestnik Metallopromyshlennosti, 1940, No. 7, pp. 55–61). (In Russian). The transformation in a high-speed steel containing carbon 0·7%, chromium 4·2%, tungsten 17·8% and vanadium 1·2% was investigated. The investigation included quenching and tempering tests under pressure. In the quenching tests the specimens were placed in a die preheated to temperatures of from 20° to 500° C., and were allowed to cool in the die under a pressure of 160 kg. per sq. mm. exerted on all sides. The tempering tests were made generally on oil-quenched specimens, pressure from all sides or unilateral pressure being applied during the whole or part of the time. Treated specimens were examined for hardness and residual austenite, the investigations being supplemented in some cases by the recording of dilatometer curves. In general, the stressed austenite obtained by quenching under pressure resulted in a higher martensite transformation temperature, development of more martensite on tempering, a smaller reduction in hardness as the tempering temperature was raised and a higher secondary hardness than in the case of the oil-quenched steel. Tempering under pressure had substantially the same effects. Cutting tests with tool tips have demonstrated the marked improvement resulting from quenching under pressure.

CORROSION OF IRON AND STEEL

The Principles of Electrolytic Corrosion of Metals. G. E. Coates. (Chemistry and Industry, 1944, Aug. 26, pp. 306–309). Only a very small amount of corrosion need take place before a metal acquires a considerable negative potential relative to the solution ; this negative potential favours the return of the positive ions and in the absence of other factors it would grow until a state of equilibrium is attained when the corrosion would cease. Thus the study of wet corrosion is the study of the manner in which a metal may lose this negative charge. The following types of mechanism influencing the rate of metallic corrosion are discussed : (1) The discharge of positive ions in solution ; (2) the reaction with oxygen or oxidising agents forming negative ions, and (3) metallic conduction to another electrode at which mechanism (1) or (2) may occur. Cases of practical importance involving these processes are described.

Corrosion of Steel by High Temperature Steam. F. Alton. (Engineering and Boiler House Review, 1944, vol. 58, Aug., pp. 202–209). A report is presented on long-time corrosion tests on specimens of 46 different alloys exposed in a pipe line to the attack of steam at high pressure and at temperatures in the 700–1775° F. range. The alloys varied in composition from relatively pure nickel, through nickel-copper alloys, stainless steels,

medium- and low-alloy steels to plain carbon steels and alloy cast irons. A uniform and adherent scale formed on all the steels at 925° F., except the stainless steel which only pitted slightly. The adherence of the scale was greater at 925° than at 1100° F. The addition of 0·50% of molybdenum to a plain carbon steel greatly improved its corrosion resistance in steam at 1100° F., but 2% added to a 12%-chromium steel lowered its corrosion resistance. Analysis of the scale indicated that the resistance of the alloy steels was due to a thin and very dense tightly-adhering inner layer of scale and a dense middle layer containing a higher concentration of the alloying elements than that present in the steel base.

New Design of Humidity Cabinet for Corrosion Testing. F. Todd. (Industrial and Engineering Chemistry, Analytical Edition, 1944, vol. 16, June, pp. 394–397). A cabinet for accelerated corrosion tests on small metal panels is described. With this apparatus very uniform conditions of temperature and humidity can be maintained.

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MINERAL RESOURCES

A Survey of the Iron Ore Resources of Central Sweden. P. Geijer and N. H. Magnusson. (Jernkontorets Annaler, 1944, vol. 128, No. 9, pp. 379–403). (In Swedish). A geological survey of the iron ore resources of central Sweden has recently been completed. A full report is now being printed as a separate publication, and the present paper constitutes a summary of the report by those who carried out the survey. The total ore reserves are now estimated at 220 million tons, as compared with Tegengren's estimate in 1909 of 183 million tons.

Geology of the Manganiferous Iron-Ore Deposits at Boston Hill, New Mexico. L. P. Entwistle. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1712: Mining Technology, 1944, vol. 8, May).

Exploration on the Stillwater Chromite Deposits, Stillwater and Sweetgrass Counties, Montana. P. T. Allsman and E. W. Newman. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1751 : Mining Technology, 1944, vol. 8, Sept.). Trenching, sampling and core-drilling in Stillwater and Sweetgrass Counties, Montana, by the Bureau of Mines have established that there are over five million tons of chromite ore containing more than 20% of chromic oxide. An account of this exploratory work is given.

Vanadium in Iron Ores and Its Behaviour in Ore Preparation. W. Luyken and H. Kirchberg. (Mitteilungen aus dem Kaiser-Wilhelm-Institut für Eisenforschung, 1944, vol. 27, No. 3, pp. 29–42). The geochemistry of vanadium and the vanadium content of iron ores in different parts of the world are discussed. The titanium-bearing magnetites are richest in vanadium, the content being about 0·25%. The amount of vanadium in German iron ores varies considerably, the maximum being 0·13%. In concentrating the titanium-bearing magnetites, the vanadium remains with the iron and most of the titanium goes on to the waste heap.

ORES—MINING AND TREATMENT

Metallurgical Methods of Increasing the Iron Recovery in the Concentration of Lean Ores. M. Tigerschiöld. (Jernkontorets Annaler, 1944, vol. 128, No. 9, pp. 423–450). (In Swedish). Three methods of concentrating lean ores are critically examined and their application to Swedish ores is discussed. These are : (1) Roasting hematites in a reducing atmosphere with subsequent magnetic concentration ; (2) the Krupp-Renn process ; and (3) the production of sponge iron from lean ore with subsequent concentration. An account is given of a large magnetising-roasting plant which has been working in Manchuria on an ore containing about 38·5% of iron to produce a concentrate containing 58–60% of iron. Small-scale trials in Sweden have shown that there are good prospects of producing a concentrate from Kärrgruve hematite with an iron recovery of over 92% by magnetising-roasting. Successful tests in the concentration of sponge iron made from Bålsjöberg and Örling ores are also reported.

Double Pick-Up Features New Magnetic Separator. (Engineering and Mining Journal, 1944, vol. 145, July, p. 85). A brief description is given of the new Memco-Hoster wet-type magnetic separator for iron ore. The crude ore fed to the machine contains 35% of iron ; the tailings contain 1% of iron and the concentrates 66%.

The Fontana Steel Plant and Its Raw Materials Supply. G. D. Ramsay. (American Institute of Mining and Metallurgical Engineers: Mining and Metallurgy, 1944, vol. 25, Sept., pp. 423–426). A general description of the integrated steelworks completed in 1943 at Fontana, California, is given, together with an account in greater detail of the ores and coal and the methods of preparing them for use.

Thaws Frozen Iron Ore in Special Furnace Building. (Steel, 1944, vol. 115, July 31, p. 88). A description is given of a large thawing chamber for thawing out frozen truck-loads of iron ore. Oil-fired combustion units are provided, and the products of combustion are circulated round the trucks by fans. The chamber can take two 90-ton truck loads at a time, and the heating treatment lasts only 20 min.

REFRACTORY MATERIALS

(Continued from p. 1 A)

Carbon Refractories for Blast Furnaces. J. M. Robitschek. (Iron Age, 1944, vol. 153, Feb. 3, pp. 48–53). The advantages of various carbon-brick linings over the conventional firebrick are discussed from the point of view of iron-making in the United States. The properties of American and European fireclay bricks for blast-furnaces are compared. Several indirect advantages resulting from the use of carbon blocks, such as the elimination of inside cooling plates and the much slower accumulation of furnace bear, are pointed out, and the American iron industry is advised to pay greater attention to these refractories.

Uses of Permeable Refractories for Furnace Construction. R. H. Anderson, D. C. Gunn and A. L. Roberts. (Journal of the Institute of Fuel, 1944, vol. 17, Aug., pp. 169–182). The theory is advanced that the property of high permeability in refractories can be used to achieve a marked increase in furnace efficiency. Experimental furnaces were built for use with a clean fuel (town gas) in which the products of combustion did not leave the chamber directly by a flue, but passed first through a wall of permeable refractory material into an annulus. The effect of the outward passage of gases is to tend to decrease the temperature gradient, and therefore the conduction loss; as this is achieved by waste heat, the fuel consumption of the furnace will be reduced if the same temperature conditions are maintained. The experiments confirmed this, and tests were made with heat-treatment furnaces on an industrial scale, the fuel consumption and temperatures being carefully controlled and recorded. The results, which are presented in tables and graphs, indicate that at temperatures below 500° C. (at the hot face) the fuel savings involved are 10% or less, but that the 500–1300° C. range offers the best field for immediate development, for here savings of the order of 40% are likely to be achieved.

Alumino-Silicate Refractories. J. H. Chesters. (Iron Age, 1944, vol. 153, June 8, pp. 48–53, June 15, pp. 79–85). The chemical and mineral compositions, physical structure and properties of refractory clays and of firebricks containing alumina in the 20–70% range are discussed, with notes on the distribution of the raw materials in Great Britain and the United States. In the second part of the paper the manufacture, properties and steelworks applications of high-alumina bricks are described.

Insulation of Steel-Plant Equipment. J. H. Chesters. (Iron Age, 1944, vol. 154, Aug. 24, pp. 42–49; Aug. 31, pp. 47–52, 90). Factors affecting the selection of a refractory insulating material for different furnaces are considered, and two forms of apparatus for determining the

thermal conductivity of materials at low and at high temperatures are described. Data on the properties of diatomite, asbestos, vermiculite, basic materials, and of fireclay and silica insulating bricks are presented. In the second part of the paper heat-transmission calculations are dealt with, and Pluck's slide graph for determining interface temperatures and heat losses through furnace walls is illustrated and explained. Finally, the factors affecting the properties of anti-piping compounds for steel ingots are considered.

FUEL

(Continued from pp. 2 A–3 A)

The Recent Development of Air Turbines with Closed Cycle. H. Lent. (Stahl und Eisen, 1944, vol. 64, Mar. 9, pp. 153–158). The Swiss engineers J. Ackerett and O. Keller of Maschinenfabrik Escher Wyss propose to use air as the working medium in a closed-cycle turbine-compressor plant. The new principle aims at dispensing with the high working temperatures met in the open-cycle combustion turbine and to use much simpler machinery than that of the modern steam turbine plant. Ackerett and Keller have developed a " double isothermal cycle for air " which is independent of the atmospheric pressure and temperature and is only dependent on the available temperature gradient. The output can be varied by altering the pressure level while maintaining the same temperature gradient. Results obtained from an experimental plant confirm that the total efficiency remains substantially constant for great variations in load. The indifference of the cycle to the pressure level has the great advantage that with higher pressures the size of the plant becomes smaller. Compared with a steam turbine plant, the new prime mover dispenses with the whole complicated feed-water system with all its auxiliaries, but the problem of heating the air to 700–800° C. presents difficulties which it is proposed to solve by using a steel recuperator. It is believed that for plants producing more than 10,000 kW. a total thermal efficiency of 41·6% could be achieved, which, with one intermediate superheating stage, might be brought up to 46·2%.

Regenerator Calculations. A. Schack. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Nov.–Dec., pp. 101–118). Formulæ for calculating the heat transfer in regenerators and the temperature changes in the air and gas flowing through them are presented and discussed.

Thermal and Economic Considerations Concerning a Blast-Furnace-Gas-Fired Mixer with Recuperator. M. Steffes. (Stahl und Eisen, 1944, vol. 64, Aug. 3, pp. 506–510). A description is given of the piping layout and recuperators used for heating a 1200-ton metal mixer using blast-furnace gas and air, both preheated in metal recuperators. Heat balances and heat flow diagrams are presented and the economic aspects of the installation are discussed.

Fires Forging and Heat-Treatment Furnaces with Pulverized Coal. R. A. Campbell and J. H. Loux. (Steel, 1944, vol. 115, Oct. 2, pp. 90–94). An illustrated description is given of the pulverised coal-firing system installed for large forging and heat-treatment furnaces at a works in Nova Scotia. Unit pulverisers with individual circulating systems were put in instead of a large central plant. The unit system supplies pulverised coal to a number of burners, and returns any excess coal to the pulveriser; with this system individual burners can be adjusted without affecting the feed to the remaining burners.

The Economic Utilisation of Coke-Oven Gas. E. W. Smith. (Journal of the Institute of Fuel, 1944, vol. 17, Aug., pp. 188–189). The difficulties

of co-ordination between the coke-oven industry and town-gas undertakings are pointed out, with special reference to the way in which the fluctuating seasonal demands of gasworks can be met. The possibility of using surplus coke-oven gas for steam-raising when the boiler plant is equipped with means for alternative firing is discussed, and the practical implications of the Fischer-Tropsch process of synthetic oil production from the point of view of coke-oven-gas utilisation is considered.

Inflating Blast-Furnace-Gas Mains with Coke-Oven Gas. L. Tofft. (Chicago District Blast Furnace and Coke Association : Blast Furnace and Steel Plant, 1944, vol. 32, Aug., pp. 936–937). A method of preventing air from getting into blast-furnace-gas mains is described. On a multi-furnace plant where there are extensive mains it is undesirable to use steam to keep the pipes at a positive pressure during a shut-down. At the works in question coke-oven gas was used, and details are given of the pipe connections and control valves which were fitted for this purpose.

PRODUCTION OF IRON

(Continued from pp. 3 A–5 A)

Determination of the Profiles of Blast Furnaces. W. Goldsbrough and S. G. Throssell. (British Steelmaker, 1944, vol. 10, Sept., pp. 388–402; Oct., pp. 443–453). Equations are given and calculations explained for determining blast-furnace profiles. Two examples of calculations of this type are given: (1) For a furnace producing hematite from brown iron ore and (2) for one producing foundry iron from roasted Cleveland ironstone. The theoretical part of the paper is based on earlier work by M. Pavloff and B. Osann.

The Aerodynamic Turbine in the Steel Industry. C. Keller and R. Ruegg. (Stahl und Eisen, 1944, vol. 64, Mar. 30, pp. 201–206). The recent introduction of steel recuperators for heating blast-furnace blast in a continuous cycle instead of the intermittent cycle of the brick stove has achieved a considerable reduction in power consumption and led to smaller heat losses. A promising possibility of further economic improvement is the application of the aerodynamic turbine as the prime mover for blast-furnace blowers. This turbine requires an air heater of similar construction to the modern steel recuperator (*see* Lent's paper, p. 41 A). The objection that gas turbines require several bulky heat exchangers to achieve a high efficiency is not valid for the air turbine, as the dimensions can be reduced considerably because of the clean working medium and the higher pressure levels employed. The air heater has to-day the same dimensions as a modern steam boiler of the same output. Several schemes are outlined: (1) Where the blower is driven by an air turbine, and a combined air heater heats the blast for the furnace and the cycle air which works the turbine; in this the path of the combustion gases is divided into two. The air coming from the blower is heated in the low-temperature section and the cycle air in the high-temperature section. The high pressure employed in the air-turbine plant improves the heat transmission in the high-temperature zone of the air heater. Experience gained with a pilot plant shows that the problem of heating air to high temperatures has been solved. (2) A combination of Cowper stoves and an air-turbine-driven blower. (3) A plant in which the blower compresses the blast to a higher pressure than that required for the furnace; in this the compressed air is heated in the first stage of an air heater and then expanded to about furnace pressure in an air turbine coupled to a blower; the required additional power is produced by an aerodynamic turbine. The expanded air

is then heated to the temperature required for the furnace. The advantage of this system is that the heat consumption for an equivalent volume of air is about 6% lower than with scheme (1). It would be advantageous to arrange the plants for these three schemes in such a way that the air turbine is divided into high- and low-pressure stages, the low-pressure stage driving the blower at a rate varying with the required load, and the high-pressure stage driving the compressor for the cycle air. Electric power could also be generated by the air turbine if it were made large enough. Details of a combined blast and cylce-air heater capable of heating 100,000 kg. of blast and 109,000 kg. of cycle air per hr. are given. The heating surface for the cycle air is about one-tenth of that for the blast. The cycle-air heater lowers the temperature of the combustion gas to the safe level necessary for the blast-heater tubes. The heating of the cycle air by the combustion gases is carried out in uniflow, which is possible because the temperature gradients between the combustion gases and the air are large and uniflow ensures relatively low and evenly distributed tube temperatures. The blast-furnace blast is heated by cross-current flow. An additional burner is fitted between the two stages to provide a means of controlling the blast temperature independently of the cycle-air temperature. When heating the blast in Cowper stoves, thorough cleaning of the blast-furnace gas is necessary, whereas with the new process such careful cleaning is not necessary.

Postwar Importance of Beneficiated Iron-Bearing Materials. C. E. Agnew. (Steel, 1944, vol. 115, Sept. 4, pp. 98–100; Sept. 11, pp. 112, 174–177). The effects of using an all-sinter burden on fuel economy at the blast-furnace and on the over-all economy of steel production are considered. With an all-sinter burden the productive capacity of the furnace is governed by the capacity of the furnace hearth to smelt the material delivered to it from the shaft; as the sinter is easily prepared in the shaft (compared with soft ore), maximum production is promoted with a heavy burden, a light blowing rate and slower stock travel. The governing factor in sinter-burden practice is the thermal problem of supplying the amount of heat necessary to do the work of the hearth. The lighter blowing rate and absence of volatile matter in the sinter result in a smaller volume of gas, and assist in conserving heat within the furnace. The proportion of sinter in the burden may have to be limited to the ability of the blowing plant to supply air at an increased rate. An all-sinter burden requires less fuel, so that less gas is produced, but at the same time more gas is required for the increased stove capacity necessary to maintain the higher blast temperatures. It is claimed that the direct use of furnace gas in gas turbines for the blowing and electrical generating equipment would offset the reduction in gas volume, because the thermal efficiency of the gas turbine is greater than that of the steam turbine.

The Thermal Dissociation of Carbon Monoxide in Contact with Iron and Its Alloys and in Mixtures with Carbon. F. Körber, H. Wiemer and W. A. Fischer. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Sept.–Oct., pp. 43–52). An investigation is described in which carbon monoxide was passed over iron powder and mixtures of iron with carbides and free carbon in a small electric furnace at steadily increasing temperatures, the object being to determine the effect of these solids on the dissociation of the gas. A temperature of 900° C. was reached before pure iron powder markedly increased the rate of dissociation; it is probable that this was due to chemical reactions between the iron and the gas, not to the iron acting as a catalyst. Additions of up to 1·5% of graphite to the iron had no effect. When cementite was present, however, there was a sharp peak in the carbon-dioxide production curve between 500° and 600° C., with only 0·4% of carbon present as cementite. With increasing amounts of

cementite in the powder at 500–600° C. this peak increased in height, whilst at 900° C. the peak was reduced in height, until at 1·2–1·5% of carbon it had disappeared. With steel powder, either loose or pressed, the dissociation was much less. With falling temperatures the effect of the powders was much less because the effective surface area was reduced by the sintering action. Magnetic determinations showed that when carbon monoxide was passed over iron powder at up to 500° C. the latter was entirely converted to cementite. The rate of this reaction depended on the fineness of the powder. No formation of higher carbides or of any oxide was observed at this temperature.

Midland Iron Works Centenary. (Iron and Coal Trades Review, 1944, vol. 149, Oct. 27, pp. 628–630). A brief account is given of the history of wrought-iron manufacture at the Midland Ironworks, Rotherham, which started production in 1844. An illustrated description of the present-day puddling furnaces and rolling mills is also given.

Powder Metallurgy. F. V. Lenel. (Mechanical Engineering, 1943, vol. 65, July, pp. 489–492). The physical properties of parts made from iron-powder briquettes by sintering are considered in relation to the conditions of manufacture and the qualities of the powders used. A note on the application of iron-powder products to small-arms parts is added.

FOUNDRY PRACTICE

(Continued from pp. 5 A–9 A)

First Report on the Basic Cupola by the Melting Furnaces Sub-Committee. (Institute of British Foundrymen : Foundry Trade Journal, 1944, vol. 74, Sept. 14, pp. 25–28 ; Sept. 21, pp. 55–59 ; Sept. 28, pp. 71–74). The Melting Furnaces Sub-Committee of the Technical Committee of the Institure of British Foundrymen has been investigating the use of stabilised dolomite clinker and other basic materials for lining cupolas, and now publishes its First Report. When working a cupola so that desulphurisation is obtained, *i.e.*, with a minimum of oxidation and producing iron containing 2·80% of total carbon and 1·50% of silicon, the life of the lining in the combustion and melting zones was 12–16 weeks ; the lining in the well withstood 26 weeks' operation, during which about 4000 tons of metal were melted. Studies on dephosphorisation of iron in basic-lined cupolas for steelmaking by the cupola/converter process indicate that : (1) Steel scrap higher in phosphorus can be used ; (2) higher-phosphorus pig-iron can be used instead of hematite in the cupola charge ; (3) steel can be produced lower in phosphorus than that normally obtained from the acid-lined converter ; and (4) the process has the advantage of off-setting the increase in the phosphorus which occurs in the steel produced because of the metallic losses in the cupola and converter. No abnormal pick-up of carbon was noted when operating basic-lined cupolas under conditions favourable for desulphurisation.

Cupola Practice. D. J. Reese. (American Foundrymen's Association : Foundry, 1944, vol. 72, Aug., pp. 82, 196–200 ; Sept., pp. 79, 202–206). The advantages of the cupola as a melting unit and cupola practice in general are discussed. A coke charge of about 75 lb. per square foot of hearth area is recommended. Calculations relating to the coke used for melting and that supplying carbon to the iron are made, and the importance of obtaining reliable data on the weights of the materials charged is emphasised.

Cupola Raw Materials. H. W. Gillett. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Sept., pp. 1–15). The position

of the foundry industry is reviewed with respect to the raw materials for the cupola. The variations to be expected in raw materials and the problem of obtaining uniform quality of products are discussed.

Briquetting " Swarf " at Warner & Swasey. G. W. Birdsall. (Steel, 1944, vol. 115, Sept. 18, pp. 112–114, 178–180). The system developed at a large engineering works at Cleveland, Ohio, for collecting, crushing and briquetting swarf is described and illustrated. The material is compressed into briquettes 3 in. in dia. and $1\frac{1}{2}$ in. thick, weighing $1\frac{1}{2}$ lb. in the case of crushed steel turnings and 2 lb. when made of cast-iron borings. Two men operate the crushing and briquetting machinery, and they produce about 1000 briquettes per hr.

Factors Involved in Superheating Gray Cast Iron and Their Effect on Its Structure and Properties. A. W. Schneble, jun., and J. Chipman. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Sept., pp. 113–158). The parts played by nucleation and by gases in contributing to the effects of superheating cast iron on its properties and structure were studied by casting bars in a 10-lb. high-frequency furnace constructed so that the bars could be cast in the same atmosphere as that in which the iron was melted and superheated. The melts were prepared *in vacuo*, air, carbon monoxide, hydrogen and nitrogen. The superheating effects are not the results of nucleation processes, but of the presence of gases, carbon monoxide being the most important. Although melting in hydrogen produced the same type of tensile-strength curve as carbon-monoxide, the two gases have opposite effects on the structure of cast iron. Generally, hydrogen is not present in sufficient quantity in most melting operations to account entirely for all the superheating effects. Very little is known about the method by which the above gases influence the structure of cast iron. Carbon monoxide may act as a catalyst in the decomposition of iron carbide, thus accelerating graphitisation. Hydrogen may remove carbon monoxide from the melt to form methane and water vapour, thus hindering graphitisation. Nitrogen has little effect on the properties and structure of the metal. Melting *in vacuo* reduces the tensile strength considerably and causes a pronounced change in the microstructure.

Malleable Cast Iron. J. A. Wylde. (Metallurgia, 1944, vol. 30, June, pp. 86–89; July, pp. 123–126, 129; Aug., pp. 191–195; Sept., pp. 245–248). A brief historical outline of the development of whiteheart and blackheart malleable iron castings is given, and the production and metallurgical characteristics of the two types are described, with notes on their uses in shipbuilding and in electrical, automobile, agricultural, railway and general engineering.

Testing Moulding Sand and Moulding Sand Economy. E. Diepschlag. (Giesserei, 1944, vol. 31, Mar., pp. 37–43). Some limitations in the German standard specification for testing moulding sands are pointed out, and formulæ are presented for calculating the density and moisture content from data determined experimentally. The changes in the sand temperature at increasing distances from the mould-metal interface are discussed.

Gas Developed in Molds. N. J. Dunbeck. (Foundry, 1944, vol. 72, Sept., pp. 95–96, 178). Data on the quantities of gas evolved from a variety of moulding sands and core sands at different temperatures are presented and discussed.

Centrifugal Casting of Aircraft Engine Cylinder Liners. (Machinery, 1944, vol. 65, Nov. 9, pp. 505–511). An illustrated description is given of the mass production plant set up by the Ford Motor Company for making aero-engine cylinder liners of centrifugally cast steel. There are twenty-three mould-spinning machines arranged on a continuous conveyor, which travels round an oval track. The steel is poured at a minimum temperature of 2940–2960° F. into the mould, which is spun at about 800 r.p.m.

The rate of production is expected to be 400 liners per hr. with fifteen men working on the moulding line.

Conditioning of Steel Castings to Standards of Quality. P. Ffield. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Sept., pp. 173–204). The improvement of steel castings by the investigation of internal defects by means of radiographic and magnetic testing and by flame-gouging is discussed. Types of defects, radiographic standards of acceptability and interpretation of results are dealt with briefly. The Bethlehem Steel Company's specifications for the magnetic-powder inspection of flanges on castings which have been radiographed are given in an appendix.

PRODUCTION OF STEEL

(Continued from pp. 9 A–11 A)

Granite City Steel Company Increases Production Facilities. C. Longenecker and R. Vaill. (Blast Furnace and Steel Plant, 1944, vol. 32, Aug., pp. 919–926). An illustrated description is given of the extensions to the plant of the Granite City Steel Company, where three 175-ton open-hearth furnaces have been erected, together with two soaking pits, a reheating furnace and additions to the rolling mill.

Soviet Iron and Steel Industry in Wartime. V. Rikman. (Iron and Coal Trades Review, 1944, vol. 149, Nov. 3, pp. 665–667). A brief account is given of the development of the iron and steel industry in the U.S.S.R. and the movements which were necessary owing to enemy occupation of some of the industrial regions. In electric-furnace steel production before the war the Soviet Union was second only to the United States. Under war-time conditions the making of alloy steel in basic open-hearth furnaces was developed. The preparation of ferro-chromium in ordinary blast-furnaces has also been developed in the Urals.

Some Economic Factors in Metallurgy. C. B. Snodgrass. (Metal Treatment, 1944, vol. 11, Autumn Issue, pp. 139–144). The complex nature of the problems facing the metallurgical industry in the post-war period are considered in some detail. The two main factors under discussion are the great increase to be expected in primary producers, with the consequent decrease in world markets, and the fact that the metallurgical industry is based upon the phenomenon of "wasting assets." The solution may be approached through technological efficiency and a study of the distribution of labour and raw materials. The necessity of working to a world economy is emphasised.

The General Fundamentals of Water Supply, Especially at Iron and Steel Works. F. Schreier. (Stahl und Eisen, 1944, vol. 64, Sept. 7, pp. 584–588). The fundamentals of an iron and steel works' water supply are considered, the points discussed including sources of supply, quality requirements, degrees of contamination, mechanical and biological purification, impurities in waste water from a steelworks, circulation systems, increasing the use of cleaned water and co-operation with local authorities and Water Boards.

Steel Plant Maintenance. A. G. Henry. (Iron and Steel Engineer, 1944, vol. 21, Sept., pp. 43–45). Differences in the nature of steel-plant-maintenance problems in peace-time and in war-time are pointed out and several examples of war-time maintenance are described.

Maintenance of Steel Plant Instruments and Controls. W. J. McGregor. (Iron and Steel Engineer, 1944, vol. 21, Sept., pp. 54–55). The organisation of a department for maintaining instruments and controls

at a steelworks is briefly described and the importance of regular inspection and cleaning is emphasised.

Some Problems of Electrical Maintenance. H. L. Coe. (Iron and Steel Engineer, 1944, vol. 21, Sept., pp. 56–58). The organisation of maintenance work for electrical plant and equipment is discussed and the following methods of reducing maintenance costs are amplified: (1) More emphasis on planned preventive maintenance; (2) better and more rapid training of maintenance personnel; (3) greater degree of standardisation in the equipment and materials used in maintenance; and (4) more attention to new and improved methods.

Stopper Heads X-Rayed to Eliminate Pouring Trouble. C. B. Jenni. (Foundry, 1944, vol. 72, Sept., pp. 68–69, 208). A system of X-ray inspection for stoppers for ladles of molten steel is described. In practice this has been found to justify the time and expense, because it prevented those with internal defects from being used, with a consequent reduction in the number of leaky stoppers.

Automatic Manipulation of Open-Hearth Furnace Doors. C. C. Downie. (British Steelmaker, 1944, vol. 10, Oct., pp. 440–442). The disadvantages of hydraulic and electrically operated doors on open-hearth furnaces are pointed out, and a brief description is given of equipment comprising a small winch, rope-drum, electric block and limit switch, all accommodated in a closed housing mounted upon rails above the door. Remote control can be fitted, and a timing clock registers the period during which the door is open.

A Survey of Slag-Control Methods. W. O. Philbrook and A. H. Jolly, jun. (American Institute of Mining and Metallurgical Engineers: Blast Furnace and Steel Plant, 1944, vol. 32, July, pp. 793–797; Aug., pp. 938–942). Methods for estimating the composition of slag for open-hearth control purposes were investigated. The slag-pancake method is fairly reliable with lime/silica ratios in the 1·0 to 2·5 range, is less reliable when this ratio is 2·5 to 3·0, and may be subject to serious errors for ratios exceeding 3·0. The examination of polished slag surfaces by reflected light with a metallurgical microscope is useful for following the maturing of the slag in the course of a heat, but by itself it is not a sensitive means of estimating the slag composition. An estimate of both basicity and the iron and manganese oxide contents of the slag can be made from its colour in the finely powdered state, for the colour varies with the composition; with further study this might be useful for control purposes, but it is probably no better than the examination of pancakes. A new method was discovered, but it has not yet been tried in full-scale operations; it consists of measuring the slight alkalinity of a suspension of the powdered slag in water by means of a glass-electrode *p*H meter. It is thought that this method is equal in reliability to the pancake method with the lime/silica ratio in the 1·5 to 2·5 range. Procedures similar to those used for determining free lime in cement clinker are useless for open-hearth slag control.

Basic Electric Melting Procedure for High-Quality Alloy Steels. A. L. Ascik. (American Society for Metals, Oct., 1944, Preprint No. 26). The reasons for lack of agreement among steelmakers as to a standard melting procedure for making alloy steels in basic electric furnaces are discussed, and that favoured by the author is described. The difference between the physical properties in the longitudinal and transverse directions found in forged ingots is taken as a measure of the heterogeneity, and the way in which the manganese/silicon ratio in the charge, the nature and composition of the slags and other factors in the melting practice affect the heterogeneity are pointed out. Deoxidation with silicon and manganese compounds is preferred, as aluminium reacts with the residual

oxygen to form solid aluminium oxide which remains permanently in the steel.

Furnace Atmospheres in Arc Furnaces and Their Effect on the Hydrogen and Nitrogen Contents of the Steel. L. Villner and A. Norrö. (Iron and Steel Institute, 1944, Translation Series, No. 200). An English translation is presented of a paper which appeared in Jernkontorets Annaler, 1944, vol. 128, No. 3, pp. 105–134 (*see* Journ. I. and S.I., 1944, No. II., p. 106 A).

Surface Defects on Ingots and Billets. G. Helmer. (Jernkontorets Annaler, 1944, vol. 128, No. 9, pp. 457–503). (In Swedish). In 1939 Jernkontoret appointed a committee, consisting of B. D. Enlund, T. Lindblom and the author, to investigate the causes and methods of preventing surface defects in ingots and billets. In this paper the author reviews the committee's investigations, as well as other published and hitherto unpublished work, and presents his own recommendations based on this information. A detailed description is attempted of the mechanism of the formation of cracks in ingots of different shape. The effects of four types of ingot mould coatings (tar, tar oil, mould varnish and a tincture) are studied and the removal of surface defects by chiselling, flame deseaming, grinding and machining are discussed from the technical and economic points of view.

Fundamental Principles Involved in Segregation in Alloy Castings. R. M. Brick. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1764 : Metals Technology, 1944, vol. 11, Sept., pp. 3–10). Fairly rapid solidification of an alloy over a range of temperatures leads to "coring," *i.e.*, a difference in composition between the centre of a dendrite and its periphery ; this can usually be detected only by microscopical examination. The dendritic form of solidification leaves interdendritic channels, which, filled with the low-melting-point constituents, reduce the large differences in composition found in cored structures to relatively low differences detectable only by macro-etching. Examples of these two forms of solidification are discussed.

A Review of Factors Underlying Segregation in Steel Ingots. B. M. Larsen. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1764 : Metals Technology, 1944, vol. 11, Sept., pp. 13–32). The following significant factors affecting segregation in the solidification of steel ingots are discussed : (1) The heat evolved during freezing ; (2) the tendency for skeleton-like crystals to grow faster along the original axis of formation than in other directions ; (3) the higher iron content of the crystal nuclei, which creates an adjacent layer of lower freezing point ; (4) keeping the liquid metal in motion during freezing ; (5) the difference in pressure at the top and bottom of the ingot ; (6) the appreciable difference in the specific gravities between liquid metal and solid crystallites ; and (7) the evolution of gas from the carbon-oxygen reaction in the ingot.

The Relation of Open-Hearth Practice to Segregation in Rimmed Steel. J. W. Halley and G. L. Plimpton, jun. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1764 : Metals Technology, 1944, Sept., pp. 37–55). *See* p. 11 A.

Segregation in a Large Alloy-Steel Ingot. S. W. Poole and J. A. Rosa. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1764 : Metals Technology, 1944, vol. 11, Sept., pp. 58–68). A detailed report is presented of an investigation by sulphur prints and by quantitative chemical analysis of a 15-ton steel ingot. The steel was produced in a 70-ton basic-lined electric furnace, and it was top-poured at a temperature of about 2830° F. The ladle analysis of the steel was : Carbon 0·380%, nickel 1·86%, chromium 0·81%, vanadium

0·07% and molybdenum 0·34%. All the elements segregated to some extent, carbon the most and vanadium the least. Most of the segregation occurred in the lower third of the ingot, this area being low in the alloying elements. Slight enrichment of manganese and sulphur occurred at the centre of the upper third portion.

FORGING, STAMPING AND DRAWING

(Continued from pp. 12 A–14 A)

High Tensile Bolt Production. (Wire Association: Wire and Wire Products, 1943, vol. 18, Oct., pp. 631–634, 643–644, 648–649, 659, 674, 686). The paper takes the form of answers by various members of the Wire Association to an enquiry by a British member concerning the production of high-tensile steel bolts on both double-blow cold-heading machines and "National Boltmakers" machines. A considerable percentage of cracked heads was experienced in the heading process on both types of machine.

Coating Compounds Contribute to Increased Production in Drawing Wire and Tubing. J. H. Richards, jun. (Steel, 1944, vol. 115, Sept. 18, pp. 136–140). The advantages of using solutions of lubricating compounds instead of lime coatings for the drawing of alloy steel wire and tubes are cited.

Drawing High Speed Rope Wire. J. C. Aiken. (Wire and Wire Products, 1943, vol. 18, Oct., pp. 601–603, 668). Methods are described for drawing rope wire at more than the usual speeds by means of the continuous machine equipped with water-cooled blocks. Procedures for meeting rigid tensile and torsional test requirements are discussed.

Electric Equipment for Wire Drawing Machines. A. L. Thurman. (Wire and Wire Products, 1944, vol. 19, July, pp. 420–424, 454; Aug., pp. 488–492). The requirements of motors for single-block and multi-block wire-drawing machines are discussed in the first part; in the second part the use of tensiometers on wire-drawing machines and the application of the Thyratron control to electrical winding reels are described.

The Graphical Solution of Wire Mill Mathematical Problems. E. J. Crum. (Wire and Wire Products, 1943, vol. 18, Oct., pp. 585–598, 666). A number of charts are provided which enable the rapid determination of (*a*) the relationship between the weight per foot and the diameter of steel wire, (*b*) the relationship between the length per pound and the diameter, (*c*) the linear speeds of wires being drawn on a cylindrical block, (*d*) the pounds of wire produced per hour from different diameters at different drawing speeds, and (*e*) the percentage reduction in area in drawing from one diameter to another and other data of value to wire-drawers. All the data are for round steel wires.

Sub-Sieve Diamond Powders for Die Work. P. L. Herz. (Wire and Wire Products, 1944, vol. 19, Aug., pp. 479–484, 516; Sept., pp. 547–549, 572–574). The importance of uniform grain size in diamond powders used for grinding wire-drawing dies is pointed out, the unsatisfactory quality of sieved powders, in this respect, being emphasised. A technique is described in detail for crushing bort (a fine crystalline form of carbon) and three grading methods are considered, namely, air-flow, fluid-flow and decantation, in relation to the uniformity of grain size which each produces.

ROLLING-MILL PRACTICE

(Continued from pp. 14 A–15 A)

Pilot Rolling Mills Aid in Checking Physical Specifications. R. L. Hartford. (Steel, 1944, vol. 115, Aug. 28, pp. 114–116). The use of pilot rolling mills to enable test rollings to be made on a small scale is described. Such mills have enabled determinations to be made of the effects of various rolling operations upon the physical properties of the steel.

Energy Consumption in Hot Rolling. G. Wallquist. (Jernkontorets Annaler, 1944, vol. 128, No. 7, pp. 249–308; No. 8, pp. 309–369.) (In Swedish). In an endeavour to find more widely applicable formulæ for calculating the energy consumed in hot-rolling, an extensive investigation has been carried out in which determinations were made of the effects of the type of steel, the rolling temperature and the type of bearings on the power consumed in rolling different sections. The results are given in tables.

The Manufacture, Testing and Protection of Steel Pipes for Water Mains and Services. J. H. Smith. (Transactions of the Manchester Association of Engineers, 1943–44, pp. 57–73). Brief descriptions are given of the following methods of manufacturing steel pipes: (1) The Fretz-Moon process; (2) various welding processes; and (3) the production of tubes from hollow blooms by dies, by rolling and by the pilger mill. In the second part of the paper the causes of the corrosion of the outside and inside of buried water mains are considered and particulars of the latest types of bitumen protective coatings are given.

Improvements in Rolling-Mills. A. Schiffers. (Stahl und Eisen, 1944, vol. 64, Aug. 31, pp. 561–564). Brief descriptions are given of a number of items of improved equipment for soaking-pits and rolling-mills. These include an ingot tipper, a stamp for numbering billets, a ramp for guiding billets between the two upper rolls of a three-high stand, an automatic shear for bar ends, wire-mill guides, &c.

HEAT TREATMENT

(Continued from pp. 15 A–19 A)

Experiments on Sodium-Cyaniding of High-Speed Steel Prior to Hardening. J. McIntyre. (American Society for Metals, Oct., 1944, Preprint No. 29). Tests are described which show that the hardness and life of high-speed steels can be improved by treatment in a cyanide bath before the normal hardening procedure. The tests were carried out on an 18/4/1 tungsten-chromium-vanadium steel and two molybdenum high-speed steels. The treatment consisting of holding for $1\frac{1}{2}$ hr. in a sodium-cyanide bath at 790° C., cooling in air, cleaning, heating the 18/4/1 steel to 1260° C. (1235° C. for the molybdenum steels), quenching in oil and tempering twice for 1 hr. at 565° C. A hardness of Rockwell C 68–70 could be obtained at the centre of 1-in. dia. bars by this means.

The Nitriding Process: Factors Influencing Ammonia Consumption. C. V. Snell. (Metal Progress, 1944, vol. 46, Aug., pp. 299–304). Methods of reducing ammonia consumption in the nitriding of steels are given. A 60% drop in consumption occurs if the composition of the inflowing gases is so adjusted as to give 60% of dissociated ammonia in the exhaust gases; the dissociation may be allowed to increase to 85% without detriment to the case, provided that a preliminary nitriding at 30% dissocia-

tion has been in force for a short time. By using a high ammonia dissociation the "white layer" may also be eliminated. At lower temperatures (940–950° C.) the ammonia consumption is lower by 20%; the Rockwell hardness is higher and the case depth is reduced. The gas pressure in the nitriding chamber should be kept at 1 in. of water. The effect of furnace construction and maintenance on ammonia consumption is also dealt with.

A Note on the Relationship between Preliminary Heat Treatment and Response to Nitriding of Some Nitriding Steels. C. C. Hodgson and H. O. Waring. (Iron and Steel Institute, 1945, this Journal, Section I.). In the nitriding of some chromium-bearing steels it has been found that the hardness of the case is influenced considerably by the preliminary heat treatment to which the steel has been subjected. This is particularly true of the pre-nitriding tempering treatment. Steels of the chromium-molybdenum and chromium-molybdenum-vanadium types are considerably affected. A chromium-molybdenum steel containing a notable amount of aluminium is scarcely affected at all in this respect. The details of the experimental work are described and illustrated, and possible explanations of the results obtained are discussed.

Nitriding Hardened High-Speed-Steel Tools. J. G. Morrison. (Mechanical Engineering, 1944, vol. 66, Aug., pp. 539–542). The nitriding of high-speed steel in molten cyanide baths is described, and the results of hardness, impact and bend tests on nitrided tools are presented and discussed. The process is carried out in a bath consisting of sodium cyanide 70% and potassium cyanide 30% held at about 1050° F. The impact test is too severe to simulate machining conditions. If tools are already giving poor performance because of excessive impact or misalignment, nitriding may lower rather than improve the cutting efficiency, but the liquid nitriding of hardened high-speed-steel tools increases the tool life considerably where relatively light cuts are taken.

Isothermal Quench Baths Applied to Commercial Practice. H. J. Babcock. (Iron Age, 1944, vol. 153, Feb. 3, pp. 44–47; Feb. 10, pp. 62–69). The theory of Davenport and Bain's S-curves for indicating the beginning and end of the transformation of austenite in quenched steels is explained, and their practical application for the heat treatment of steel using molten salt baths is described.

Application of High Temperature Conveyer Belts.—II. S. C. Alexander. (Industrial Heating, 1944, vol. 11, Aug., pp. 1242–1252). A description is given of wire-woven conveyor belts used for carrying parts through heat-treatment furnaces.

Proposals for the Alteration and Recommendations for the Operation of Sodium-Nitrate Patenting Baths. O. Peltzer. (Stahl und Eisen, 1944, vol. 64, Aug. 10, pp. 513–518). The patenting of wire in sodium-nitrate baths instead of in molten lead has been successfully carried out in full-scale production in Germany. The conversion of a plant from lead to sodium-nitrate operation is discussed in detail, particularly with regard to the changes in the heating equipment for the baths.

Metallurgical Aspects of the Bofors Anti-Aircraft Gun. W. M. Mitchell. (Metal Treating Institute: Industrial Heating, 1944, vol. 11, Aug., pp. 1231–1234). The methods by which the manufacture of the Swedish Bofors gun was adapted to American practices and standards are described with special reference to the heat treatment of the gun parts.

Heat Treating Broaches. W. L. Gibbons. (Steel, 1944, vol. 115, Sept. 11, pp. 106–110, 166). The heat-treatment procedure at a works making high-speed steel broaches up to 7 in. in dia. is described.

Manufacture and Heat Treatment of Hairsprings. J. W. Whittaker. (Metal Treatment, 1944, vol. 11, Autumn Issue, pp. 193–198). A tech-

nique developed by Metropolitan-Vickers Electrical Company, Ltd., for the manufacture and heat treatment of fine instrument hairsprings is described in detail. The springs in question have torque requirements of the order of 0·04 mg. cm. per degree. This necessitates an accuracy to within 3·3% of 0·001 in. which is the spring thickness A rolling technique, heat-treatment in an inert atmosphere and the special oven devised for the simultaneous production of large numbers of the springs of the required standard are described.

Practical Aspects of the Selection of Frequency and Time Cycles for the Processing of Metallic Parts with Induction Heating. W. E. Benninghoff and H. B. Osborn, jun. (American Society for Metals, Oct., 1944, Preprint No. 30). Factors affecting the selection of the correct frequency, current input and time cycle for induction hardening are discussed in a practical manner, with particular reference to 9600-cycle motor-generator equipment, as this has a wider range of applications than other frequencies. A simple guide to the selection of the lowest possible frequency for heating the induction of round bars above the magnetic transformation point, when the magnetic permeability of the material is unity, is the equation $F = 180/R^2$, where F is the frequency in cycles per second and R the bar radius in inches. The interval between the end of the heating period and the beginning of the quench is a most important control factor. Increasing this interval gives additional time for heat penetration and carbide solution, and may be used instead of increasing the power or current time.

Induction Hardening of Plain Carbon Steels : A Study of the Effect of Temperature, Composition and Prior Structures on the Hardness and Structure after Hardening. D. L. Martin and Florence E. Wiley. (American Society for Metals, Oct., 1944, Preprint No. 31). An investigation is reported of the factors affecting the hardness and structure obtained in the induction hardening of bars of fine steels with carbon in the range 0·17–1·10%. Alternating currents with a frequency of 530,000 cycles per sec. were used. The temperature is the most important variable, and this must be high enough for the diffusion of the carbon to be sufficiently rapid. For a 0·45%-carbon steel 850° C. is usually satisfactory. The prior structure is important if the rapid formation of austenite at a low temperature is desired. A sorbitic structure transformed to austenite at a lower temperature than did any of the other structures studied.

Induction Hardening Advantageous in Fabrication of Gun Parts. (Machinist, 1944, vol. 88, Nov. 4, pp. 96–97). Illustrated descriptions are given of some of the induction-hardening processes used by a Toronto firm in making parts for Bren guns.

Timed Quenching. M. L. Frey. (Metal Progress, 1944, vol. 46, Aug., pp. 308–309). The old-established practice of timed quenching is reviewed in the light of recent advances in the knowledge of the mechanism of quenching and tempering operations. In timed quenching the part is withdrawn from the cooling agent after a limited time so that the residual heat is sufficient to temper it automatically.

How to Avoid Heat-Treat Distortion by Die Quenching. G. W. Birdsall. (Steel, 1944, vol. 115, Aug. 28, pp. 84–94). The distortion of steel components by heat treatment can be avoided in gears, shafted parts and symmetrical flat shapes by controlled quenching while still within the die. In this paper are described and illustrated a number of die-quenching machines of the rotating type, contact between the quenching medium and the part being controlled centrifugally. The Gleason die-quenching press is also described. In this machine the part is lowered mechanically into the quenching medium. Mention is also made of controlled quenching in the die of asymmetrical machine-gun parts.

Working with Quenching Fluids. (Machinist, 1944, vol. 88, Oct. 14, pp. 113–124). The theory and practice of quenching are reviewed in detail. The cooling characteristics of brine, water, water-glycerine mixtures, sodium-hydroxide solution, oils and air are discussed with data on: (*a*) Their comparative quenching rates; (*b*) the effect of the temperature of the quenching bath; and (*c*) the dimensions of the part. Finally, different types of quenching equipment are described.

Water Quenching Large Steel Castings. R. A. Gezelius. (Steel Founders' Society of America: Foundry, 1944, vol. 72, Sept., pp. 74–77, 176–178). *See* Journ. I. and S.I., 1944, No. II., p. 153 A.

Hot Quenching and Austempering. H. J. Elmendorf. (Metal Progress, 1944, vol. 46, Aug., pp. 311–314). The mechanism of the hot-quenching and austempering process and its advantages and disadvantages are described in the light of recent knowledge.

Martempering. B. F. Shepherd. (Metal Progress, 1944, vol. 46, Aug., pp. 309–311). A process in which steels are quenched at high temperatures and subsequently cooled in air is described. The process aims at producing a steel containing 100% martensite, and has been called "mar-tempering."

WELDING AND CUTTING

(Continued from pp. 19 A–22 A)

Boron Treated Electrode Coatings Aid Cast Iron Welding. J. A. Neumann. (Iron Age, 1944, vol. 153, Feb. 3, pp. 56–58). Boron introduced into the coating of Monel welding-electrodes appears to behave as a flux, and permits thorough binding between the Monel and the cast iron, thus giving an excellent weld. The mechanism of this process is discussed, and micrographs of treated and untreated welds are included.

New Process Bonds Aluminium Cooling Fins to Steel Cylinder Barrels. (Industrial Heating, 1944, vol. 11, Aug., pp. 1270–1272). A new process recently developed by the Fairchild Engine and Airplane Corporation, New York, enables aluminium cooling fins to be permanently bonded to the steel cylinders of aircraft engines. For military reasons details of the process are not given.

Recent Progress in the Application of Welding to Steel. W. F. Hess. (American Iron and Steel Institute: Steel Processing, 1944, vol. 30, June, pp. 370–371, 392; July, pp. 430–431, 435; Aug., pp. 501–503). Factors affecting the successful application of welding are discussed. Two ways of expressing the energy input in arc welding are explained. In the first, this is given in joules per inch of weld, and it is equal to the arc voltage times the current divided by the welding speed in inches per second. In the other method the weld energy input is specified in terms of the volume of metal deposited per pass; this can be done because the amount of metal deposited is directly related to the joules per inch used.

Low-Temperature Welding of High-Speed Steel. (Machine Shop Magazine, 1944, vol. 5, Oct., pp. 94–98). A process of repairing high-speed steel tools is described in which a proprietary brand of copper-base welding rods is used to make the joint.

Flame Cutting Under Water. G. R. Reiss. (Steel, 1944, vol. 115, Sept. 4, pp. 80–81, 136–138). An account is given of a method of cutting sheet steel with an oxy-acetylene flame in which distortion is eliminated. The plates are actually cut under water, and it is suggested that the water medium carries away more rapidly the heat generated by the flame, thus minimising the hardening of the cut edge which normally

occurs. Straightening operations are avoided by this means, cutting is more accurate, and the margin of excess metal required for the cut is reduced.

MACHINING

(Continued from pp. 23 A–24 A)

Increased Efficiency by Increasing the Cutting Capacity. H. Stevens. (Stahl und Eisen, 1944, vol. 64, Aug. 31, pp. 564–571). A method has been devised for determining the optimum value of the product "feed × cutting speed" for machining with high-speed steel. Data for determining this value in a rapid and simple manner under any given conditions of strength, size of the work and its carbon content, chip dimensions, tool life and lathe capacity are presented in tables and graphs.

Increasing the Efficiency of Machining by Lead Additions to Steel. H. Schrader. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Sept.–Oct., pp. 65–76). The results of extensive machining tests on twelve different carbon and low-alloy steels with and without additions of lead are compared and discussed. The lead addition improved the machinability in all cases, and the tool life was longer, especially when cutting the softer steels. The hardenability of the unalloyed steels was reduced slightly by the lead addition. The special measures necessary in the production of lead-bearing steel are described.

Short-Time Methods for Determining the Machinability of Steels and the Life of Tools when Turning with Fine Cuts. H. Lüpfert. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Sept.–Oct., pp. 89–98). The results of a number of machinability and tool-life tests made by measuring the cutting pressure and temperature and by measuring the bluntness of the tool with the Leyensetter pendulum apparatus are presented and discussed.

Surface Finish and Its Measurement. R. E. Reason. (Journal of the Institution of Production Engineers, 1944, vol. 23, Oct., pp. 347–372). An account is given of the various instruments now in use for the detection of imperfections in metal surfaces. Particular attention is paid to the stylus type of instrument, and the imperfections which may be revealed by its use are classified.

A Case for the Qualitative Inspection of Surface Finish. H. P. Jost. (Machinery, 1944, vol. 65, Nov. 2, pp. 483–486). Surface-finish inspection by qualitative analysis consists of comparing the finish of the part tested with that of a standard part of known roughness. The following ways of doing this, and their sensitivity, are described: (1) The visual comparison method; (2) the photomicrographical method; (3) the comparison micrographical method; and (4) the friction dynamometer method.

PROPERTIES AND TESTS

(Continued from pp. 28 A–33 A)

Dependability of Engineering Property Tests. H. A. Reece. (Steel, 1944, vol. 115, Oct. 9, pp. 130–132, 296–298). The need for careful selection of the correct type of test to apply to iron and steel for particular purposes is emphasised, and the results of tensile and fatigue tests at different temperatures on a high-strength cast iron are given.

The Evaluation of Structural Steels from Test Results. W. Kuntze. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Nov.–Dec., pp. 127–135).

It has often been found that steels have exhibited brittle behaviour in spite of the fact that specimens had a high elongation in the tensile test. Theoretical considerations which account for this phenomenon are discussed, and the proposal is made that a steel should be judged by the value of the ratio of the local to the general extension in the tensile test, and it is shown that the safe range for this ratio is between 10 and 30. The value of the reduction in area in the notched-bar tensile test as a measure of the tendency to embrittlement and the weldability of steel is also dealt with.

The Relationship of " Rheology " to " Strength of Materials." A. C. Vivian. (Metallurgia, 1944, vol. 30, Sept., pp. 237–240). The properties used in studies of the strength of materials are much more concerned with elastic than with plastic behaviour. The present standard " ductility " of a ductile metal, for example, is always grossly inaccurate. Certain properties of matter and materials which are described as " rheological," since they appertain to flow or to shape-change due to force, are now being studied by the British Rheologists' Club. The relationship between rheology and a study of mechanical properties, aspects of which are discussed in this paper, appears to be a very close one.

The Effect of Fiber on Notched Bar Tensile Strength of a Heat-Treated Low-Alloy Steel. G. Sachs, J. D. Lubahn, L. J. Ebert and E. L. Aul. (American Society for Metals, Oct., 1944, Preprint No. 14). The effect of fibre direction on a low-alloy nickel-chromium steel was investigated using plain and notched specimens cut longitudinally and transversely from bars $1\frac{1}{2}$-in. and $2\frac{3}{4}$ in. in dia. The bar diameter and position from which the specimen was cut had little effect on the results obtained with longitudinal plain and notched specimens. In the transverse direction the ductility was considerably less than in the longitudinal direction. Neither bar size nor fibre direction had any effect on the ultimate strength.

The Effects of Notches of Varying Depth on the Strength of Heat-Treated Low-Alloy Steels. G. Sachs, J. D. Lubahn and L. J. Ebert. (American Society for Metals, Oct., 1944, Preprint No. 15). This report is the concluding one of a series on factors affecting the strength of notched steel specimens (*see* Journ. I. and S.I., 1944, No. I., p. 100 A). The combined effects of notch depth and notch radius on the tensile strength of a nickel-chromium steel heat-treated to strength levels between 145,000 and 240,000 lb. per sq. in. were studied. The notched-bar strength increases with increasing notch depth, approximately linearly, only if the strength of the steel is rather low or the notch radius is large. On the other hand, sharply-notched high-strength steels have a minimum strength at intermediate notch depth. There are two different influences at work, the ratio of the transverse to the longitudinal stress (which the authors call " triaxiality ") and the stress concentration. The cohesive strength of a heat-treated steel rises with increasing triaxiality.

A Metallographic Quality Test for Malleable Iron. C. T. Eakin and H. W. Lownie. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Sept., pp. 261–272). The effect of primary graphite, primary cementite and pearlite upon the properties of malleable cast iron to A.S.T.M. Specification *A*-47–33, is described. Charts are provided whereby the tensile properties of the iron may be determined from a knowledge of these three constituents. A practical sampling and laboratory procedure for making the test is given in outline.

A Comparison of Aluminium and Titanium Deoxidation for Preventing Strain-Aging Embrittlement in Low-Carbon Steel. G. F. Comstock and J. R. Lewis. (American Society for Metals, Oct., 1944, Preprint No. 24). The strain-ageing of low-carbon steel, melted in an induction furnace, deoxidised with silicon, aluminium or titanium and forged from small

ingots, was studied by three methods to ascertain how the deoxidising agent, the forging practice and the heat treatment affected the strain-ageing. The three methods were : (1) By impact tests after cold-drawing with and without subsequent ageing; (2) by impact tests after tensile straining, with and without ageing; and (3) by Brinell hardness tests at increasing temperatures up to 500° F. The results by the three methods were in close agreement; the first was preferred as being the most imformative. If sufficient titanium is used it is just as effective as aluminium in preventing strain-ageing embrittlement. The minimum amount required to equal 2 or 3 lb. of aluminium per ton of steel depends on the forging practice and the heat treatment. The amount of ferro-carbon-titanium (containing 20% of titanium) which is usually effective is 15 lb. per ton of steel, although 10 or 12 lb. appear to be effective with the best conditions of forging and heat treatment.

A Study of Work-Hardening and Reannealing of Iron. A. Balicki. (Iron and Steel Institute, 1945, this Journal, Section I.). The paper is an account of an extensive investigation into the changes in various properties of Armco-iron wires resulting from twelve different degrees of work-hardening, and subsequent annealing in a vacuum. Particular care was taken to secure comparable sets of results. On the same samples, or on exactly similar material, the following properties were determined : Hardness, electrical resistivity, elastic limit from the bend-test, plastic limit, ultimate stress, uniform elongation, springiness, thermo-electric properties and microstructure. Attention was paid to the effect of the time of annealing and to the order of the inaccuracy caused by ageing. Some other effects of which account should be taken are also indicated in the paper. A survey of the changes induced by reannealing confirms the presence of three phenomena : Strain-ageing, recrystallisation and crystal growth.

The main change, that of recrystallisation, of a work-hardened wire is shown by all the curves as occurring in the same range of temperatures (which is contrary to the findings of Tammann); this suggests that all the changes are a result of the same fundamental cause. The recrystallisation range is lowered when the degree of work-hardening is increased. From the change in the course of the curves which characterise work-hardened iron three regions of plasticity were distinguished. A short description of the working hypothesis of work-hardening is given and also an account of the theory of recrystallisation developed by Krupkowski and the author. The applicability of the theory to the experimental results is demonstrated by recalculation of some of the experimental curves. The calculated and experimental values show satisfactory agreement, although not as good as was previously obtained when copper and nickel were similarly treated. Evidence is produced that irregularities are caused by strain-ageing. A point of theoretical importance is that the energy of activation of the recrystallisation process of iron is independent of the degree of reduction and is nearly the same (31,000 cal. per mol) as previously determined for copper (33,000 cal. per mol) and nickel (32,400 cal. per mol). The frequency coefficient c is found to depend upon the degree of cold-work in a similar manner to that in which certain properties increase with increasing degrees of reduction. The value of c depicts the nature and the history of the metal.

The Influence of the Heat Treatment of Steel on the Damping Capacity at Low Stresses. L. Frommer and A. Murray. (Iron and Steel Institute, 1945, this Journal, Section I.). An electromagnetic method of inducing torsional oscillations in freely-suspended cylindrical steel bars was used for measurements of the damping capacity up to a maximum stress of 100 lb. per sq. in. There were six specimens, each 3 in. in dia. and 3 ft.

long, all from the same melt, with a composition normal to a 0·6% carbon steel. Measurements were made in the normalised, 830° C.-oil-quenched and the fully-tempered conditions; by successive heat treatment these conditions were repeated. The measurements showed that the damping value was a characteristic property which varied for each condition and was reproduced through two heat-treatment cycles. Within the experimental accuracy the torsional damping capacity is independent of oscillation frequency up to 7000 cycles per sec. and of stresses up to 100 lb. per sq. in. The damping values as expressed by the logarithmic decrement are $0{\cdot}5 \times 10^{-4}$ for the tempered condition, and $0{\cdot}7 \times 10^{-4}$ for the oil-quenched condition. The Brinell hardness values showed changes similar to these, and it is suggested that this correspondence is due to the state of aggregation of the carbide, resulting from the heat treatment given.

Annealing Studies on Cold Rolled Iron and Iron Binary Alloys. C. R. Austin, L. A. Luini and R. W. Lindsay. (American Society for Metals, Oct., 1944, Preprint No. 40). The response of binary iron alloys to strain-hardening after cold-rolling and their behaviour during annealing were investigated. The data presented are for iron alloyed with nickel, chromium, cobalt, silicon, manganese and molybdenum. The changes in the properties were followed by hardness measurements and examination of the microstructure. In most cases the binary alloys behaved similarly to unalloyed iron in their response to cold-rolling. There were significant differences in behaviour during annealing, for alloying with molybdenum, chromium or manganese displaced the softening point of the cold-worked samples to higher temperatures than that at which the unalloyed iron softened.

Factors Controlling Graphitization of Carbon Steels at Subcritical Temperatures. C. R. Austin and M. C. Fetzer. (American Society for Metals, Oct., 1944, Preprint No. 42). An investigation of the effects of mechanical and thermal treatment at above the critical point on the subsequent graphitisation of steels at sub-critical temperatures is reported. The steels used contained 1% of carbon and small amounts of chromium, nickel and copper. Whilst compressive stress had little effect, tensile stress greatly increased the instability of the carbides as revealed by subsequent tempering. The precipitation of graphite particles was studied by macrographs and hardness determinations. Pretempering water-quenched steels at 670° C. before prolonged tempering at 715° C. had a marked effect in promoting graphitisation, whereas pretreatment at 715° C. had no such effect. The presence of oxygen in the medium surrounding the steel specimens accelerated the rate of graphitisation at sub-critical temperatures. This effect was strongest in the steels quenched from the highest temperature (1000° C.), whereas normalised steels, with few exceptions, were not materially affected.

Some Notes on the Discarding of Mine Winding-Ropes. K. Paterson. (Journal of the Institution of Civil Engineers, 1944, vol. 23, Nov., pp. 17–28). Details are given of the signs of failure which led to the discarding of five large winding-ropes, together with the results of tests carried out on the suspected portions of the ropes after they had been taken out of service.

The Mechanism of Failure of 18-Cr 8-Ni Cracking Still Tubes. C. L. Clark and J. W. Freeman. (American Society for Metals, Oct., 1944, Preprint No. 12). An investigation of the cause of the failure by cracking of 18/8 stainless steel tubes used in petroleum distillation for periods up to 97,500 hr. is reported. From the test results and examination of the microstructure it is concluded that failure is due to structural changes at the grain boundaries, consisting of the precipitation and growth of highly alloyed ferrite resulting from the decomposition of austenite. When

these precipitated areas have grown to a certain size, cracks, at first nearly submicroscopic in nature, will occur under certain conditions of time, temperature and stress. When this cracking has occurred the tube is permanently brittle, but at any time prior to the cracking, the tube is either ductile or can have its ductility restored.

Cracking and Embrittlement in Boilers. H. N. Boetcher. (Mechanical Engineering, 1944, vol. 66, Sept., pp. 593–601). Descriptions with micrographs are given of several types of cracks found in boiler steel. These include cracks arising from folds, ageing, corrosion fatigue, creep, intercrystalline oxidation and caustic embrittlement. A new type of cracking has been experienced in recent years in water-wall and generating tubes of a few boilers operating at very high pressure. The lowest pressure at which it is known to have occurred is 1200 lb. per sq. in. The characteristics of this trouble closely resemble those of hydrogen embrittlement. It always occurs under a layer of hard, dense, magnetic iron oxide. Alternative theories which might account for the formation of this oxide are advanced.

Some Causes of Failure of Medium and High Duty Helical Compression Springs. G. H. Jackson. (Journal of the Junior Institution of Engineers, 1944, vol. 55, Oct., pp. 9–19). The subject of failure in high-compression springs is dealt with from the point of view of the quality and condition of the wire of which the spring is made. Attention is confined to steel wires only; the effects of fatigue, superimposed stresses (as a result of surge), corrosion, embrittlement, defects originally present in the billet or bloom, decarburisation and mechanical defects are all considered.

Failures in Cold-Headed Machine Bolts. H. R. Isenburger. (Steel, 1944, vol. 115, Sept. 18, pp, 116–118). The frequent failure of cold-headed bolts by fracture across the top of the shank was investigated by X-rays, micrographs and macrographs. It was found to be due to non-metallic inclusions or "dirty" steel, a very small amount of foreign material in the zone of maximum cold-work being sufficient to cause fracture under quite low stress.

Effect of Screw Threads on Fatigue. S. M. Arnold. (Mechanical Engineering, 1943, vol. 65, July, pp. 497–505). The paper consists of a review of the literature on investigations concerning the fatigue testing of threaded members and screwed joints. A list of 167 references is appended.

A Comparison of Microhardness Indentation Tests. D. R. Tate. (American Society for Metals, Oct., 1944, Preprint No. 1). The Tukon and Eberbach microhardness testing machines are described. Both are satisfactory for applying indenting loads of 1 kg. or less. The Tukon tester can be used with both Vickers and Knoop indenters when a suitable adapter is provided, and the two machines give results in good agreement with loads of between 200 g. and 1 kg.

Further Developments of the End-Quench Hardenability Test. C. R. Wilks, E. Cook and H. S. Avery. (American Society for Metals, Oct., 1944, Preprint No. 6). As the Jominy end-quench test does not yield data readily applicable to the cooling of plates, another test method has been sought, and the simple expedient of quenching the bar simultaneously from both ends has been developed. It was thought that, once the dimensions of the double-end-quench test specimen had been established to approximate the cooling of the section under investigation for a constant analysis, a general equation for the time-temperature relationships between section and test-bar size, that would be independent of composition, would follow. Experimental evidence supporting this hypothesis is presented in the form of an initial report on an investigation of the relation of hardenability, quenching rate and section size to the mechanical properties of cast alloy steels after various heat treatments.

A Hardenability Test for Low Carbon and Shallow-Hardening Steels. O. W. McMullan. (American Society for Metals, Oct., 1944, Preprint No. 7). Some limitations of the Jominy end-quench hardenability test are pointed out, and a new test method is described which is specially applicable for shallow-hardening steels. For this test round or rectangular specimens are prepared with the back face cut at an angle to the face which is to be quenched, so as to form a wedge. By careful selection of the angle between the two faces, full hardness is developed at the thin end, and the increased length of the oblique face across which the hardness readings are taken enables the changes in hardness to be followed more accurately. Some results obtained by this procedure are pointed out.

The Effect of Carbon Content on Hardenability. E. S. Rowland, J. Welchner, R. G. Hill and J. J. Russ. (American Society for Metals, Oct., 1944, Preprint No. 16). End-quench test results on two series of steels, one containing 1·5% of chromium and 0·27% nickel, and the other 0·25% of chromium, 1·75% of nickel and 0·25% of molybdenum are reported. In each series the carbon content ranged from 0·20% to about 1%. The tests were made on annealed and on normalised specimens held at quenching temperatures of up to 925° C. for up to 4 hr. The similarity in the behaviour of the two series was so marked that the effects of prior structure, holding time and hardening temperature can probably be predicted for these steels when only the carbon content is known. In the eutectoid carbon range the change in hardenability is enormous with variations in the conditions, but, in general, the steel obeys the same laws as those found for steels lower in carbon. The normalised prior structure produces higher hardenability than the annealed structure under all reasonable conditions of time and quenching temperature.

Air Hardenability of Steels. C. B. Post, M. C. Fetzer and W. H. Fenstermacher. (American Society for Metals, Oct., 1944, Preprint No. 17). Experimental data are presented to show that the cooling velocity at the centre of round bars is controlled by the ratio of the surface area to the volume and is independent of the hardening temperature. An air-hardenability gradient test is described. This test enables the hardening properties of a given air-hardening steel to be determined when the sizes range from 1 in. in dia. × 8 in. long to 6 in. in dia. × 6 in. long (area/volume ratios from 4·25 to 1). Small changes in the hardening temperature affected the hardness of the steels with over 1% of carbon, but had little effect on the lower carbon steels. In general, the variation in the hardenability of air-hardening steels from heat to heat was found to be negligible for a given type of analysis, so that the data presented could be used for predicting the hardness of similar steels after similar heat treatment.

End-Quench Test Reveals Changes in Steel Induced by Heat. C. M. Parker. (Machinist, 1944, vol. 88, Oct. 21, pp. 100–103). The Jominy end-quench hardenability test is described, and some curves from test data are presented which show the usefulness of the test in demonstrating the effect of chromium and nickel on the hardenability of steel.

Magnetic Powder Inspection of Large Castings. J. F. Cotton. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Sept., pp. 205–231). The problems involved in the magnetic inspection of large castings are reviewed with the object of evaluating quantitatively the variables encountered in such a method of inspection. It is shown that a standardised technique must be established before standards of magnetic-powder inspection can be prepared.

Organic Coating Aids Magnetic Inspection. G. C. Close. (Iron Age, 1944, vol. 154, Aug. 31, p. 41). At an American aircraft works it was the practice to apply a flash coating of cadmium to parts which were to be tested by the magnetic powder method; this provided a bright background

against which the magnetic particles stood out sharply. As cadmium became very scarce a substitute had to be found. Particulars are given of an organic coating which is easily prepared, is cheaper than cadmium and can be applied with a spray gun; it has proved to be very suitable for the above purpose.

Magnetic Sorting Bridge. (Engineering, 1944, vol. 158, Nov. 10, p. 366). A brief description is given of an electronic device known as the Salford magnetic sorting bridge. It consists of two balanced coils and an amplifier with a cathode-ray oscillograph which portrays the hysteresis loop of the part under test. The apparatus has two objects; the first is to make sure that the part tested has the desired properties before any work is done on it, and the second is to ascertain whether the properties have been altered by the process of manufacture. A master part with the desired properties is placed in one coil, and the part to be tested in the second coil. Differences in the properties are easily detected by changes in the shape of the loop.

Permanent Magnets and Their Application. H. E. Hutter. (Mechanical World, 1944, vol. 116, Oct. 6, pp. 369–371). Some alloy steels for permanent magnets and a number of their applications for scientific research, industrial and domestic purposes are described.

Interferometer Measurements on the Expansion of Iron. J. B. Saunders. (Journal of Research of the National Bureau of Standards, 1944, vol. 33, Aug., Research Paper No. 1597). Several sets of data on the expansion of relatively pure iron by different observers and different procedures are compared. The results show good agreement between data taken with those interferometer methods that are free from the tilting of spacers and air-film errors, whereas the failure to eliminate these sources of error produces data which cannot be duplicated except by chance.

The Scaling Behaviour of Hard Alloy Welded Deposits with Different Cobalt Contents. H. Cornelius. (Stahl und Eisen, 1944, vol. 64, Aug. 17, pp. 529–532). An investigation of the scaling resistance of eight alloys containing different amounts of cobalt in air and in the products of combustion from leaded and non-leaded petrol at 900–1100° C. is reported. In air, welded deposits containing 1–36% of cobalt behaved just the same as alloys containing 60–65% of cobalt. Alloys with 30–36% of cobalt having equal scaling resistance to the exhaust gases from both leaded and non-leaded petrol can be prepared, the cobalt being replaced by iron and the other elements remaining approximately the same.

Properties of Metals and Alloys at Sub-Zero Temperatures. J. W. Donaldson. (Metal Treatment, 1944, vol. 11, Autumn Issue, pp. 161–170). Investigations carried out during the past five years on the mechanical properties of metals at sub-zero temperatures are briefly reviewed. Results of tests on 18/8 chromium-nickel steel, cast steel, wrought steel, weld metal, austenitic steels and some nickel steels are given.

Notes on Recent Trends and Uses of Alloy Steels. B. B. Morton. (Mechanical Engineering, 1944, vol. 66, Aug., pp. 543–546). The properties of low-alloy steels and trends in their use are discussed. There has already been a pronounced increase in the use of low-alloy steels before 1938, and the effects of the present war have been to widen their use and to extend them into the field of heat-treated steels. The National Emergency steels are too new for any but the most tentative predictions to be made; there are, however, indications that some of them will have a permanent place in the post-war years.

Rates of Tempering in Cobalt Steels. E. A. Loria. (American Society for Metals, Oct., 1944, Preprint No. 4). The effect of cobalt on the tempering characteristics of steels containing about 0·7% of carbon and up to

11·2% of cobalt was studied by means of hardness measurements. The cobalt promoted a small increase in the hardness after tempering. Since the cobalt in all probability is not present in any large proportion in the carbide, its effect is to strengthen the ferrite in which the gradually coarsening carbides are dispersed.

The Effect of Tungsten, Molybdenum and Vanadium in Chromium-Bearing Steels for Hot-Working. R. Hohage, W. Völker and R. von Tinti. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Sept.–Oct., pp. 57–64). The effect of additions of 1–12% of tungsten, 0·5–5% of molybdenum and 0·2–1·0% of vanadium on the properties of a 1·5%-chromium steel were investigated. The effects of the hardening and tempering temperatures on the hardness and microstructure were studied. Determinations were made of the elastic limit at 400° C., the Brinell hardness at 650° C., the sensitivity to cracking at 600–650° C. and the thermal conductivity at 400° and 500° C. A special method for determining the crack sensitivity at high temperatures is described. Tungsten gave the steel a high resistance to hot-cracking; vanadium increased the strength at high temperatures. The thermal conductivity was reduced by increasing additions of alloying elements.

The Suitability of Various Heat-Treatable Steels for Nitriding. H. Cornelius and W. Trossen. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Sept.–Oct., pp. 77–88). Specimens of twenty-two low-alloy steels containing aluminium, chromium, molybdenum, niobium, boron and thorium, either separately or in combination, were tested after long-time tempering at 450°, 500° and 550° C., both before and after nitriding in ammonia for periods of 150, 100, 50 and 25 hr. Tests of the hardness and depth of the nitrided case led to the conclusion that to attain a Vickers hardness of 650 a steel containing carbon 0·25–0·35%, silicon 0·75–0·9%, manganese 1·7–1·8%, chromium 0·75–0·90% and vanadium about 0·25% was the most suitable.

Chromium Steels of Low Carbon Content. R. Franks. (American Society for Metals, Oct., 1944, Preprint No. 33). An investigation of the properties of low-carbon steels containing up to 25% of chromium is reported. Wrought 3%-chromium steel with up to about 0·10% of carbon is at least five times as resistant to corrosion in industrial atmospheres as ordinary low carbon steel. The 3%-chromium steel has good resistance to oxidation at temperatures up to 650° C. It does not harden on cooling in air from 900° C. Annealing at sub-critical temperatures (725° C.) softens this steel and imparts high ductility and toughness; good toughness is maintained down to −125° C., provided the steel is first annealed at 725° C., but if it has been normalised at 900° C., the toughness at sub-zero temperatures is not so great. The low-carbon 3%-chromium steel has good fabricating properties, and when welded the toughness is satisfactory at temperatures down to −40° C.

Characteristics and Properties of Some Cast Chromium-Molybdenum Steels. N. A. Ziegler and W. L. Meinhart. (American Society for Metals, Oct., 1944, Preprint No. 34). Thermal studies, micro-examination, and mechanical and welding tests were carried out on steels containing carbon 0·05–0·30%, chromium 2·5–9% and up to 1·5% of molybdenum. The steels remained relatively soft after slow cooling (at 3·2° C. per min.) from the austenitic state regardless of the chromium and molybdenum contents, provided the carbon did not exceed 0·07%. Steels with 2·5–9% of chromium, 0·13–0·20% of carbon and 0·5% max. of molybdenum did not develop suppressed transformations, and remained relatively soft after slow cooling from the austenitic state. Rapid cooling, however, made them hard. Increasing the molybdenum to the 0·5–1·5% range caused slightly greater hardness on slow cooling, and pronounced hardness and

suppressed transformations on rapid cooling. When the carbon content exceeded 0·2%, the steels had marked suppressed transformations, and became hard after either rapid or slow cooling from the austenitic state.

Segregation of Molybdenum in Phosphorus-Bearing Alloyed Gray Cast Iron. F. B. Rote and W. P. Wood. (American Society for Metals, Oct., 1944, Preprint No. 35). Experiments were made to determine the behaviour of molybdenum in grey cast iron containing increasing amounts of phosphorus up to 1·83%. Cylinders 3 in. in dia. × 4·5 in. long were cast with a thermocouple in the centre. Molybdenum formed a complex quaternary eutectic with phosphorus, iron and carbon in the iron studied. For every unit of phosphorus above the amount soluble in iron (*i.e.*, above 0·12%) 1·3 units of molybdenum were removed from the matrix. The composition of the eutectic which formed as small pellets on the surface of the cast cylinder was carbon 2·10%, silicon 1·10%, nickel 1·65%, phosphorus 4·40%, molybdenum 5·85%, remainder iron. The addition of small quantities of molybdenum to an ordinary grey iron in which no attempt is made to keep the phosphorus content low is useless because practically all the molybdenum segregates to the phosphide eutectic.

The Effect of Cold-Rolling on the Structure of Hadfield Manganese Steel. N. P. Goss. (American Society for Metals, Oct., 1944, Preprint No. 41). The structure of strain-hardened 14%-manganese steel was compared by X-ray diffraction technique with that of the alloy in the strain-free state. The diffraction diagrams for the manganese steel strip after being reduced 87% by cold-rolling showed that it was still austenitic; α-iron diffraction lines could not be found on the patterns. The evidence obtained supported Niconoff's theory that the extreme hardness obtained is due to the smallness of the crystallites. The maximum hardness attainable by cold-rolling is of the order of Rockwell C 58.

NE Steels in Oil Well Drilling Equipment. B. B. Wescott. (Metal Progress, 1944, vol. 46, Aug., pp. 291–294). The tensile properties and hardness of N.E. steels as used for oil-well drilling equipment in the United States are compared with three of pre-war alloy steels. The war-time steels have proved satisfactory substitutes for the S.A.E. steels formerly used for tool joints, drill collars and rack bits. No substitute has been found, however, which has the anti-corrosion properties of the 1·5%-nickel steels used for sucker-rods immersed in oxygen-free fluid of high sulphide content.

Jessop Steel Promotes Metallurgical Research by Erecting a New Laboratory. (Blast Furnace and Steel Plant, 1944, vol. 32, Aug., pp. 943–948). *See* p. 32 A.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 33 A–35 A)

British Metallurgical Institutes and Research Associations. J. W. Donaldson. (Metallurgia, 1944, vol. 30, Sept., pp. 251–254). A general *résumé* of the work of the institutes and research associations directly connected with metallurgical research work is presented, with notes on some of the work which has been carried out.

Reflected Light in Metallurgy. E. Barber. (British Steelmaker, 1944, vol. 10, July, pp. 302–306; Aug., pp. 335–339; Sept., pp. 404–408; Oct., pp. 436–439). In Part I. of this series the methods of illuminating specimens for examination with metallurgical microscopes are explained; these methods include dark and light field illumination, plane glass reflectors,

prism reflectors, oblique light and the polarising illuminator. In Part II. the optical principles of the microscope are explained, with details of the construction of objectives and eye-pieces. The method of ascertaining the combined magnifications of objective and eye-piece is given. In Part III. a formula for calculating the resolving power of a microscope and the relationship between the numerical aperture and the resolving power are explained. In Part IV. the use of filters and the mounting of specimens are dealt with.

Fractography—A New Tool for Metallurgical Research. C. A. Zapffe and M. Clogg, jun. (American Society for Metals, Oct., 1944, Preprint No. 36). Fractography is defined as the micrographic study of cleavage facets on fractured specimens. The technique and advantages of fractography are described and discussed. It is regarded as a supplementary tool for metallurgical research, and it offers many of the advantages of single-crystal methods. The plane of weakness, which is often the plane of greatest interest, becomes *ipso facto* the field of observation, and the internal structure of the grain, unaffected by polishing and etching, can be studied. Examples of numerous metallic specimens and of rocksalt are reproduced.

Cleavage Structures of Iron-Silicon Alloys. C. A. Zapffe and M. Clogg, jun. (American Society for Metals, Oct., 1944, Preprint No. 37). The application of fractography (*see* preceding abstract) to the study of the iron-silicon system is described. At 1·59% of silicon all the patterns were of the mosaic type found in pure iron and were consistent with a single α-phase. At 4·24% of silicon, although micrographs showed only the same α-phase, fractographs revealed some subtle structural change already occurring within that phase. At 11·84% of silicon micrographs still showed only one phase, but fractographs revealed patterns which were quite distinct from those of the above lower-silicon alloys; these are taken as evidence of super-lattice formation as the composition Fe_3Si is approached. At 18·10% of silicon two phases occurred, and the cleavage pattern of one, which is ascribed to FeSi, was distinctive because of its complete lack of crystallographic markings. The second one was α-phase, or a modification of it, within which unique structures appeared; these structures are ascribed to precipitation of the compound Fe_5Si_3. Many of the cleavage patterns resist explanation on the classical grounds of deformation and cleavage, and point clearly to the pre-existence of an intracrystalline structure which must exert a tremendous influence upon the entire nature of the material.

Grain Shape and Grain Growth. D. Harker and E. R. Parker. (American Society for Metals, Oct., 1944, Preprint No. 38). Arguments are presented to prove that the ability of a metal to show grain growth depends not on the grain size, but only on the grain shape. When all grains in a metal have flat faces making angles of 120° with adjacent faces there can be no grain growth, no matter what the grain size may be. The distribution of the angles at which grain boundaries meet on the surface of a metallographic specimen is a criterion for the growth stability of its grain structure. This distribution is calculated for the ideal stable structure in which only 120° angles exist between adjacent faces. It is shown experimentally that the distribution of grain-boundary junction angles approaches the calculated one as a metal is annealed, irrespective of the final grain size.

Rapid Hand Polishing of Micro-Specimens. A. L. Schaeffler. (Metal Progress, 1944, vol. 46, Aug., pp. 285–287). A polishing technique for metallographic specimens of hard alloy steels is described. The specimens are first rough-polished with emery, then wax-lapped, and finished with Gamal, a proprietary material of which γ-alumina is the principal constituent.

The Application of X-Ray Crystal Analysis Methods to Engineering Problems. F. Brech. (Transactions of the Manchester Association of Engineers, 1943–44, pp. 179–196). The theory underlying the application of X-rays to produce diffraction patterns to yield information on the chemical composition, the texture and the residual stresses in a metal or alloy are explained. An example is described of the application of diffraction patterns to distinguish between those steel bars which will distort on machining down for manufacture into bolts and those which will not.

Improved Sensitivity in Double Exposure Radiography. J. Rigbey. (American Society for Metals, Oct., 1944, Preprint No. 2). In order to overcome the decreased sensitivity inherent in double-exposure radiographs, a return is made to the stereoscopic method of exposing two separate films. An X-ray technique for determining the depth of flaws in castings using the image of a lead marker as a reference point is described in detail.

Factors Influencing Exposure Times in the Radiography of Metals. R. L. Durant. (Metal Treatment, 1944, vol. 11, Autumn Issue, pp. 151–156, 160). The use of experimentally-determined exposure times, under a wide variety of conditions, for the routine examination of metals by gamma rays is discussed. The factors which influence the density of the exposed photographic film are also dealt with.

The Interpretation of Radiographs, Particularly of Aircraft Parts. L. W. Ball. (American Society for Metals, Oct., 1944, Preprint No. 3). The principles and procedure for a method for the interpretation of radiographs are outlined and methods are suggested for assessing the acceptability of defective parts.

Magnetic Analysis of Metals. N. S. Akulov. (Vestnik Metallopromyshlennosti, 1940, No. 4–5, pp. 123–129). (In Russian). Developments in Russia in the field of magnetic inspection and analysis are surveyed. A number of automatic or semi-automatic instruments have been constructed for the control of heat-treated steel parts. Of considerable interest is the use of magnetic suspensions to show up the macro- and microstructure of steel specimens. This method has given valuable results in connection with the study of changes due to deformation and precipitation. It is especially useful as a means of studying the microstructure of stainless and heat-resisting steels. In 1933 the author suggested the use of γ-$Fe_2O_3 + Fe_3O_4$ suspensions in place of iron-powder suspensions in the magnetic method of detecting cracks. A simple magnetising apparatus, not requiring accumulators, and operating directly from the A.C. lighting mains, has been developed. Other new magnetic instruments include: (*a*) An instrument for the automatic recording of the preferred orientation in rolled steel—a record is obtained in 1 min.; (*b*) an instrument for recording magnetic changes during fatigue testing (in the course of a fatigue test the permeability passes through two peaks, the second occurring shortly before failure of the specimen); (*c*) a simple instrument to record the rate of transformation of the austenite on quenching at different temperatures (with the help of this instrument an accurate S-curve can be constructed in a day); and (*d*) simple magnetic instruments for measuring the thickness of corrosion-resisting coatings.

Modification by Heat-Treatment of Cast Structures and Properties. H. T. Angus. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 74, Oct. 19, pp. 131–135; Nov. 2, pp. 171–177). An examination is made of the effect of solidification structures on the heat-treatment and properties of two cast commercial steels, namely, a medium-carbon steel and a high-carbon high-silicon steel, both made in acid electric furnaces by the duplex process of continuous pouring for producing castings weighing about 50 lb. The mechanical properties obtained after various

heat treatments of the medium-carbon steel are given, and the effect of heavily segregated dendrites and associated non-metallic inclusions is discussed. The properties and structures obtained after normalising and spheroidising the high-silicon steel are considered.

A Further Note on the Microstructure of High-Silicon Acid-Resisting Iron. J. E. Hurst and R. V. Riley. (Iron and Steel Institute, 1945, this Journal, Section I). Further observations on the nature of the "barley-shell" structure (*see* Journ. I. and S.I., 1944, No. I., pp. 221 P–224 P) are recorded. The well-defined barley-shell structure observed by W. Wrażej (Journ. I. and S.I., 1944, No. I., pp. 227 P–236 P) on pure (electrolytic) iron and steel, was found to be produced only when the etchants containing hydrofluoric acid were contaminated with silica or silicon-bearing compounds. The dissolved silica in the etchant could be derived from a number of sources, as, for example, the glass vessel in which the reagent had been stored or from a specimen of high-silicon iron alloy previously etched in the same etchant. The barley-shell structure was invariably produced on high-silicon iron alloys even when the etchant was free from silicon.

The fact that silicon or its compounds, either in the etching reagent or in the specimen, are essential to the formation of the barley-shell structure will have to be taken into consideration in any explanation of its exact nature.

Isothermal Transformation and End-Quench Hardenability of Some N.E. Steels. R. L. Rickett, J. G. Cutton, C. B. Bernhart, jun., and J. R. Millikin. (American Society for Metals, Oct., 1944, Preprint No. 5). Isothermal transformation diagrams were constructed for a number of chromium-nickel-molybdenum steels, which were also subjected to end-quench hardenability tests. The microstructure of the steels was examined. At temperatures above about 1000° F. proeutectoid ferrite and pearlite were formed. In the range of about 900–1000° F. acicular ferrite containing "dots," presumably carbide particles, formed; in some of the steels this was followed by the appearance of a dense dark-etching constituent during the later stages of transformation.

The Dimensional Stability of Steel. Part I. Subatmospheric Transformation of Retained Austenite. S. G. Fletcher and M. Cohen. (American Society for Metals, Oct., 1944, Preprint No. 27). A combined X-ray and dilatometric procedure is described for making quantitative studies of the decomposition of residual austenite in steels at subatmospheric temperatures, this being important in obtaining dimensional stability in heat-treated steel gauges. Charts are presented which show how the decomposition of the austenite is affected by different low temperatures and by the time of previous ageing at room temperature. This ageing lowers the temperature at which the retained austenite begins to transform and reduces the amount decomposed on cooling to any selected temperature. The transformation continues down to − 155° to − 160° C., irrespective of the holding time at room temperature.

A Study of Sub-zero Treatments Applied to Molybdenum-Tungsten High-Speed Steel. R. G. Kennedy, jun. (American Society for Metals, Oct., 1944, Preprint No. 28). The effect of cooling to sub-zero temperatures on hardened tool steel (containing about 3·75% of chromium, 1·75% of tungsten, 1% of vanadium and 8·25% of molybdenum) was studied by different kinds of tests. Continuous cooling to − 84° C. from 1205° C. resulted in the transformation of at least 95% of the austenite which would have transformed if the temperature had been lowered to − 190° C. The formation of additional martensite during sub-zero cooling caused an increase in the specific volume, hardness and strength as compared with ordinary hardening. The increments of hardness and strength gained by

sub-zero hardening were maintained at tempering temperatures of at least 425° C., but at 560° C. they disappeared. Sub-zero cooling after tempering 1 hr. at 560° C. produced little change in hardness, strength or plasticity when hardening temperatures of 1205° C. or higher were employed. With lower hardening temperatures, sub-zero cooling after tempering caused some increase in hardness and strength and a marked increase in plasticity.

The Partition of Molybdenum in Steel and Its Relation to Hardenability. F. E. Bowman. (American Society for Metals, Oct., 1944, Preprint No. 18). Previous investigations of the effect of molybdenum in heat-treated steel (*see* preceding abstract) were extended by studying the distribution of molybdenum between the ferrite and carbide formed isothermally at 1200° and 1100° F. from austenite and by tempering martensite. Molybdenum migrates to the carbide when the isothermal transformation takes place at these temperatures. The complex carbide $(Fe,Mo)_{23}C_6$, previously reported at 1300° F. continues to form at lower sub-critical transformation temperatures. An explanation for the increased hardenability produced in steels by molybdenum is offered; this is based on the effect of the molybdenum migration and of the altered carbide structure on the nucleation rates.

The Rate of Diffusion of Molybdenum in Austenite and in Ferrite. J. L. Ham. (American Society for Metals, Oct., 1944, Preprint No. 19). The coefficients of diffusion in pure austenitic and ferritic iron-molybdenum alloys and in pure austenitic iron-molybdenum-carbon alloys are reported. The coefficient for molybdenum was found to be much greater in ferrite than in austenite. The presence of carbon slightly increased the coefficient of diffusion of molybdenum in austenite. A mechanism is proposed to explain the effect of molybdenum on the shape of the S-curve on the basis of nucleation rates, diffusion data and the distribution of the molybdenum. Equations are presented expressing the relative rates of homogenisation in terms of the diffusion coefficients.

The Ar′ Reaction in Some Iron-Cobalt-Tungsten Alloys and the Same Modified with Chromium. W. P. Sykes. (American Society for Metals, Oct., 1944, Preprint No. 25). The Ar′ transformation was studied in an iron alloy containing 29·5% of cobalt and 17% of tungsten and in similar alloys with small additions of chromium. The times required for the initiation and completion of the $\gamma \rightarrow (\alpha + \varepsilon)$ transformation were determined approximately. The effect of chromium is to delay the onset of the transformation and to depress the position of the $(\alpha + \gamma + \varepsilon)$ field in which the γ-phase is stable.

CORROSION OF IRON AND STEEL

(Continued from pp. 35 A–36 A)

The Attack on Iron in Electrolytes in Small-Scale Tests. Part I. G. Masing. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Nov.–Dec., pp. 141–145). Laboratory corrosion tests were carried out in which pairs of plates, all of the same size, of iron and iron, iron and copper, and iron and platinum were partially immersed in a 3%-NaCl solution with the tops of the plates connected by a conductor. The loss in weight of each plate was measured after a suitable immersion period. When the plates were covered with paraffin at the water-line, each plate of a pair corroded equally, but with no protection at the water-line, the total loss in weight of each pair increased in the order iron, copper, platinum. This is due to the higher electrochemical potential at the water-line which brings the

specific activity of the metal into play. A calculation is made which demonstrates that there is a drop in potential in the electrolyte at the water-line.

The Weather-Resistance of Galvanised Steel Wires and Wire Ropes. G. Schikorr. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Nov.–Dec., pp. 147–150). An investigation of the corrosion of galvanised steel wires and wire ropes in the atmosphere of Berlin is reported. The conclusions were: (1) The corrosion in winter was much greater than in summer and it varied directly with the amount of sulphur in the atmosphere; (2) the rate of corrosion per annum remained constant over a period of years; (3) a galvanised wire rope resisted corrosion much better than a single strand of galvanised wire; this is explained by the electrochemical protective effect which the internal less-attacked parts of the wire rope has upon the external parts; and (4) the testing of thin galvanised wires by repeated dipping in copper-sulphate solution and counting the number of dips until the copper coating appears is not a satisfactory method of determining the corrosion resistance.

Replacing Non-Ferrous Metals with Nitrided Steel. V. S. Chernyak and G. A. Asinovskaya. (Vestnik Metallopromyshlennosti, 1940, No. 11–12, pp. 32–34). (In Russian). It has been found that non-ferrous parts on gas-welding equipment which have to resist corrosion by moist oxygen on heating can be successfully replaced by plain carbon steel rendered corrosion-resistant by nitriding under appropriate conditions. Depending on the service conditions to be met, the plain carbon steel parts are nitrided at 550–620° C. for 30–90 min. in ammonia dissociated to the extent of 25–40%. The parts are allowed to cool to 60° C. in the gas. The thickness of the nitrided layer amounts to 30–40 μ. Tests in moist oxygen showed that the resistance to both corrosion and wear of the nitrided parts was satisfactory. Nitrided carbon steel becomes passive in water, its potential approaching that of stainless steel.

Corrosion of Galvanized Coatings and Zinc by Waters Containing Free Carbon Dioxide. L. Kenworthy and M. D. Smith. (Journal of the Institute of Metals, 1944, vol. 70, Oct., pp. 463–489). Premature failures in cold- and hot-water tanks led to an investigation of the effect on corrosion of the free-carbon-dioxide content of the water. Conductivity-water (water of low electrical conductivity), a hard public supply water and a mixture of the two were tested at room temperature, and the hard water was tested also under conditions approximating to those in domestic hot-water systems. Various carbon-dioxide contents were artificially maintained, and both zinc and galvanised mild steel were tested under each condition. Results show that increase in free-carbon-dioxide content increases the attack in all cases. In conductivity-water, pitting was most serious with the lowest carbon-dioxide content used, but in the hot supply water there was a critical carbon-dioxide content giving maximum rate of penetration. Various types of zinc sheet showed little difference in behaviour in conductivity water. Electrochemical protection of bare steel by zinc was found to cease after a short time, both in conductivity-water with a low carbon-dioxide content (8–18 days) and in hot supply water (3–4 days); in the cold supply water, however, this protection is maintained for many months, until all the zinc coating has been removed. In hot water, as much as 50–80% of the original coating may remain when failure by pitting occurs. The importance of gas bubbles in initiating pitting in hot tanks is discussed. Recommendations are made with regard to the type and thickness of coatings as well as the treatment of water supplies.

Supplemental Protection for Black-Oxide Finishes. M. Weisberg and E. A. Parker. (Industrial Heating, 1944, vol. 11, Aug., pp. 1304–1314).

Results of a series of salt-spray tests on the effectiveness of different types of corrosion inhibitor on steels coated with black-oxide finish are presented and discussed. Adhesion tests on synthetic-resin refrigerator enamels applied to the same type of surface are also discussed.

Corrosion Prevention Improved by New Wrapping System. (Steel, 1944, vol. 115, Sept. 4, p. 94). A method of wrapping is described which affords complete protection from corrosion of highly finished steel parts. The wrapping material consists of a double "kraft" paper the interior of which is impregnated with cellulose acetate.

Corrosion Protection of Metals. K. G. Compton. (American Society for Metals: Canadian Metals and Metallurgical Industries, 1944, vol. 7, July, pp. 27–28, 53). The mechanism of the corrosion of metals and the inhibitory effects of natural and applied coatings are briefly reviewed.

ANALYSIS

Colorimetric Determination of Nickel in Steel. G. R. Makepeace and C. H. Craft. (Industrial and Engineering Chemistry, Analytical Edition, 1944, vol. 16, June, pp. 375–377). The method of W. M. Murray and S. E. Q. Ashley for the colorimetric determination of nickel in steel (*see* Journ. I. and S.I., 1938, No. I., p. 360 A) is critically examined and data on the stability under various conditions of the red colour of oxidised nickel dimethylglyoxime are presented. A modified method which gives a very stable red colour and is particularly suitable for routine work is described.

Colorimetric Method for Determining Phosphorus in a Combined Form of Analysis. W. B. Sobers. (American Foundryman, 1944, vol. 6, Sept., pp. 2–4). Supplementing a previous paper on routine analyses of cast iron and steel (*see* Journ. I. and S.I., 1944, No. II., p. 65 A) colorimetric methods of determining phosphorus are described.

A Rapid Colorimetric Process for the Determination of Phosphorus in Steel. K. A. F. Schmidt and K. Kutil. (Stahl und Eisen, 1944, vol. 64, Aug. 17, pp. 539–540). A method similar to that of G. Bogatzki (*see* Journ. I. and S.I., 1939, No. I., p. 128 A) for the rapid determination of phosphorus in steel is described. Whereas Bogatzki treats the phosphoric acid first with ammonium molybdate and then with ammonium vanadate, the present authors obtain a better colour effect by reversing this order.

Methods for the Photometric Determination of Cobalt in Tool Steels and High-Speed Steels in the Presence of Iron and Other Alloying Elements. G. Bogatzki. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Nov.–Dec., pp. 125–126). A photometric method for the determination of cobalt in tool steels which makes use of the colour produced when an ammoniacal solution of cobalt is treated with potassium ferricyanide is described. The determination takes 25 to 30 min.

The Determination of Oxygen in Steel. A. Gotta. (Archiv für Eisenhüttenwesen, 1943, vol. 17, Sept.–Oct., pp. 53–55). A new method for determining oxygen in steel is described. The sample is prepared in thin strips about 0·7 mm. thick and 10 mm. long; these are buried in a fine powder consisting of 1 part of aluminium mixed with 3 parts of alumina in an alumina crucible which is heated up to 1300° C. in an electric furnace. The aluminium diffuses into the steel and combines with the oxygen in it. The amount of this alumina is determined by wet methods. The method is not so difficult as the hot-extraction procedure and, as the sample is not melted, the errors which this may cause do not arise.

The Use of Perchloric Acid in the Determination of Nitrogen in Iron and Steel. H. Kempf and K. Abresch. (Archiv für das Eisenhütten-

wesen, 1943, vol. 17, Nov.–Dec., pp. 119–124). The methods of determining nitrogen in iron and steel are examined with a view to their improvement and simplification, and the literature on the use of perchloric acid is reviewed. Some discrepancies which have arisen when perchloric acid was used are explained. A new procedure by which the nitrogen in plain carbon steel can be determined in 10 min. is described.

Spectrographic Analysis of Iron and Steel. H. F. Kincaid. (Transactions of the American Foundrymen's Association, 1944, vol 52, Sept., pp. 248–260). The use of the spectrograph at the Farmall Works of the International Harvester Company is described (*see* Journ. I. and S.I., 1944, No. II., p. 117 A).

A Graphic Calculation for Spectrochemical Analyses. A. P. Vanselow and G. F. Liebig, jun. (Journal of the Optical Society of America, 1944, vol. 34, Apr., pp. 219–221). A graphic calculator for computing quantitative spectrographic analyses is described. It embodies the better features of several calculators described in the literature and the authors' own ideas.

Rapid Routine Methods for Slag Analysis. R. A. Willey. (Foundry, 1944, vol. 72, Sept., pp. 87, 180–184). Rapid methods of determining the magnesia, lime and ferrous oxide in acid open-hearth furnace slags are described.

REFRACTORY MATERIALS

(Continued from pp. 40 A–41 A)

Significance of a Steel-Plant Refractories Testing Program. W. S. Debenham. (Bulletin of the American Ceramic Society, 1944, vol. 23, Sept. 15, pp. 317–320). The purposes of a routine programme for testing steelworks' refractories and the problems involved in its organisation are considered. A testing programme can bring about economies by: (*a*) Checking the quality of incoming refractories; (*b*) evaluating commercially available refractories to determine the applications for which they are most suited; (*c*) evaluating new refractories; and (*d*) providing the necessary co-operative link between consumer and producer. Some acceptance tests for super-duty bricks, ladle bricks, stopper heads and silica bricks are described.

The Action of Steam and Sulphur Dioxide on Refractory Materials. Part I. The Action on Fireclay at 1200° C. L. R. Barrett, N. E. Dobbins and A. T. Green. (Transactions of the British Ceramic Society, 1944, vol. 43, Sept., pp. 155–170). An attempt was made to separate the reactions between fired fireclay and the components of a sulphureous atmosphere, namely, oxygen, steam, sulphur dioxide, sulphur trioxide and sulphuric acid vapour. The results obtained at 1200° C., along with some previous observations at other temperatures, are described. It was found that the shrinkage and reduction in porosity induced by reheating bars for periods up to 222 hr. at 1200° C. were greater in steam and in sulphur dioxide than in air, oxygen, sulphur trioxide or sulphuric acid vapour, where the shrinkages were similar. Bars heated in steam and pure sulphur dioxide experienced a colour change from light biscuit to grey or brown, and were shown to contain ferrous iron. The extra shrinkage and the colour change in steam and sulphur dioxide would seem to be due to the absence of free oxygen, which permitted the steam and sulphur dioxide to exert a reducing action. Reheating increased the modulus of rupture in each case. The presence of steam aided vitrification, whilst sulphur trioxide nullified to some extent the effect of reheating. This may be ascribed to the loss of bond, since in this case notable proportions of lime and magnesia were removed as sulphates on leaching with water. The removal of sulphates was also observed to a lesser extent with the bars that had been treated in sulphur dioxide and sulphuric acid vapour. The quantity of sulphate formed depended on the temperature of the reaction and the atmosphere in which cooling took place, being most when it was sulphureous and least when it was inert.

The Action of Steam and Sulphur Dioxide on Refractory Materials. Part II. The Action on Fireclay at 600° C. L. R. Barrett, N. E. Dobbins and A. T. Green. (Transactions of the British Ceramic Society, 1944, vol. 43, Sept., pp. 170–179). Fireclay specimens in the form of bars measuring $4{\cdot}5 \times 0{\cdot}5 \times 0{\cdot}5$ in. were exposed at 600° C. to atmospheres of steam, oxygen, sulphur dioxide, sulphur trioxide and oxygen in the ratio of 2 to 1 by volume, and sulphuric acid vapour for periods up to 222 hr. The alterations in weight, length, porosity and transverse strength were determined. On exposure to the sulphuric acid vapour and to the mixture of sulphur dioxide and oxygen, considerable increases in weight occurred, accompanied by slight contraction, marked decrease in porosity and increase in the modulus of rupture. On extracting these specimens with hot dilute hydrochloric acid solution, appreciable quantities of the fireclay material passed into solution, pointing to the formation of sulphates of all

the bases present, but more particularly iron and aluminium sulphates. Reheating at 1000° C. in air of the specimens exposed to sulphuric acid vapour or to sulphur dioxide and oxygen resulted in the decomposition of the sulphates, accompanied by a permanent expansion of the same order of magnitude as the contraction occurring when the sulphates were formed.

The Action of Chlorine on Refractory Materials. Part III. The Effect of Chlorine Treatment on the Physical and Chemical Properties of Some Commercial Products. H. M. Richardson, F. H. Clews, L. R. Barrett and A. T. Green. (Transactions of the British Ceramic Society, 1944, vol. 43, Sept., pp. 179–185). Continuation of a series of papers (*see* Journ. I. and S.I., 1943, No. I., p. 73A). Firebrick, magnesite, chrome and diatomite refractory products were treated with chlorine at 1000° C. Firebrick was found to be relatively resistant to attack, although enough iron oxide was removed to cause resistance to disintegration by carbon monoxide at 450° C. Volatilisation of the constituents of magnesite and chrome products resulted in a large reduction in the cold crushing strengths of these bricks. Purification of the diatomite was effected by the chlorine treatment, with the result that the working temperature of the new product could be increased to about 1100° C.

Ferrous and Magnesium Chromites. Specific Heats at Low Temperatures. C. H. Shomate. (Industrial and Engineering Chemistry, Industrial Edition, 1944, vol. 36, Oct., pp. 910–911). Specific heats of ferrous chromite and magnesium chromite were measured throughout the temperature range 52–298° K. Ferrous chromite has two anomalies in its specific heat curve, one peak occurring at 75° and the other at 135° K. The following molar entropies at 298·16° K. were computed : Ferrous chromite $34{\cdot}9 \pm 0{\cdot}4$ E.U. (entropy units) and magnesium chromite $25{\cdot}1 \pm 0{\cdot}2$ E.U.

High-Temperature Heat Contents of Ferrous and Magnesium Chromites. B. F. Naylor. (Industrial and Engineering Chemistry, Industrial Edition, 1944, vol. 36, Oct., pp. 933–934). The heat contents above 25° C. of high-grade synthetic samples of ferrous chromite and magnesium chromite were determined in the temperature range 25–1500° C. The heat contents and entropies at temperatures increasing from 400° K. by 100° stages to 1800° K. are given in a table.

Industrial Survey of Refractory Service Conditions in Electric Furnaces Used in Steel Manufacture. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 230–240). Detailed descriptions are given of the construction of refractory linings for basic and acid electric arc furnaces and for basic and acid coreless induction furnaces.

Thermal Insulation : The Properties and Processes which Control the Flow of Heat in Continuous and Intermittent Heating Operations. H. R. Fehling. (Journal of the Institute of Fuel, 1944, vol. 18, Oct., pp. 15–25). The author presents a comprehensive theory governing thermal insulation with special reference to the influence of porosity and heat capacity on the insulating efficiency of the material used. All composite walls represent at least three thermal resistances put in series. Hence, the heat flow through them is controlled by the strongest resistance in the series, which should be the insulation. The application of an insulating layer not only affects the temperature gradients throughout the wall, but also the thermal resistances of the surfaces. For heavy insulation the suitability of the material as regards the maximum safe temperature is more important than its conductivity. In all intermittently heated walls there is a fairly sharp dividing line between a zone of strongly fluctuating temperatures near the hot face and a nearly stationary temperature zone near the cold face ; this leads to different specifications for the most efficient materials for the two zones. The heat stored in insulating bricks is only a fraction

of that absorbed by ordinary refractory walls in the same time, the degree of reduction being much greater than the degree of reduction in weight, because the product of the conductivity and heat capacity is the controlling factor.

FUEL

(Continued from pp. 41 A–42 A)

Reactivity of Solid Fuels. A. A. Orning. (Industrial and Engineering Chemistry, Industrial Edition, 1944, vol. 36, Sept., pp. 813–816). Reactivity indices are presented for 68 coals and 35 cokes. The indices, varying inversely with the reactivity, were found to correlate with the volatile-matter content on the dry ash-free basis. The indices are highest for cokes, decrease rapidly with increasing volatile-matter content in the anthracite range, are relatively constant for medium-volatile bituminous coals, and again decrease rapidly for the high-volatile coals and lignites.

Gas Flow Control Systems. W. J. Roberts. (Coke and Smokeless Fuel Age, 1944, vol. 6, Oct., pp. 191–193). Devices for the continuous metering and regulation of fuel gas and air supplies to furnaces and automatic proportioning instruments for maintaining two variables in constant proportion are described.

Rare and Uncommon Elements in Coal. F. H. Gibson and W. A. Selvig. (United States Bureau of Mines, 1944, Technical Paper No. 669). This brief report summarises the available information on the rare elements in coals, not only those mined in the United States, but also in other countries. A brief account of the common inorganic constituents of coal ash and the mineral matter in coal is included.

Proposed Method of Test for Expansion Pressure of Coal during Coking. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 293–296). An apparatus for measuring the swelling pressure of coking coal is described. It consists of an electrically heated chamber 28 in. wide by 44 in. high, with one fixed and one movable wall mounted on a carriage with ball-bearing wheels. Details of the proposed testing procedure are also given.

Proposed Method of Test for Expansion Properties of Coal for Use in By-Product Coke Ovens by Means of the Bethlehem Test Oven. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, p. 314–361). The apparatus for testing the swelling properties of coking coals described in this paper consists of a horizontal carbonising chamber heated by a combustion chamber under the floor, and having a weighted rectangular piston on top, the movements of which are followed by a recording device.

Proposed Method of Test for Plastic Properties of Coal by the Davis Type Plastometer. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 297–300). A method is proposed for determining the changes in the plastic properties of coal at different temperatures. The apparatus consists of a horizontal electric furnace 2 in. in inside dia. × 12 in. long with temperature-recording equipment. An 18-g. sample of the coal is put in a small steel retort, which is slowly rotated within the furnace. A shaft carrying four rabble arms extends into the retort, and the changes in the plasticity of the coal are determined by measuring the torque exerted by the coal on the rabble arms.

Proposed Method of Test for Plastic Properties of Coal by the Gieseler Type Plastometer. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 301–305). A method for determining the plastic properties of coal is described. This is the Gieseler plastometer

method, in which the sample is packed in a crucible containing a vertical spindle with four rabble arms; the upper end of the spindle carries a drum with a cord which passes over a pulley carrying a pointer. A 40-g. weight is suspended at the end of the cord. When the crucible is heated by immersion in a bath of molten metal at 300° C. and the brake on the spindle is released, the movement of the rabble arms in the coal is measured by the cord moving the pointer in front of a dial.

The Gieseler Method for Measurement of the Plastic Characteristics of Coal. G. C. Soth and C. C. Russell. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 1176–1189). A modified Gieseler plastometer and its mode of operation are described, together with a method of calibrating the apparatus so as to give test results in absolute units. Difficulties met with in the measurement of the plastic characteristics of coal are discussed. It is shown that the plastic range of the usual coking coals is between 340° and 510° C. when the coals are heated at the rate of 3° C. per min. The range of viscosities measurable by the modified Gieseler plastometer lies between $6{\cdot}4 \times 10^7$ and $6{\cdot}6 \times 10^2$ poise.

Proposed Method of Test for Carbonization Pressure of Bituminous Coal. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 306–309). A small-scale laboratory method is described for measuring the swelling pressure of coking coals. A rectangular silica-brick chamber made in two halves is used; one half is fixed and the other is mounted on ball-bearing wheels. The pressure exerted on the movable wall is transmitted to bellows connected to a Bourdon gauge.

The Mode of Combustion of Coke. J. Hiles and R. A. Mott. (Fuel in Science and Practice, 1944, vol. 23, Nov.-Dec., pp. 154–171). Details are recorded of a study by the Midland Coke Research Committee of the combustion and gasification of coke under a variety of conditions of size and quality of fuel, rate of combustion and size of the combustion unit. A modified two-zone theory of combustion and gasification is postulated; this explains satisfactorily the detailed studies reported, and reconciles the need for modification of the conventional two-zone theory envisaged by Bangham and Bennett and by the work of Chukanov and his associates. The oxygen of an air supply to a bed of coke is virtually used up within ½–9 in. of the point of air entry whatever its rate of supply within a range sufficient to burn coke at 1–65 lb. per sq. ft. per hr. The height of the oxidation zone is but little affected by the rate of air supply, or by the specific reactivity of the fuel, and is determined essentially by the size of the fuel. In general, the dominant factor which influences the reactions in a burning bed of coke is particle size.

PRODUCTION OF IRON

(Continued from pp. 42 A–44 A)

Strength of Concrete as Related to Abrasion of the Blast-Furnace Slag Used as Coarse Aggregate. F. Hubbard and H. T. Williams. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 1088–1094). The results of an examination of the properties of blast-furnace slags, together with strength tests on concrete made with slag, are presented. Results of abrasion tests on slag show a marked similarity between the Los Angeles and the Deval test results. It is suggested that abrasion tests are not significant in determining the quality of blast-furnace slag aggregate, as judged by the compressive and bend strengths of the concrete.

Electrometallurgical Treatment of Ores. C. Hart. (Year Book of the American Iron and Steel Institute, 1944, pp. 162–175). The use of electric furnaces for making ferro-alloys and alloy steels is reviewed and some developments in electrode and refractory brick manufacture are commented on. The Bureau of Mines has developed an electrolytic method of producing manganese of 99·9% purity. The process was technically successful, but the production of 5 tons per day by two plants soon saturated the market.

Nitriding Ferroalloys. R. H. Steinberg and D. Steinberg. (Metals and Alloys, 1944, vol. 19, Apr., p. 859; vol. 20, Sept., p. 631). High-nitrogen ferro-chromium has been used in steel furnaces to refine the grain when making high-chromium steels. To determine whether other ferro-alloys would take up nitrogen, experiments were conducted with 2-g. samples in powder form by holding them at temperatures up to 875° C. for 1 hr. in ammonia. Low-silicon (16%) ferro-silicon nitrided easily, but 78% ferro-silicon did not take up any nitrogen at all. The optimum temperature was 400–600° C. for nitriding iron, 600–800° C. for pure manganese, 600–700° C. for ferro-manganese and 600° C. for 16% ferro-silicon.

Powder Metallurgy. A. Manufacturing Methods and Limitations. J. H. Bull. (Australian Institute of Metals: Australasian Engineer Science Sheet, 1944, Sept 7, pp. 2–4). Methods of manufacturing parts by powder metallurgy are described. In some instances the pressing, sintering and sizing are done in separate operations each requiring its special plant, but in others, such as making cemented-carbide forming dies, the three operations can be combined by hot-pressing the preheated powder.

Powder Metallurgy. B. Applications, Uses and Trends. J. H. Robinson. (Australian Institute of Metals: Australasian Engineer Science Sheet, 1944, Sept. 7, pp. 5–7). The advantages of applying powder metallurgy to the manufacture of a wide variety of articles are outlined.

FOUNDRY PRACTICE

(Continued from pp. 44 A–46 A)

Quality Control in the Foundry. E. W. Harding. (Canadian Metals and Metallurgical Industries, 1944, vol. 7, Aug., pp. 39–46; Sept., pp. 37–40). *See* Journ. I. and S.I., 1944, No. II., p. 4A.

The Application of Radiography to the Improvement of Foundry Technique. R. Jackson. (Iron and Steel Institute, 1945, this Journal, Section I). The principles involved in the taking and interpretation of a radiograph of a casting are given and the nature and extent of the defects which are revealed are discussed. Examples are given of the application of radiography to the examination of steel castings and of the methods adopted to improve the quality of castings thus examined. Tests carried out on steel castings improved by these means have shown their strength to be much greater than had been previously recognised.

Modern Cast Iron. A. J. Milgate. (Australian Institute of Metals: Australasian Engineer, 1944, vol. 44, Sept. 7, pp. 29–35). A brief history of the development of cast iron is followed by an account of some aspects of the manufacture of high-duty cast irons of tensile strengths of the order of 25 tons per sq. in. Factors in cupola operation and inoculation processes are dealt with, and some of the practical difficulties associated with the production of high-duty cast iron reviewed. Notes on heat treatment and flame-hardening are included, and the more important physical properties of this rapidly developing material are listed.

Steel Foundry Pouring Practice. J. H. Hall. (Foundry, 1944, vol. 72, Oct., pp. 74–75, 210–216). The advantages and disadvantages of different types of ladle used in steel foundries and the design of nozzles and stoppers for bottom pouring are discussed.

The Investigation of Dried Moulding Sands. H. Baresch. (Giesserei, 1944, vol. 30, Apr., pp. 55–59). Methods of testing dried moulding sand for steel castings are discussed. From tests it was observed that, when a green sand had a maximum permeability with 3% of moisture, dry sand of the same type had a maximum permeability when the initial moisture content was 4%. Curves showing the effect of initial moisture content on the strength and permeability of a number of sands are presented.

The Principles of the Testing of Mould and Core Sands and Their Development Possibilities with Special Reference to the Fischer Testing Apparatus. W. Goetz. (Giesserei, 1944, vol. 30, June, pp. 91–95). Having regard to the advancement in the knowledge of the properties and characteristics of moulding sands, improvements in sand-testing apparatus have been introduced. The fundamentals of sand testing are considered and some recent improvements which have been developed by a German firm making test apparatus are described.

New Knowledge and Proposals on the Subject of Testing Moulding Sand. W. Reitmeister. (Giesserei, 1944, vol. 31, Sept., pp. 136–141). Factors affecting the apparent density of moulding sand mixtures are discussed and data are presented relating this property to the strength and permeability.

A New Moulding Process for Grey Iron Castings. K. Grunke. (Giesserei, 1944, vol. 31, Aug., pp. 126–128). The Gödel cement-sand moulding process is briefly described. In this, a cement-sand is used as pattern sand, and the remainder of the mould is filled up with silver sand. No hand or pneumatic ramming is necessary. Examples of two large castings, the moulds for which were prepared in this way, are described and illustrated.

Making Practice Bombs. (Canadian Metals and Metallurgical Industries, 1944, vol. 7, Sept., pp. 22–27). The foundry technique developed by the Grinnell Company of Canada, Ltd., for the mass production of practice bombs, which are grey iron castings weighing 11¾ lb., is described.

High-Strength Centrifugal Castings. S. P. Perry. (Iron Age, 1944, vol. 154, Oct. 5, pp. 52–53). Recent researches resulting in the production of centrifugally-cast steels having strengths of the order of 180,000 lb. per sq. in. and a ductility of 10% are briefly described. The tests were carried out on specimens of tubular shape of 6¾ in. outside diameter, 1-in. wall and 7⅝ in. in length.

Economic Compressed-Air Supply in Foundries. H. Allendorf. (Giesserei, 1944, vol. 31, Aug., pp. 117–125). The theory and practice of the production of compressed air are discussed and data are presented on costs of production, compressor ratings, air requirements for various operations in the foundry, pipe-line resistance and other causes of loss.

Efficient Utilization of Compressed Air in the Foundry. J. L. Yates. (Foundry, 1944, vol. 72, Oct., pp. 80–81, 238). Recommendations are made on the maintenance of air compressors, piping and compressed-air equipment in foundries, and a chart is presented for the rapid determination of the pressure drop when air passes through pipe lines of different sizes.

PRODUCTION OF STEEL

(Continued from pp. 46 A–49 A)

The Western Steel Industry. J. R. Mahoney. (Iron Age, 1944, vol. 154, Sept. 21, pp. 67–73, 166–168; Steel, 1944, vol. 115, Sept. 25, pp. 100–102, 124–128). The economic aspects of the new iron and steel works at Geneva, Utah, and at Fontana, California, are discussed. The costs of raw materials for these works are compared with those applying to plants in the eastern and southern states, and the probable extent of the post-war markets for the western steel industry is dealt with.

Republic Builds for D.P.C. T. J. Ess. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. R-18–R-38). A comprehensive illustrated description is given of the integrated blast-furnace and steel plant at South Chicago which was completed in 1943 for the Defense Plant Corporation. It is being operated by the Republic Steel Corporation. The plant includes one blast-furnace with a 27-ft. hearth rated at 1250 tons of pig-iron per day, four 250-ton tilting open-hearth furnaces and nine basic-lined arc furnaces. The nominal capacity of the plant is 750,000 tons of ingots per annum.

The Use of Graphite in the Open-Hearth Charge. J. O. Griggs. (Blast Furnace and Steel Plant, 1944, vol. 32, Sept., pp. 1065–1067). The advantages and disadvantages of charging graphite and scrap in the open-hearth furnace instead of molten pig-iron to obtain the required carbon are discussed. It is expected that after the war large quantities of steel scrap will become available at reduced prices; there will therefore be less inducement to use hot metal. In the basic process there is much less furnace-bottom maintenance when graphite is used because there is 60% less silicon in the melt than when working with hot metal. The amount of phosphorus will also be less by about 50%, and this means the fluxes can be reduced by about 75%.

Cheap High-Duty Steels. J. Edmiston. (Metallurgia, 1944, vol. 30, Oct., pp. 304–305). As large quantities of steel scrap containing small amounts of copper, nickel and molybdenum will become available from unwanted guns, armament, etc., at the end of the present war, it is suggested that these steels should be used as a basis for manufacturing high-duty alloy steels.

Basic Open-Hearth as a Producer of Alloy and Special Steels. C. D. King. (Metal Progress, 1944, vol. 46, Sept., pp. 478–480). Some methods and equipment which have contributed to the increase in the production of high-quality steels in basic open-hearth furnaces in the United States are pointed out. These include the careful selection of raw materials, improved slag control, more rapid analyses and better temperature control.

Electric Furnace Quality *versus* Open-Hearth. G. Soler. (Metal Progress, 1944, vol. 46, Sept., pp. 480–483). Some advantages of the electric furnace in producing clean steel of high quality are discussed. The practice of working with two slags has advantages in flexibility of control, and, as the slag volume is small in relation to the amount of steel, its composition can be adjusted much more rapidly with small additions than the comparatively large volume of slag in open-hearth furnaces.

FORGING, STAMPING AND DRAWING

(Continued from p. 49 A)

Forging Die Design. J. Mueller. (Steel Processing, 1944, vol. 30, Sept., pp. 563–564). Three pairs of dies used in the manufacture of a gear selector fork are described and illustrated.

Designing Rubber Press Tools. J. Albin. (Iron Age, 1944, vol. 154, Sept. 14, pp. 50–57; Sept. 21, pp. 59–66). Illustrated descriptions are given of presses for forming parts out of sheet metal by placing the blank over a block of the desired shape which rests on a plate on the bolster; the descending ram has layers of rubber cemented together on its face. The sheet metal is formed round the block by the rubber under pressure. The advantages of the process in producing shapes which cannot be obtained by pressing with metal to metal dies are pointed out.

Progress in Hot Cropping Operations. A. G. Arend. (Drop Forger, 1944, vol. 24, Nov., pp. 106–108). The application of hot saws and the layout of forging machines in relation to the hot saw which cuts off the billets or blanks to the required size are discussed.

Bending and Forming Heavy Steel Sections Hydraulically. C. W. Hinman. (Steel Processing, 1944, vol. 30, Aug., pp. 490–492). Modern presses and dies are described which are used for straightening shafts and for making angles, Z bars and other shapes out of flats.

Shell Forging—Metal Flow and Associated Forces. M. D. Stone. (American Society of Mechanical Engineers: Steel Processing, 1944, vol. 30, June, pp. 353–354, 382–383; Aug., pp. 493–497, 503). Efficiency in the forging of shells is considered with special reference to the depth obtainable by piercing, the shape of the mandrel, the optimum piercing speed and the optimum billet shape. The use of roller-type dies and "upsetter" methods for shell forging are described.

Induction Heating for Forging Shells. (Metals and Alloys, 1944, vol. 20, Sept., pp. 631–635). The manufacture of 105-mm. shells at the Salem plant of the Mullins Manufacturing Corporation is described by a series of 15 illustrations with explanatory notes. The heating of the billets prior to forging is done by induction heating units.

The Manufacture of Large Marine Engine Forgings. R. Liston. (Journal of the West of Scotland Iron and Steel Institute, 1943–44, vol. 51, Part V., pp. 107–121). A detailed account is given of the forging of large marine-engine crankshafts, intermediate shafts and boiler drums. Factors governing the choice of ingot are discussed and a large hydraulic forging press is described.

Stainless Steel Stampings for Ship Parts. W. J. Meinal. (Metals and Alloys, 1944, vol. 20, Sept., pp. 616–619). Descriptions are given of the construction of a ship's funnel and a hatch cover for an aircraft carrier both of which were made of stainless steel stampings and sheet assembled by welding.

History and Growth of the Tungsten Carbide Die. A. R. Zapp. (Wire and Wire Products, 1944, vol. 19, Sept., pp. 543–546, 569, 570–571). The history of the discovery and development of the tungsten-carbide die for drawing wire is described.

Tension and Velocity Controls; Applications of the Mechanical Variable Speed Transmission. J. H. Gepfert. (Wire and Wire Products, 1944, vol. 19, Oct., pp. 682–691, 729). Mechanical methods of varying the speed of wire-mill and wire-drawing machinery and methods of controlling the speed are described.

HEAT TREATMENT

(Continued from pp. 50 A–53 A)

A Microstructural Guide to the Heat Treatment of a Plain Carbon Steel. R. J. Randebaugh and W. E. Fontaine. (Metals and Alloys, 1944, vol. 20, Aug., pp. 359–364). A series of five charts is presented which indicate the effects of heat-treating a bar of a 0·4% carbon steel in different ways. The charts consist of a combination of cooling curves and micrographs of the structures obtained.

Heat-Treatment Plant for Alloy Steel Bars and Sections. (Engineering, 1944, vol. 158, Nov. 24, p. 407; Engineer, 1944, vol. 178, Dec. 1, pp. 438–439). A new heat-treatment plant at the works of the Hallamshire Steel and File Co., Ltd., Sheffield, is described. It is designed for the continuous production of annealed, normalised, or fully hardened and tempered alloy steel bars up to 6 in. in dia. and for other shapes. It consists of four recuperative producer-gas-fired furnaces, two oil-quenching tanks, one water-quenching tank and four loading racks in line served by an electric charging machine.

Leaf Spring Production. J. Ade. (Steel, 1944, vol. 115, Sept. 25, pp. 80–81, 136). The continuous heat-treatment plant of the Standard Steel Spring Company is described and illustrated. This is specially designed for the heat treatment of leaf springs from 1 ft. to 6 ft. in length. Both the hardening and tempering furnaces have a chain link conveyor running through them, the speed of which is adjustable so that the heating time can be controlled.

Heat Treating 5-in. Navy Shells. (Iron Age, 1944, vol. 154, Oct. 5, pp. 64–65). A new conveyor type of heat treating plant used by the Houdaille-Hershey Corporation at Decatur, Illinois, for the production of 5-in. shell cases is described and illustrated.

The Siliconizing of Steel. D. McPherson. (Machinery, 1944, vol. 65, Nov. 23, pp. 571–573). The Ihrigising process is described and details of the structure and composition of the silicon case are given. The steel parts are packed in either silicon carbide or ferro-silicon and heated to about 1000° C. Chlorine is fed into the container or into the furnace chamber to accelerate the reaction. The siliconised case on iron and steel is characteristically soft and ductile having a Rockwell hardness of 80–85 B. The case retains oil in microscopic pores in the surface, and siliconised parts can be made self-lubricating simply by boiling in lubricating oil.

A Formulation of the Carburizing Process. F. E. Harris and W. T. Groves. (Metal Progress, 1944, vol. 46, Sept., pp. 488–496). Ideal carbon concentration gradients are constructed from data obtained by a mathematical analysis of results of gas-carburising experiments described in earlier papers by F. E. Harris (*see* Journ. I. and S.I., 1944, No. I., p. 17 A and 1945, No. I., p. 16 A). These ideal gradients can be applied to predict the results which would be obtained under a variety of carburising conditions.

Essential Characteristics of Controlled Atmospheres. W. D. Vint. (Metallurgia, 1944, vol. 30, Oct., pp. 293–296). Brief descriptions of three types of plant for producing protective atmospheres for annealing furnaces are given; these types are: (1) The cracked ammonia regenerative system; (2) the burnt town's gas plant; and (3) the blast-furnace-gas cracker. The effects of moisture, hydrogen, methane, the CO_2/CO ratio and sulphur in the atmosphere are discussed.

The Principles and Practice of Lithium Heat Treating Atmospheres. Part I. Neutral Heat-Treating Atmospheres. C. E. Thomas. (Industrial

Heating, 1944, vol. 11, Sept., pp. 1405–1416). After a discussion of the requirements of heat-treating atmospheres it is shown how lithium can be applied to keep furnace atmospheres neutral and entirely free from water vapour. A description is given of a small muffle furnace in which a partially burned gaseous fuel is first used to vaporise lithium so that the mixture enters the muffle containing the parts to be heat-treated. The reactions of lithium with water vapour, oxygen and carbon monoxide are explained and it is shown that there can be no water vapour in the atmosphere with lithium present and in this dry state the atmosphere is inert so that steels of different carbon content can be treated without oxidation or loss of carbon.

Fundamental Principles and Applications of Induction Heating. (Sheet Metal Industries, 1944, vol. 20, Aug., pp. 1389–1396; Sept., pp. 1573–1581; Oct., pp. 1757–1765; Nov., pp. 1941–1949, 1970). The principles of induction heating are explained and descriptions of several applications including surface hardening, hardening the edges of strip, brazing, and heating railway rolling-stock tyres to shrink them on to wheels are given with details of the water-cooled coils of diverse shapes. Factors affecting the choice of equipment for particular purposes are reviewed.

Continuous High Frequency Heat Treatment. (Steel Processing, 1944, vol. 30, Sept., pp. 578–579). A brief description of the "Megatherm" high-frequency induction unit is given. This is used by one firm for the surface hardening of steel pins ½ in. in dia. × 2½ in. long at a rate of 75 per min.

Continuous Hardening of Steel Bars by Induction. (Steel Processing, 1944, vol. 30, Sept., pp. 574–575). An illustrated description is given of a heat-treatment plant for the induction-hardening of steel bars up to 1¼ in. in dia. The plant comprises four hardening units to which current is supplied by two 9600-cycle, 440 V., generating sets. Each unit carries two water-cooled copper inductor coils, one for hardening and one for tempering. The bars are fed end to end into constant-speed rolls which carry them forward through the coils. Bars ⅞ in. dia., for example, pass through the coils at 22·8 in. per min. The four units have a capacity of 1 ton of ⅞-in. dia. bars per 2 hr. and they require only two operators.

Subzero Treatment to Improve Tool Life. S. M. DePoy. (Metals and Alloys, 1944, vol. 20, Sept., pp. 645–649). The effect of refrigeration on the hardness, impact strength and cutting capacity of high-speed steels is discussed and the results of tests are given in which tools were heat-treated in the following manner: (1) Held for 2½ min. at temperatures increasing from 2175° to 2410° F.; (2) quenched for 3 min. in a salt bath at 1100° F.; (3) cooled in air down to 100–150° F.; (4) washed thoroughly; and (5) tempered twice for 2 hr. at 1050° F. with and without an intermediate cold treatment at −105° F. for comparison purposes. In all cases the cold treatment resulted in increased hardness, lower impact strength and longer life of the tool.

Extending Cutting Tool Life by Refrigeration and Nitriding. C. M. Lucas and H. M. Hartley. (Iron Age, 1944, vol. 154, Oct. 12, pp. 74–77, 168). High-speed-steel cutting tools which suffer from low hardness, carbide segregation and incorrect grinding in the as-received condition may be improved by refrigeration followed by salt-bath tempering or nitriding, regrinding and honing. Operations of this kind are described and some typical defects in tools are illustrated. The application of salt-bath nitriding in the tool room as a means of increasing the life of taps, chasers and milling cutters is also described.

Facilities for Water Quenching Steel Castings. R. H. Swartz. (Steel Founders' Society of America: Foundry, 1944, vol. 72, Oct., pp. 82–84, 230–236). One of the papers presented at a symposium on the water-

quenching of steel castings is reproduced (*see* Journ. I. and S.I., 1944, No. II., p. 153 A).

WELDING AND CUTTING

(Continued from pp. 53 A–54 A)

Recent Progress in the Scientific Application of Welding to Steel. W. F. Hess. (Year Book of the American Iron and Steel Institute, 1944, pp. 115–131). The manner in which the following factors affect the cooling rates of steel after electric welding are discussed : (1) The thermal properties of the steel ; (2) the heat energy input per unit length of weld ; (3) the plate thickness ; (4) the plate temperature ; and (5) the design of the joint. A method of determining the best welding procedure to apply to a particular steel is described.

Metallic Arc Welding Electrodes. H. Lawrence. (Steel, 1944, vol. 115, Sept. 11, pp. 98–101, 144–146 ; Sept. 18, pp. 110–111, 166–176 ; Sept. 25, pp. 86–88, 130–134 ; Oct. 2, pp. 98–103, 114–116 ; Oct. 9, pp. 136–138, 304–308 ; Oct. 16, pp. 104). Continuation of a series of articles (*see* Journ. I. and S.I., 1944, No. II., p. 157 A). The properties of welds made with various types of stainless steel electrodes, electrodes for hard-surfacing and cast-iron, nickel, nickel-alloy, copper and aluminium electrodes are discussed with notes on the best welding technique to adopt.

Determination of Cooling Rates of Butt and Fillet Welds as a Result of Arc Welding with Various Types of Electrode on Plain Carbon Steel. W. F. Hess, E. F. Nippes, L. L. Merrill and A. P. Bunk. (Welding Journal, 1944, vol. 23, Aug., pp. 376-S–391-S). This investigation of cooling rates after welding is an extension of an earlier one (*see* Journ. I. and S.I., 1944, No. I., p. 56 A). The conclusions reached were : (1) The cooling rates obtained on welding with austenitic electrodes on mild steel plate are the same as when mild steel electrodes are used ; (2) when making butt welds on 1-in. mild steel plate at 72° F. the cooling rates are the same regardless of whether low-carbon steel or alloy-steel electrodes are used ; (3) the cooling rates for both bare and coated electrodes are essentially the same ; (4) the cooling rate for the first pass is slower than for the last pass when making V welds in $\frac{1}{2}$-in. plate ; (5) about 45% more metal is deposited by austenitic than by ferritic electrodes for the same energy input ; and (6) last-pass butt welds cool more rapidly than first-pass butt welds, whilst first-pass fillet welds cool more rapidly than second-pass fillet welds.

The Spot Welding of 0·0375-In. Aluminized Low-Carbon Steel. H. W. Brown. (Welding Journal, 1944, vol. 23, Sept., pp. 458-S–473-S). A comprehensive report is presented of an investigation of the strength of spot welds in aluminium-coated steel sheet. The material used was a low-carbon steel sheet (carbon 0·10%) coated with aluminium on both sides, the coating thickness being 0·0013–0·0020 in. The welds were made with a 75-kVA. A.-C. spot welding machine. The effects of variations in the welding conditions are pointed out.

The Copper-Brazing of Steel Assemblies. J. D. Jevons. (Overseas Engineer, 1944, vol. 18, Nov., pp. 28–31, 50). A description is given of the copper-brazing process. Under industrial conditions it consists of assembling steel components with definite clearances and in such a manner that they remain in the desired position until the joint is made ; pure copper in the form of wire, foil or powder (made up as a paste) is placed in a position from which, when molten, it will flow readily into the spaces to be filled ; the assembly is heated and held for 5–10 min. at about 1120° C. in a reducing atmosphere and then cooled, usually in the same atmosphere, by passing through a tunnel furnace with a cooling zone.

Stresses in Welded Structures. H. C. Boardman. (Year Book of the American Iron and Steel Institute, 1944, pp. 103–114). The problem of stresses in welded structures is dealt with by considering the types and magnitudes of the stresses which would be set up by butt-welding two plates measuring 13 × 6 × 1 in. under various conditions, such as with and without preheating and with and without cold deformation of the plates in the elastic and plastic ranges.

Low-Frequency Induction Heating. O. Kreisel. (Welding Journal, 1944, vol. 23, Aug., pp. 710–712). The application of heavy low-frequency currents for the local heating by induction of heavy plates and pipes prior to welding and for normalising after welding is discussed.

Controlled Low Temperature Stress Relief. T. W. Greene and A. A. Holzbaur. (Welding Journal, 1944, vol. 23, Aug., pp. 369-S–371-S). The stress-relieving of long welds in large structures such as ships is discussed, and a method is advocated which consists of simultaneously heating two bands of the steel, one on each side of the weld, with special oxy-acetylene torches. Tests have proved that the temperature required to attain adequate stress relief is not more than 400° F.

Stud-Link Anchor Chain. (Canadian Metals and Metallurgical Industries, 1944, vol. 7, Oct., pp. 38–39). The manufacture of 2⅛-in. manganese steel stud-link anchor chain at a works in Vancouver is described. The bar-stock is cut in an oxy-acetylene cutting machine into 25-in. lengths for each link. These are heated and formed in a bending machine at a rate of 100 links per hr. The links are next welded in an automatic resistance welding machine with electronic control, the welding time per link being 25–35 sec. After burning and grinding off slag and surplus metal, the chains are annealed and tested.

Welded Structures *vs.* Meehanite Castings. E. J. Charlton. (Iron Age, 1944, vol. 154, Sept. 21, pp. 74–77, 168–173). In an earlier paper L. F. Williams stressed the advantages of Meehanite castings as compared with welded steel for diesel-engine components (*see* Journ I. and S.I., 1944, No. II., p. 122 A). In the present paper the advantages of welded components are put forward and it is shown that a direct-cost comparison of welded and cast parts cannot be made.

The Oxy-Acetylene Process in the Steel Mill. (Blast Furnace and Steel Plant, 1944, vol. 32, Sept., pp. 1076–1079). Some applications of the oxy-acetylene torch and lance in the steelworks are described and illustrated.

CLEANING AND PICKLING OF METALS

(Continued from p. 24 A)

Modern Metal Cleaning Processes. (Metal Treatment, 1944, vol. 11, Autumn Issue, pp. 179–188). A comprehensive review is given of up-to-date methods for cleaning ferrous and non-ferrous metals. Problems encountered in every-day production are discussed from a practical point of view. The cleaning of metal in bulk is given special consideration.

Precleaning Low Carbon Steel for Spot Welding. J. Gauthier. (Iron Age, 1944, vol. 154, Oct. 5, pp. 66–67). A simple procedure is outlined for the removal of oil, dirt and oxide from low-carbon steel sheets, in order to ensure structurally sound welds and a surface to which paint or plating readily adheres. The method is similar to that used for the precleaning of aluminium prior to spot-welding.

Descaling of Hot Steel before and during Rolling and Forging. G. Wallquist. (Iron and Steel Institute, 1944, Translation Series, No. 202).

An English translation is presented of a paper which appeared in Jernkontorets Annaler, 1943, vol. 127, No. 3, pp. 61–89 (*see* Journ. I. and S.I., 1944, No. I., p. 21 A).

Pipe Cleaning. L. R. Robinson. (Iron and Steel Engineer, 1944, vol. 21, Sept., pp. 46–50). Appliances are described for cleaning deposits from the inside of blast-furnace-hearth coolers, boiler tubes, condenser tubes, chemical and water pipes, sewers, drains and gas pipe-lines.

Cleaninghouse Practice. F. P. Spruance. (Wire and Wire Products, 1944, vol. 19, Oct., pp. 671–675). Pickling practice is reviewed and some micrographs of wire sections are presented which demonstrate the effect of pickling alloy steel and mild steel wire in 5% (by volume) sulphuric acid with and without additions of 20 lb. of rock salt to each 100 gal. of acid and with and without additons of an inhibitor called "Rodine."

The Latest United Semi-Continuous Pickler. C. F. Buente. (Blast Furnace and Steel Plant, 1944, vol. 32, Sept., pp. 1068–1070; United Effort, 1944, vol. 24, Sept., pp. 11–15).

Semicontinuous Strip Pickler Handles Individual Coils. C. F. Buente. (Steel, 1944, vol. 115, Sept. 25, pp., 90–92). A semicontinuous pickling plant for cleaning hot-rolled strip prior to cold-rolling is described. The coils are dealt with separately, so that no automatic welders and looping pits are required. The leading end of the coil is picked up by rubber-covered pinch rolls and fed into a leveller unit, which in turn feeds it into the first tank. This tank contains hot water, which removes broken scale and raises the strip temperature. As the strip is hot when it enters the pickling tank, the action with the acid is accelerated.

COATING OF METALS

(Continued from pp. 25 A–28 A)

Factors Underlying Specifications for Electrodeposited Metallic Coatings. E. M. Baker. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 191–194). The relative merits of using data from minimum thickness determinations and from salt-spray tests in specifications for metal coatings on steel are considered.

Chromium Plate in Engineering Applications—Its Thickness and Finishing. T. G. Coyle. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 556–561). Engineering applications of chromium plating are discussed, and recommendations are made regarding the optimum plating thickness for various tools, gauges, moulds, drawing dies and plugs.

Aluminimum as an Addition to Zinc Baths for Galvanising. H. Bablik and F. Götzl. (Archiv für das Eisenhüttenwesen, 1943, vol. 17, Nov.–Dec., pp. 151–152). The addition of small amounts of aluminium to molten zinc for galvanising is discussed with particular reference to the reactions which take place and the use of the correct flux to bring about the desired effect without wasting aluminium.

Galvanising Fluxes. A. T. Baldwin. (Hot-Dip Galvanisers' Association: Iron and Steel, 1944, vol. 17, Oct., pp. 645–646). The use of hydrochloric-acid and zinc-ammonium-chloride flux washes prior to hot-dip galvanising is discussed. With the latter flux there is a considerable reduction in the amount of dross formed. Iron accumulates as a sludge in zinc-ammonium-chloride solution, and when the amount reaches about $\frac{3}{4}$ oz. per gal. it should be removed by raising the *p*H value to about 5·75 and adding hydrogen peroxide.

Corrosion-Resistance of Steel is Greatly Extended. R. M. Thomas. (Steel, 1944, vol. 115, Oct. 2, pp. 74–78, 118–124). The principles applied in the electrolytic coating of steel with zinc are explained and the subsequent treatment of zinc and cadmium plated surfaces in a chromate bath is described. Brief immersion in the chromate bath followed by rinsing and drying leaves a uniform, opaque, olive-green coating which is extremely resistant to salt spray and atmospheric corrosion. It is possible to treat the film before drying in such a way as to produce a coloured precipitate within the film.

A Reappraisal of Electrogalvanizing. E. H. Lyons, jun. (Wire and Wire Products, 1944, vol. 19, Oct., pp. 646–648, 732, 733). The advantages and disadvantages of the electrolytic zinc-plating process for wire are reviewed. The coating is much more uniform in thickness than with the hot-dip process, and the drag-out losses of zinc are less. The rubber lining of the tank sometimes contaminates the electrolyte, and may cause the coating to be rough. Zinc acts as a good die lubricant, and electrolytically coated wire can be drawn to fine gauges at high speeds with little wear on the die.

Tinning Cast Iron for Babbitted Bearings. T. E. Eagan. (Metals and Alloys, 1944, vol. 20, Sept., pp. 625–629). A description is given of the Kolene process of tinning cast iron so as to obtain a bond for babbitt metal in the manufacture of bearings. The graphitic carbon in the iron prevents the tin from bonding, but in the Kolene process the part is immersed in a molten salt bath (at 700–950° F.) containing a catalyst to promote oxidation of the carbon to CO_2 which combines with constituents of the bath to form carbonates. The part is then rinsed, put into a second alkali bath at 250° F., rinsed again, dipped about half a minute in 20–30% hydrochloric acid, passed into a flux bath and finally to the tinning bath.

Lead Plating. J. L. Bray. (Steel, 1944, vol. 115, Oct. 9, pp. 128–129, 288–292). The process of electroplating lead on steel is described with particulars of the fluosilicate and the fluoborate solutions. The fluosilicate bath is cheaper for large-scale operations. The fluoborate bath yields a satisfactory deposit directly on to steel, and the coating has a finer grain and is denser than that obtained with fluosilicate solutions.

Properties of Plated Lead Coatings on Steel. K. G. Soderberg. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 562–568). The literature on the properties, particularly the corrosion resistance, of solid lead and of lead-coated steel is reviewed and the results of some of the author's own tests are reported.

Chemically Treated Steels for Food Cans. C. M. Cosman. (Iron Age, 1944, vol. 154, Sept. 28, pp. 54–58). The German technical literature on the use of phosphate and lacquer coatings on steel sheet for food cans as a means of replacing tinplate is reviewed.

Russian Practice with Protective Phosphate Coatings. (Sheet Metal Industries, 1944, vol. 20, Nov., p. 1918). Particulars are given of a phosphatising process widely used in Russia, in which manganese phosphate is employed. The bath is a 3% solution of $Mn(H_2PO_4)_2$ and $Fe(H_2PO_4)_2$, the ratio of manganese to iron being between 6 and 9. The results of corrosion tests on steel phosphatised by this process are presented.

Capillarity of Metallic Surfaces. E. R. Parker and R. Smoluchowski. (American Society for Metals, Oct., 1944, Preprint No. 13). Observations made on the spreading of liquid metals on solid metals are discussed, considerations being limited to the notion of surface energy and to the contour of the solid metal surface. Tests are described in which the rate of spread of liquid silver on steel at 1200° C. was observed. The cleanliness of a metal surface is not the only factor in the spreading, for the presence

of capillary grooves, machined or etched, and of grain boundaries greatly increases the rate of spread. The influence of grinding, polishing and etching was also studied.

Electrostatic Spraying and Detearing. H. Forsberg. (Iron Age, 1944, vol. 154, Oct. 19, pp. 50–54). Electrostatic methods of spraying small parts with paint and of removing excess paint after dipping and before baking are described. For applying the paint the part is suspended between vertical electrodes charged at 100,000 V. Both paint and time are saved, and the coating is very uniform. The process of removing paint drops (called "detearing") has been described by H. P. Ransburg (*see* Journ. I. and S.I., 1944, No. I., p. 100 A).

PROPERTIES AND TESTS

(Continued from pp. 54 A–62 A)

Mechanical Properties of Case-Hardened Materials. S. L. Gertsman and I. H. MacPherson. (Canadian Metals and Metallurgical Industries, 1944, vol. 7, Oct., pp. 32–33). Two methods of testing developed by the Bureau of Mines, Ottawa, for case-hardened bars are described. The specimens were $10\frac{1}{4}$ in. long $\times$ $\frac{7}{16}$ in. dia. In the bend test the bar was freely supported on 8-in. centres, and pressure was applied with a 12-in. radius bend block. The first crack in the case was accompanied by a clear and distinct sound, and the deflection in inches when this sound was heard was called the "case break point." The second test consisted of allowing a 45 ft.-lb. blow to strike the bar midway between the two supports which were also 8 in. apart. These tests provided evidence which enabled improvements in the methods of production to be made.

The True Stress-Strain Tension Test—Its Rôle in Modern Materials Testing. C. W. MacGregor. (Journal of the Franklin Institute, 1944, vol. 238, Aug., pp. 111–135; Sept., pp. 159–176). The fundamentals of the true stress-strain tension test are discussed. Its use in revealing the effects of the testing conditions, such as high and low temperatures, velocity of testing and size and shape of the specimen, are dealt with in Part I., and the effects of heat treatment, alloying additions, cold-work, overheating and grain direction are considered in Part II.

The Effect of the Rate of Straining on the Strength of Cast Iron. F. Bischof. (Giesserei, 1944, vol. 31, Sept., pp. 141–142). An investigation is described in which specimens of two cast irons were fractured in a tensile machine, the load being applied at three different rates. In both cases the tensile strength increased with increasing rate of straining.

The Technical Cohesive Strength and Other Mechanical Properties of Metals at Low Temperatures. D. J. McAdam, jun., and R. W. Mebs. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 661–703). Diagrams representing the technical cohesive strength of metals at room temperature and at selected low temperatures are presented and discussed. The influence of low temperatures on the form of the diagram afforded a basis for interpreting the effect of low temperature on the strength, ductility and total work of notched and unnotched specimens. A study was also made of the quantitative variation of the technical cohesive strength and other mechanical properties. This variation was examined with reference to the usual arithmetical scale of temperatures, as well as with reference to a temperature scale of equal thermodynamic efficiencies. The use of the latter scale revealed relationships of considerable theoretical and practical importance.

Improving the Impact Stress Endurance of a Carburized Gun Part. K. B. Valentine. (Metal Progress, 1944, vol. 46, Sept., pp. 467–472). An investigation is described the purpose of which was to select a steel and a suitable hardening treatment which would increase the life under repeated impacts of a hammer for a small automatic gun. A low-alloy nickel-chromium-molybdenum steel was selected; this was carburised, held in a salt bath at 480° F. for 2 hr. and then shot-peened. Subsequent surface treatment by plating with chromium, cadmium, tin or lead, or by blueing at 450° F. decreased the endurance, but phosphate coating had no detrimental effect.

Relationships between the Magnitude of the Notched-Bar Toughness and the Type of Fracture in the Bend Test on Weld Metal. K. Albers. (Zeitschrift des Vereines Deutscher Ingenieure, 1943, vol. 87, Oct. 30, pp. 677–682). The value of the notched-bar impact test and of the bend test on weld metal as methods of determining the weldability of structural steels is discussed. Impact tests were carried out on specimens which had been subjected to various degress of cold work, but no relationship was found between the impact values and the type of fracture obtained in the bend test on weld metal.

On the Transition from a Ductile to a Brittle Type of Fracture in Several Low-Alloy Steels. P. G. Jones. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 547–553). The results of Charpy impact tests on low-alloy steels containing various combinations of copper, nickel and chromium are reported, the object being to study the conditions tending to cause brittle fracture. For a given hot-rolled ductile steel these conditions were thought to be a high rate of straining, a non-uniform distribution of stress, a triaxial state of stress and a low temperature. The suddenness of transition from ductile to brittle fracture varied considerably with the steels tested.

A Test for Measuring Drawability of Deep-Drawing Steels. F. W. Boulger and F. B. Dahle. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 503–518). The reasons for the failure of present standard tests to evaluate the suitability of sheet metals for deep-drawing operations are discussed, and an experimental method of testing is proposed. This is a cup-drawing test which has been found satisfactory. Test data are presented which indicate that whilst increases in carbon and sulphur in the steel decrease the drawability, variations in the manganese, copper and phosphorus within certain limits do not affect this property.

Hyperbolic Sine Chart for Estimating Working Stresses of Alloys at Elevated Temperatures. A. Nadai and P. G. McVetty. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 735–745). A method of plotting results of creep tests in a manner which departs from the usual double-logarithmic scales of stress and minimum creep rate is described. The method is based upon the hyperbolic sine function for expressing the dependance of the rate of flow on the stress. The new chart enables available creep-test data to be surveyed in an easily comparable form. Examples illustrate its application to a variety of materials to obtain estimates of stress for low specific creep rates.

Fluorescent Penetrant Inspection. G. Ellis. (Steel, 1944, vol. 115, Oct. 16, pp. 100–102, 164). A method of crack detection is described in which the part is sprayed or brushed with a fluorescent solution which penetrates into cracks and pores; surplus solution is wiped off and the part is dried and inspected in a dark room under " black " light which causes any cracks or defects to be shown up by yellowish-green lines.

The Effect of Overstressing and Understressing in Fatigue. J. B. Kommers. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 749–762). Fatigue tests were carried out on smooth and

square-notched specimens of annealed ingot iron and a 0·32%-carbon steel, and on smooth specimens of cold-drawn ingot iron. In the over-stressing tests specimens were first subjected to stresses 10%, 20%, or 30% above the endurance limit, after which the load was reduced so as to determine the new endurance limit. In the understressing tests the increased endurance limit was determined after increasing numbers of cycles at stresses below the endurance limit. Curves showing the decreases and increases in the endurance limit brought about by overstressing and under-stressing respectively are presented. A reduction in the original endurance limit was obtained by comparatively few cycles of overstress, but a considerable number of cycles of understress was necessary to obtain an increase in the endurance limit.

Second Progress Report on the Effect of Size of Specimen on Fatigue Strength of Three Types of Steel. H. F. Moore and D. Morkovin. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 109–120). Fatigue tests were carried out on notched specimens of two plain carbon steels (carbon 0·20% and 0·35%) and on one heat-treated chromium-molybdenum steel, and the results were correlated with the results of tests on unnotched specimens reported previously (*see* Journ. I. and S.I., 1943, No. II., p. 163 A). A tendency for the endurance limit of the steels tested to become constant for notched specimens 1 in. in dia. and larger was noted. A diagram is presented which enables an approximate estimate of the endurance limit of members of any size to be made when the endurance limit of a geometrically similar specimen of the same metal but of different size is known.

Some Notes on the Indentation Hardness Test. J. M. Lessells. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 803–814). The relationship between the indentation hardness and other properties of steel are discussed. Whilst the hardness and tensile strength, as well as the tensile strength and endurance limit, are related in the case of most heat-treated steels, serious errors can arise from the assumption that the hardness is related to the endurance limit. The ratio of tensile to fatigue strength is considerably reduced by drastic quenching, owing to the internal stresses set up, but the indentation hardness test does not detect such a condition.

Present Types of Hardness Tests. S. R. Williams. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 815–834). Descriptions are given of Bierbaum's scratch hardness tester and a number of indentation hardness testing instruments.

The Sphere of Application for Microhardness Testing. E. B. Bergsman. (Teknisk Tidskrift, 1944, vol. 74, Nov. 11, pp. 1297–1303). (In Swedish). The applications of scratch and indentation microhardness testing machines are reviewed. It is shown that the ratio of the number of cracks round the impression to the impression area or depth can be used as a measure of the brittleness of the material tested.

The Development of Controlled Hardenability. W. G. Bischoff. (Steel, 1944, vol. 115, Oct. 16, pp. 98–99, 139–140). Data are presented showing the difference between the results of Jominy end-quench tests on specimens cast in the melting shop and on specimens of the same steel after forging and machining. The practicability of including minimum and maximum hardness values at a given distance from the quenched end in steel specifications is considered.

Some Aspects of Commercial Production of Alloy Steels to Hardenability Requirements. W. G. Bischoff. (Year Book of the American Iron and Steel Institute, 1944, pp. 132–140). *See* p. 33 A.

Principles Involved in Determining Hardenability Limits of Alloy Steels. L. L. Ferrall. (Year Book of the American Iron and Steel Institute, 1944,

pp. 141–160). The work of the Alloy Technical Committee of the American Iron and Steel Institute on the determination of hardenability bands for inclusion in steel specifications is outlined. A hardenability band is the area between two curves drawn from end-quench test results; the two curves represent the high and low hardness values obtained from a number of heats representative of the normal manufacturing practice for a particular grade of alloy steel.

Selection of Automotive Steel on the Basis of Hardenability. A. L. Boegehold. (S.A.E. Journal, 1944, vol. 52, Oct., pp. 472–485). The application of end-quench test data and cooling rate curves to determine the choice of steels suitable for making heat-treated automobile parts is discussed with examples of recommended procedures.

The Effect of Phosphorus on the Hardness of Quenched Cast Iron. F. Bischof. (Giesserei, 1944, vol. 31, Apr., pp. 53–55). An investigation is described in which specimens of five cast irons containing 0·12%, 0·35%, 0·52%, 0·91% and 1·18% of phosphorus were quenched in water and in oil after holding at temperatures of 700°, 800° and 900° C. for periods up to 4 hr. After quenching from below the A_1 point, the hardness was less than that in the as-cast state, owing to the formation of pearlite; increasing the holding time did not affect the hardness in this case. At temperatures of 800° and 900° C. increasing the holding time decreased the hardness after quenching. For temperatures of 900° C. the highest hardness was obtained with the iron containing 0·52% of phosphorus. The reduction in hardness after quenching was due to the phosphorus promoting the precipitation of carbon from solid solution.

Interpretation of Creep Test Data. P. G. McVetty. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 707–727). The interpretation of creep-test data is based upon the relations between temperature, stress, deformation and time, and the effects of stress and temperature upon the deformation/time ratio are considered. Interpolation and extrapolation methods are developed for the presentation in tabular form of creep results covering any desired ranges of stress and temperature. The advantage claimed for this method of classifying data is that it permits of reasonably accurate analyses based upon a minimum number of tests.

The Deformation and Strength of Metallic Materials under Creep Stress. A. Thum and K. Richard. (Zeitschrift des Vereines Deutscher Ingenieure, 1943, vol. 87, Aug. 21, pp. 513–520). The literature on the short-time and long-time creep testing of metals is reviewed.

Thermodynamic Properties of Carbides of Chromium. K. K. Kelley, F. S. Boericke, G. E. Moore, E. H. Huffman and W. M. Bangert. (United States Bureau of Mines, 1944, Technical Paper No. 662). The preparation of the carbides of chromium is described, and the results of low and high temperature specific heat measurements are given; the data are correlated satisfactorily, and thermodynamic values and functions for the carbides are derived.

Chemical and Mechanical Properties of Some of the National Emergency Steels. W. C. Stewart and R. E. Wiley. (Journal of the American Society of Naval Engineers, 1944, vol. 56, Aug., pp. 396–411). The results of tests on 36 steels representing 10 grades of National Emergency steels are presented. The primary purpose of the investigation was to determine their suitability for making large forgings for propeller shafting and rudder stocks.

Directory of Metals—Materials for Chemical Engineering Equipment. (Chemical and Metallurgical Engineering, 1944, vol. 51, Sept., pp. 101–106). A long alphabetical list of most of the metallic materials used in the construction of process industry equipment is presented. The name of

each material is followed by the name and address of the manufacturer, the nominal chemical composition and the forms in which it is available.

Development and Application of Military and Special Steels for Ordnance Purposes. J. H. Frye. (Year Book of the American Iron and Steel Institute, 1944, pp. 66–86). Data are presented on the production of steel in the United States during 1942 and 1943 and its allocation to meet military, naval and mercantile marine requirements. The reductions in the consumption of nickel, chromium and molybdenum required for certain weapons are pointed out.

Development of Special Steels for Naval Use. E. G. Touceda. (Year Book of the American Iron and Steel Institute, 1944, pp. 87–94). *See* p. 32 A.

The Physical Properties of High Carbon Steel Rope Wire as Affected by Variations in Patenting. H. J. Godfrey. (Wire and Wire Products, 1944, vol. 19, Oct., pp. 635–642). An investigation of the effect of patenting in lead at 900°, 1000° and 1100° F. on the tensile properties of wire for wire ropes is reported. Three steels containing 0·75–0·80% of carbon were used; these were hot-rolled to 0·187 in. in dia., patented and drawn to 0·083, 0·083 and 0·096 in. respectively, for final patenting. In the tensile test the reductions in area were approximately the same after patenting at 1000° and at 1100° F., and were higher after patenting at 900° F. The tensile strengths were approximately the same after patenting at 900° and 1000° F., and were reduced by patenting at 1100° F.

New Developments in Die Materials. R. Bredenbeck. (Iron Age, 1944, vol. 154, Oct. 12, pp. 60–66). The chemical analyses, heat-treatment, hardnesses, Charpy impact values and results of friction tests on a large number of die steels are correlated in tabular form. As a result it has been possible to select, for certain operations, die steels having a much longer life than the normal.

CORROSION OF IRON AND STEEL

(Continued from pp. 66 A–68 A)

Metallic Corrosion. U. R. Evans. (Midland Metallurgical Societies: Iron and Steel, 1944, vol. 17, Nov., pp. 686–690). Problems of the corrosion of iron and steel now being studied at the Corrosion Research Section of Cambridge University are briefly reviewed under the following headings: (1) The protection of steel surfaces which have reached a condition in which ordinary painting methods fail to give protection; (2) the evolution of a safe and efficient inhibitive system for rendering cooling-waters non-corrosive; and (3) the avoidance of cracking due to corrosion-fatigue or to various forms of embrittlement.

Corrosion Ratings of Metals. (Iron Age, 1944, vol. 154, Oct. 5, pp. 59–62). Classified data originally prepared by the Westinghouse Electric and Manufacturing Company on the corrosion resistance under various conditions of several ferrous and non-ferrous alloys are given. Corrosion ratings are based upon the general deterioration of the specimen, including local pitting and intergranular corrosion, in various outdoor and industrial atmospheres. The alloys are classified into five different degrees of corrosion resistance in rural, urban and marine atmospheres, and in industrial atmospheres contaminated with ammonia, hydrogen sulphide, hydrogen chloride, sulphur dioxide and chlorine.

The Total-Immersion Corrosion Test. W. A. Wesley. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 649–658). Series of total-immersion corrosion tests were made, and the results are

presented by tables and curves in such a way as to show the reproducibility of the results. The ferrous metal tests consisted in measuring the rates of corrosion of 24 specimens selected at random from 80 specimens cut from a sheet of mild steel. Each of these was immersed in a separate jar containing 3% sodium-chloride solution through which air was bubbled at a high rate. A statistical analysis of the test results is made.

Corrosion Experiences in Chemical Process Industries. (Chemical and Metallurgical Engineering, 1944, vol. 51, Sept., pp. 94–100). The views of engineers representing many chemical process industries on corrosion problems are summarised, and an extensive table is presented which contains recommendations as to the best material to use for constructing various items of plant to resist the attack of any one of a long list of chemicals.

Some Cases of Corrosion in Engineering Practice. G. H. Stanley. (Journal of the South African Institution of Engineers, 1944, vol. 42, Apr.–May, pp. 135–163). Several cases of failure in service due to the corrosion of steel and wrought-iron pipes and of non-ferrous metal, as well as some experiments to ascertain the cause of the corrosion, are described.

Corrosion of Steel Salt-Bath Pots by Molten Alkali Nitrates. R. J. Box and B. A. Middleton. (Iron and Steel Institute, 1945, this Journal, Section I). The corrosion of steel containers of salt-bath furnaces by molten alkali nitrates was kept under observation simultaneously in several factories over two or three years. The furnaces were used for heat treatment of aluminium alloys at about 380° and 495° C., with variations in (1) the quality of steel used for the pot, (2) the shape of the pot and (3) the type of salt employed. Details of analytical control found useful are given. Results obtained indicated that with commercially available salt of guaranteed purity (sodium chloride 0·3% max.) corrosion was negligible at 380° C., and proceeded only slowly at 495° C. Where rapid attack took place the furnace design appeared to be a fairly prominent factor. For all practical purposes the use of ingot iron for the pot was a complete insurance against corrosion under the conditions described. The rate of increase of alkalinity of the melt was an index of the rapidity with which corrosion was taking place.

Studies on Stress Corrosion Cracking of Austenitic Stainless Steels Types 347 and 316. M. A. Scheil and R. A. Huseby. (Welding Journal, 1944, vol. 23, Aug., pp. 361-S–363-S). Some tests are described the object of which was to ascertain whether 18/8 stainless steels containing ferrite are more resistant to cracking in boiling calcium-chloride solution than the wholly austenitic alloys. A stainless steel containing 25–33% of ferrite did not crack when stressed and immersed in a boiling 42% $CaCl_2$ solution, whereas the specimens with no ferrite did show cracks. In $MgCl_2$ solutions boiling at 154° C., specimen of both types of stainless steel cracked in about the same time.

Corrosion Resistance of Metallic Materials of Construction. (Chemical and Metallurgical Engineering, 1944, vol. 51, Sept., pp. 107–130). A list is presented of 80 chemicals arranged in alphabetical order; under each one are given the metals and alloys which resist attack by the chemical, with information on the rates of corrosion expressed in inches of penetration per year at stated temperatures.

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ORES—MINING AND TREATMENT

(Continued from pp. 39 A–40 A)

Methods of Preparing Ore Slimes. F. Hartmann. (Stahl und Eisen, 1944, vol. 64, Sept. 21, pp. 613–618). The mechanism of the process of settling ore slimes in stationary liquids was investigated on the basis of Stokes' law. Thick slimes settled more slowly than thin ones. The amounts of solids in the slime phase of sludges containing varying amounts of solids were at first approximately the same, but with increasing settling time the thicker sludges formed thicker slimes, so that the settling ponds for thicker sludges need not be so large. In thickeners, a slow stirring action accelerated settling only when more than 150 g. of solids per litre were present; with less than about 100 g. per litre it had no effect. It took a considerable time for the slime phase to become denser in a thickener when stirring was applied. Complete clarification of ore sludges could be obtained with centrifuges even when they were high in solids. Small additions of lime to the water after centrifuging assisted greatly in the rapid precipitation of the remaining solids, but additions to the sludge entering the centrifuge had no beneficial effect. It was found that by centrifuging an ore slime containing 15% of manganese the resulting slime with 73·4% of the initial total solids contained only 8·6% of manganese, thus the recovered precipitate contained 30% of manganese.

Magnetic Separation. J. E. Hyler. (Steel, 1944, vol. 115, Sept. 25, pp. 95–96, 134). Descriptions are given of magnetic separators, mainly of the revolving magnetic-drum type, which are used for separating ores, and for removing ferrous particles from non-ferrous scrap metals, turnings, borings, &c.

The Standardisation of Ore Crushing, Screening and Sintering Plant Units. J. Klütsch. (Stahl und Eisen, 1944, vol. 64, Oct. 26, pp. 675–679). Descriptions are given of some combined ore-crushing, screening and sintering plants which are in course of construction in Germany. To meet war-time conditions the design excludes the use of rubber conveyor belts and additional rolling stock; existing ore bunkers had to be incorporated as far as possible, and the plants were to be capable of rapid erection using the least possible amount of alloy steels.

New Sintering Plant Affords Close Control of Product. J. D. Knox. (Steel, 1944, vol. 115, Nov. 20, pp. 120–123, 166). A description is given of the recently completed sintering plant at the works of the Wheeling Steel Corporation, Steubenville, Ohio. The whole plant is 700 ft. long, and the sintering machine, which has 14 wind boxes, is 89 ft. long and 6 ft. wide. Production has been at the rate of 1200 tons per day of two 8-hr. shifts.

REFRACTORY MATERIALS

(Continued from pp. 70 A–72 A)

The Importance of Refractories. J. H. Hartley. (Refractories Journal, 1944, vol. 20, Aug., pp. 335–346; Oct., pp. 435–440; Nov., pp. 481–491). The units by which heat quantity and temperature are measured, and the term "specific heat" are explained so that comparisons of the properties of refractories can be made. The history of the development of refractory materials is outlined with notes on the work of Wedgwood, Seger and

Mellor, and tables of the analyses and properties of sands, ganisters, diatomaceous earths and many types of refractory bricks are presented.

The Behaviour of Refractory Materials under Stress at High Temperatures. F. H. Clews, H. M. Richardson and A. T. Green. (Transactions of the British Ceramic Society, 1944, vol. 43, Nov., pp. 223–246). In Part I. of this paper experiments on the deformation of porcelain under tensile and torsional stresses at 900–1100° C. are described. In Part II. the results of applying compressive loads of 1, 5·5, 10·3 and 17·9 lb. per sq. in. on refractory insulating material at 1100°, 1150° and 1200° C. are reported. The tests were continued for periods up to 250 hr. In each case the rate of subsidence was at first relatively rapid, but decreased continuously with time. Preheating the material at 1330° C. for 2 hr. considerably decreased the rate of subsidence under load.

Some Considerations on the Manufacture of Refractory Insulating Products. J. F. Clements, L. R. Barrett and A. T. Green. (Transactions of the British Ceramic Society, 1944, vol. 43, Oct., pp. 187–222). A number of methods of manufacturing insulating bricks are reviewed; these include: (1) The incorporation in the mix of natural or synthetic lightweight material such as diatomite, expanded vermiculite or raw kyanite; (2) the inclusion of materials which are later removed by combustion or sublimation, of which the former is much the more important in terms of output; (3) the production of a cellular structure by mechanical beating using a frothing agent; (4) the formation of gas bubbles in a semi-fluid medium by chemical reaction; and (5) miscellaneous methods. The comparatively small amount of information in the literature relating to methods (1) and (2) is discussed in the light of modern developments in other branches of refractories manufacture. Special attention is given to the possibility of increasing the plasticity of clay-combustible mixtures by special selection of the clay, chemical or other treatment, or by the addition of a plasticising agent. A method which has been attracting much attention is that of frothing a clay slip, to which is usually added a quantity of fine grog, followed by casting, drying and firing in the normal manner. This has been described by Russian workers, whose accounts have been drawn on. Reasons are given for supposing that further improvements could be made. The process is particularly adapted to making very lightweight bricks, impossible by the combustible method. Attention is drawn to the work of G. A. Bole and his co-workers on the chemical bloating process, the products of which are comparable in properties to those made by the frothing process. An attempt is made to compare the properties of bricks made by the different processes. A bibliography with 140 references is appended.

The Practical Aspects of Reheating and Heat-Treatment Furnace Insulation. A. Stirling. (Journal of the Institute of Fuel, 1944, vol. 18, Dec., pp. 39–50). Furnace insulation is comprehensively dealt with under the following headings: (1) The materials of insulation; (2) the progress of insulation; (3) the furnace structural factors influencing insulation; (4) the furnace operating factors influencing insulation; and (5) the economics of furnace insulation. Much of the present conservatism towards insulation in reheating and heat-treatment practice has no technical justification. Insulation can reduce furnace total heat losses to a minimum only after detailed consideration of the structural and operating factors. The aim of insulation is attainable only where the difference between capital cost and economic cost is defined.

FUEL

(Continued from pp. 72 A–73 A)

The Abnormality of Flame Gases. W. T. David. (Proceedings of the Institution of Mechanical Engineers, 1944, vol. 151, No. 3, pp. 236–245). Experiments are described which show that a long-lived latent energy exists in flame gases, amounting in certain cases to as much as 15% of the heat of combustion of the unburnt inflammable mixture. The latent energy appears to reside in a proportion of the molecules formed during combustion. These molecules are probably of abnormal structure, and hold within them an energy far greater than that held by normal molecules. In virtue of this, an abnormally large dissociation obtains in flame gases. The experiments described indicate that the magnitude of the long-lived latent energy depends on various factors, among which are: (1) The composition of inflammable mixtures before combustion; (2) the pressure at which the inflammable mixture undergoes combustion; and (3) the mode by which the flame front is propagated through the inflammable mixture.

Coal Preparation. G. Mullin. (Iron and Coal Trades Review, 1944, vol. 149, Dec. 29, pp. 982–987). A detailed description is given of the Barvoys process of coal separation. In this process the separation of raw coal is carried out in a suitably shaped bath filled with a suspension of finely-ground barytes and clay in water.

The Coking Industry in Western Europe. (Coke and Smokeless-Fuel Age, 1944, vol. 6, Nov., pp. 205–206). Brief particulars are given of the leading characteristics of the French, Belgian and Dutch coking industries.

Assay of Coal for Carbonisation Purposes. J. Brown. (Society of Chemical Industry and the Institute of Fuel, Joint Meeting: Institute of Fuel War Time Bulletin, 1944, Dec., pp. 41–49). The variety of plant for the carbonisation of coal is reviewed and a description is given of the changing composition of the carbonisation products as the carbonising temperature is increased. Distinction is made between "differential" and "constant" cracking of the volatile matter. The development of carbonisation assays is given in chronological order with a brief description of the JB/UAL assay. A new constant for coal representing the yield in therms in the gas for each 22·4 lb. of volatile matter is introduced; this is referred to as T/V.

Gas Producers and Producer Gas. L. L. Swift. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. 45–48). The decline and subsequent increase in the development and use of gas producers in the United States are reviewed.

PRODUCTION OF IRON

(Continued from pp. 73 A–74 A)

Magnitogorsk at War. V. Dudadevsky. (Iron and Steel, 1944, vol. 17, Dec. 7, pp. 750–752). A brief account is given of the extensions which have been made to the Russian iron and steel works at Magnitogorsk, with illustrations of No. 6 blast-furnace from which the first iron was tapped in January, 1944. This furnace has a capacity of 1500 tons of pig-iron per day.

Blast-Furnace Moisture Control. J. J. Alexander. (Chicago District Blast Furnace and Coke Association: Blast Furnace and Steel Plant, 1944,

vol. 32, Oct., pp. 1213–1219). Experience gained in the operation of blast-furnaces by the Republic Steel Corporation with and without controlling the moisture in the blast is discussed and data on consumption and production are presented. Controlling the moisture reduced the number of casts in which the silicon content of the iron was higher than that specified. Increasing the moisture by injecting steam into the blast main increased the percentage of combined carbon in the iron.

Electrochemical and Allied Industries of the Niagara Frontier. R. B. MacMullin and W. C. Gardiner. (Electrochemical Society, Oct., 1944, Preprint No. 19). A summary of the electric power situation in the Niagara district is given, together with a brief historical account of the industries dependent on this source of power; these include the smelting of ores and the preparation of ferro-alloys.

An Experiment in Making Sponge Iron. W. E. Brown. (Engineering and Mining Journal, 1944, vol. 145, Nov., pp. 83–86). A brief account is presented of experiments in the reduction of ores containing about 50% of iron carried out at a plant in Texas by the Bureau of Mines. The ores were preheated in a chamber through which a reducing gas prepared from natural gas was passed; the sponge iron produced was drawn from this chamber without coming in contact with the atmosphere, and conveyed in closed cars to a furnace for making into wrought iron. The sponge produced contained total iron 74·4%, metallic iron 61·5%, carbon 1·55%, silica 12·86% and alumina 3·14%.

Powder Metallurgy. W. I. Pumphrey. (Automobile Engineer, 1944, vol. 34, Dec., pp. 547–548). Applications of the powder-metallurgy process to the making of tools and parts for the automobile industry are described.

FOUNDRY PRACTICE

(Continued from pp. 74 A–75 A)

Coke Properties and Cupola Melting. B. P. Mulcahy. (American Foundrymen's Association: Foundry, 1944, vol. 72, Oct., pp. 78–79, 202–208; Nov., pp. 94–97, 222–226). In the first part the reactions taking place when coke is burnt in the cupola are discussed, and details are presented of gas analyses taken at increasing distances above the tuyere level and at various positions between the wall and the centre. An outline is given of a recommended practice to produce clean metal with a minimum of oxidation at a minimum temperature of 2750° F. at the spout, and the effects of deviating from this procedure are pointed out. In the second part the manner in which the physical and chemical properties of the coke affect cupola operation are dealt with.

Automobile Castings. J. B. Duncan. (Iron and Steel, 1944, vol. 17, Dec., pp. 709–713). The production and properties of alloy cast iron for automobile castings at the works of the Midland Motor Cylinder Co., Ltd., are described, with particular reference to Chromidium for cylinder blocks and brake drums, Monikrom for camshafts and Cromol for crankshafts.

The Production and Founding of Inoculated High-Duty Cast Iron. J. L. Francis. (Institute of British Foundrymen: Foundry Trade Journal, 1944, vol. 74, Dec. 28, pp. 335–341; 1945, vol. 75, Jan. 4, pp. 7–10). After giving descriptions of the types of structure found in cast iron, the author discusses in particular that associated with high tensile strength and the technique of inoculation. A table gives the maximum permissible silicon contents for various total carbon contents prior to inoculation with 100–120 oz. of calcium silicide per ton when making different-sized cast-iron

bars. The fact that the effect of inoculation wears off if the melt stands for too long a time can be explained by diffusion of the inoculant. When irons with 2·8–3·2% of total carbon are poured into moulds their high solidification shrinkage must be allowed for by providing adequate runners and headers of the self-feeding type. Fluid shrinkage does not vary greatly with the composition of the iron. Recommendations on the positions of runners and risers, and on the pouring temperature, pouring rate and the design of castings are made.

The Electric Rocking Resistor Furnace. C. S. Johnson. (Foundry Trade Journal, 1945, vol. 75, Jan. 11, pp. 31–33). A detailed description is given of a cylindrical electric rocking resistor furnace, which has many advantages for the refining and superheating of high-duty cast irons. The furnace works on a single-phase current. The power consumed is 150–180 kW. per hr. with a current of up to 5000 amp. at a pressure which can be varied between 32 and 61 V. A heat of 1000 lb. of cold pig-iron and scrap can be heated to 1450° C. in 2 hr. using about 400 kWh. A reasonable life for a lining when being used solely for duplexing cast iron is about 1000 heats producing about 450 tons.

Chilled Iron Produced by Tellurium. J. S. Crout. (Foundry, 1944, vol. 72, Nov., pp. 89, 218–220). Methods of producing chilled iron castings by the addition of tellurium are described. The tellurium should be added just before casting. The amount required is very small and usually lies between 0·001% and 0·05%. Another way of using it is to apply a mould wash containing tellurium powder to the walls of the mould where a chilled surface is required. With a wash containing 0·01–0·05 g. of tellurium per square inch of mould surface a chill depth of $\frac{1}{16}$ to $\frac{3}{16}$ in. was obtained in castings of iron containing 3·20% of total carbon. The addition of tellurium also increases the resistance to attack by dilute sulphuric and hydrochloric acids.

Pulverized Coal Firing of Malleable Iron Annealing Kilns. L. S. Wilcoxson and D. F. Sawtelle. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 552–560). Improvements to furnaces fired with pulverised coal for producing malleable iron castings are described. In the final arrangement six car-bottom furnaces, each taking about 15 tons of castings per heat, are supplied with pulverised coal and air from a circulation system fed by a single pulverising and blowing unit. This system has proved to be both practical and economical.

Malleable Mixture Calculation and Melting Control. M. E. McKinney. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 441–453). A procedure for calculating the proportions of the available pig irons, scrap and ferro-alloy additions to use in charges for producing malleable iron is outlined and a method of controlling a 20-ton furnace fired with pulverised coal based on data provided by a flue-gas analysis indicator and a draft gauge is described.

Introductory Observations on the Rate of Solidification of Malleable Iron. B. C. Yearley, R. P. Schauss and P. A. Martin. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 483–498). The factors affecting the solidification of cast iron are discussed and tests are described in which blocks of iron suitable for malleablising, $1\frac{1}{2}$ in., 3 in. and 4 in. square and 8 in. in height, were cast in sand moulds. After periods increasing from 10 to 120 sec. the moulds were inverted and the unsolidified metal was allowed to run out; surveys were then made of the thickness of the solidified skin and thickness-time curves were constructed. These curves illustrate the effects of changes in the pouring temperature, of different sand mixtures and of chills on the rate of freezing. With the $1\frac{1}{2}$-in. blocks the freezing rate was faster with a $\frac{1}{4}$-in. thick chill than with one $\frac{3}{4}$ in. thick. No explanation is offered for this unexpected result.

First Report of the Foundry Practice Sub-Committee of the Steel Castings Research Committee. (Iron and Steel Institute, 1945, this Journal, Section I). The Foundry Practice Sub-Committee was formed to carry out practical investigations into problems encountered in the steel foundry.

Part 1 of this Report records their work on the running and risering of simple shapes used as 4-in., 6-in. and 9-in. cube castings, and castings 6 in. square by 12 in., 18 in. and 24 in. long. The effectiveness of various types of feeder heads normally used in steel foundries is illustrated by a large number of sawn sections of the experimental castings. The disadvantages entailed in using narrow heads which freeze prematurely, and the advantages of wider heads, insulated heads and the use of mould material which promotes differential solidification are illustrated. The establishment of definite principles to be applied to steel castings will help to remove some of the uncertainties which constantly confront the technical staffs of steel foundries.

The pioneer work in developing the so-called "whirl-gate" head method of feeding steel castings and recent examples of the application of this type of head and of the "atmospheric" head are described and illustrated in Part 2.

Foundry trials comparing the different types of feeding compound used on the fluid steel as it rises into the heads of steel castings are summarised in Part 3.

In Part 4 experiments designed to provide a numerical index of the tendency of different steels to hot tears in the mould are reported and attention is drawn to the fact that the degree of hot-tearing in a particular design can be minimised or often eliminated by using steel with a lower sulphur content.

In Part 5 some of the war-time activities of the Sub-Committee are set out, and some fundamental casting problems are mentioned in Part 6. With a return to more normal conditions attention will be devoted to further research under practical foundry conditions. The future progress of the steel castings industry depends on the co-operation of the foundry executives, who have already given so much of their time and experience, and on the generosity of individual manufacturers, who have carried out these practical experiments at their own expense.

Cast Steel Roll Manufacture and Application. F. H. Allison, jun. (Iron and Steel Engineer, 1944, vol. 21, Nov., pp. 59–65). An illustrated account is given of the moulding, casting, machining and heat treatment of cast steel rolls.

The Composition and Properties of Molding Sands. Part I. The Nature of the A.F.A. Clay Fraction Removed from Natural Molding Sands. L. H. Berkelhamer. (United States Bureau of Mines, Sept., 1944, Report of Investigations No. 3774). A report is presented of a comprehensive investigation of the nature, properties and effects of the clay bond in twenty-two samples of natural moulding sands representing several deposits in the United States, one in Canada and one in Scotland.

The Evaluation of Moulding Sands. W. Vaska. (Giesserei, 1944, vol. 31, Oct., pp. 155–157). The effects of the clay fraction and the sand particle size on the strength and permeability of moulding sands are discussed.

Mold Surface Properties at Elevated Temperatures. H. W. Dietert, R. L. Doelman and R. W. Bennett. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 421–439). Experiments are described the object of which was to determine the properties and behaviour of hot mould surfaces. The conclusions reached were: (1) Spalling tests furnish a practical means of determining the thermal stability of sand mixtures; (2) the confined expansion test is better than

the free expansion test; (3) materials such as silica flour, sea coal, core oil and cereal binders reduce the confined expansion; (4) additions of bentonite clay and other binders to silica sands reduce the confined expansion of the base sand; (5) lowering the flowability reduces the confined expansion; (6) when the surface of a green sand mould is heated by the molten metal its green strength is increased about ten times in the first minute, and this strength is still further increased by the addition of silica flour; and (7) when a dry sand mould is heated by the molten metal its strength is reduced during the first minute, but the dry sand develops increasing hot strength on continued heating.

Experiences with the Cement Dust Process. K. Alber. (Giessereipraxis, 1944, vol. 65, Aug., pp. 106–107). The use of cement dust in green sand moulds to eliminate the drying of moulds is described. Before inserting the cores the mould is dusted over to a thickness of about 0·5 mm., depending on the amount of moisture in the sand; the drag requires a little more of the cement dust than the cope. It is claimed that when the mould is prepared in this way, the iron is "quieter" on pouring and the castings have a smoother surface when knocked out.

Silicosis in Foundries. The treatment of men working in mines, grinding shops, brickworks and foundries where they are subjected to the risk of silicosis has received increasing attention, and in this connection three papers recently presented before the Canadian Institute of Mining and Metallurgy dealing with aluminium powder treatment are of interest. Abstracts of the papers are given below:

> **Progress in Aluminium Therapy.** W. D. Robson. (Transactions of the Canadian Institute of Mining and Metallurgy, 1944, vol. 47, pp. 172–179). A report is presented on a clinical test carried out on 34 men during the four years ending 1943. The men had been exposed to siliceous dust in mines for an average period of about 18 years. The treatment consisted of inhaling air carrying particles of aluminium (under 5 microns in dia.) for periods of up to 30 min. each day for 200 days. The conclusion was that aluminium powder treatment was followed by beneficial results in a significant proportion of cases.
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> **Aluminium Therapy in the United States.** J. W. G. Hannon. (Transactions of the Canadian Institute of Mining and Metallurgy, 1944, vol. 47, pp. 180–184). The results of aluminium powder treatment on groups of men who were exposed to silica dust in a variety of industries are reported. It was found that: (1) The treatment alleviated the symptoms of silicosis in a high percentage of cases; (2) the treatment improved the worker's health, morale and productive capacity; (3) the rapid type of silicosis was particularly responsive to aluminium therapy; (4) inhalation of aluminium powder prevents silicosis; and (5) there was no evidence of toxicity resulting from the treatment.
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> **The Engineering Aspects of Aluminium Prophylaxis.** A. W. Jacob. (Transactions of the Canadian Institute of Mining and Metallurgy, 1944, vol. 47, pp. 185–202). Particulars are given of the equipment and procedure for administering aluminium powder to workers exposed to the risk of silicosis.

Modern Machines and Equipment for Pattern-Making. R. Löwer. (Giessereipraxis, 1944, vol. 65, Sept., pp. 121–124). A number of German pattern-making machines of recent design are described and illustrated.

Patterns of Gypsum Cement. E. H. Schleede. (Foundry, 1944, vol. 72, Oct., pp. 76–77, 224–228; Nov., pp. 90, 186–196). The preparation and properties of gypsum cement are described and details are given of methods of making gypsum patterns.

The Application of Sink Heads in Grey Iron Founding. (Giessereipraxis, 1944, vol. 65, Sept., pp. 115–118). Recommendations on the use of moulds with sink heads are made and three methods of preparing such moulds for grey iron castings are described; these methods are: by drawing the pattern, by using a separate pattern for the head, and by incorporating the head in the pattern for the casting.

Radiography—Key to Better Casting Design. (Foundry, 1944, vol. 72, Nov., pp. 76–79). Some examples are given of the manner in which radiographs of castings have been used to improve the positioning of risers and the dimensions of gates and runners so as to prevent shrinkage defects.

Quasi-Bessemerising Process. W. S. Williams. (Iron and Steel, 1944, vol. 17, Dec., p. 714). The quasi-Bessemer process of making ingot moulds of hematite iron has already been described (*see* Journ. I. and S.I., 1943, No. I., p. 80 A). In this paper some data showing how the silicon and phosphorus contents of the iron influence the graphite "throw" are presented and discussed. With not more than 0·06% of phosphorus present in the iron the graphite throw increases enormously with an increase in the silicon from 0% to 1·8%.

Centrifugal Casting of Steel. C. K. Donoho. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 313–329). Different methods for the centrifugal casting of steel are described, and the advantages and disadvantages of centrifugal casting as evaluated by a comparison with the properties obtained by forging and by casting in static moulds are discussed.

Spinning Speeds of Centrifugal Casting Machines. F. G. Carrington. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 333–346). The factors affecting the determination of the optimum mould-spinning speed for centrifugal casting are discussed with examples of the calculations involved.

Design and Safe Operation of Centrifugal Casting Machines. J. G. Weber. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 384–391). Examples of the stresses set up in centrifugal casting machines are cited, with notes on safety factors and various precautions which can be taken against accidents.

Hydro-Blast. (Metal Industry, 1944, vol. 65, Dec. 22, pp. 392–393: Engineering, 1945, vol. 159, Jan. 26, p. 66). A method of removing sand cores from heavy castings is described and illustrated. Jets of high-pressure water with entrained sand are used, the water pressure being 1250 lb. per sq. in. About 25 gall. of water per min. pass through the gun and about 70 lb. of sand are entrained per min. Equipment is provided for reclaiming the sand with the fines eliminated.

Some Practical and Economic Aspects of Small Foundry Conveyorization. H. B. Nye. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 414–420). The design and layout of conveyors for sand and for moulds and of pallet, roller and overhead conveyors for small foundries are discussed.

PRODUCTION OF STEEL

(Continued from p. 76 A)

The Past and Future of Steel. C. H. Desch. (Harold Wright Lecture: Iron and Coal Trades Review, 1945, vol. 150, Jan. 5, pp. 6–7). The past achievements, present position and prospects of the steel industry are discussed. Large quantities of carbon and alloy steel scrap will become available after the war, and research work might well be directed to im-

proving the recovering of such elements as chromium in the remelting of scrap metals. More low-alloy steel is likely to be available, and steps should be taken to see that it is used to the best advantage. British manufacturers will find it essential to use only the most efficient and economical means of production, and quality will be quite as important as cheapness. Some recommendations on the training of young men for the steel industry are made.

By-Products of the Steel Industry. C. Davies, jun. (Chemical and Engineering News, 1944, vol. 22, Sept. 25, pp. 1550–1554). The by-products such as scrap, slag and mill-scale from iron and steel production and those from coke-oven batteries are enumerated and information is given on the quantities produced and the uses to which they are put in the United States.

Chemicals Used in Steel Industry. J. L. Gregg. (Chemical and Engineering News, 1944, vol. 22, Sept. 25, pp. 1555–1558, 1612). A survey is made of the chemicals used in the production of steel in the United States with data on the quantities consumed in 1943.

Proceedings of the Twenty-Seventh National Open-Hearth Conference. (American Institute of Mining and Metallurgical Engineers, Open-Hearth Conference, Apr. 20–21, 1944). The Twenty-Seventh Open-Hearth Conference of the American Institute of Mining and Metallurgical Engineers was held at Pittsburgh, with L. F. Reinartz as the General Chairman.

After the opening remarks by the General Chairman the paper by J. W. Halley and G. L. Plimpton, jun., which received the McKune Award, was presented; this dealt with the relation between open-hearth practice and segregation in rimming steel (*see* p. 11 A). The session which followed concerned the problems of obtaining maximum production; in this, E. G. Hill reviewed the problems, G. Soler spoke on the relation of open-hearth production to mill schedules, H. J. Forsyth on production and surface quality, T. S. Washburn on steel quality and production, and L. R. Berner on the planning of the delivery of ingots to the mills as practised by the Inland Steel Co.; L. F. Reinartz concluded the session by reviewing the qualifications of the foreman, with observations on what the foreman should know and do.

The Conference then divided into two sessions, one on "Combustion in the Open-Hearth Furnace" and the other on "Metallurgy of the Open-Hearth Process." The former began with a description by H. Flagg of the construction, operation and performance of an experimental plant for mixing pulverised coal and fuel oil for heating open-hearth furnaces at the works of the American Rolling Mill Co. G. E. Seil followed with a paper on the physical and mechanical aspects of combustion in the open-hearth furnace. R. Noor then analysed the literature on combustion in the open-hearth furnace. F. N. Hays dealt briefly with methods of obtaining rapid and efficient heat transfer from the flame to the bath, A. J. Fisher presented a paper on flame radiation, burner design and their relation to open-hearth operation, and J. N. Albaugh briefly discussed the use of meters and controls on tilting furnaces. R. A. Lambert concluded the session by describing some recent improvements to open-hearth fuel control.

The session on the metallurgy of the open-hearth process opened with a discussion by B. M. Larsen on slag-bath reactions; this was followed by a paper by M. Tenenbaum on the distribution of phosphorus in the basic open-hearth furnace. G. R. Fitterer next reported on the activities of the Acid Open-Hearth Research Association. The following papers were then presented: "Slag-Metal Relations in the Acid Open-Hearth" by C. E. Sims, "Sampling for and Determination of Nitrogen in Steel" by W. D. Brown, "Hydrogen in the Open-Hearth" by R. M. Scafe, "Variation and Segregation of Nitrogen in Steel" by W. D. Brown, "Determination of

Hydrogen in Steel " by J. B. Austin, " An Effect of Dissolved Gases on Casting Structure " by C. A. Zapffe, " Hardenability Intensifiers " by S. W. Poole and F. L. Becker, and " Methods and Accuracy of Boron Determinations by the Quinalizarin Method " by L. C. Flickinger.

The next session was on the operation and maintenance of basic open-hearth furnaces. It commenced with an open discussion on the effect of war-time production on furnace maintenance and new equipment for cleaning slag pockets. H. L. Allen then presented an analysis of the answers received to a questionnaire on the operation and maintenance of mixers. An open discussion on sinter and the preparation of iron ore nodules then followed, and the session concluded with the following three papers: " Aluminium Dross as a Substitute for Fluorspar " by V. W. Jones, " Use of Cupola Iron in Open-Hearth Furnaces " by W. C. Buell, jun., and " Cupola Hot Metal for Open-Hearth Furnaces " by H. C. Barnes.

The next session dealt with factors affecting the quality of steel, this subject being divided into six sections, namely, slag control, carbon control, alloy contamination, bath temperature, deoxidation and segregation. W. O. Philbrook, jun., and A. H. Jolly, jun., reviewed methods of slag control. L. B. Luellen discussed sampling and the determination of carbon during steel production. O. Pearson described how spectrographic analyses were made at the Gary Works of the Carnegie-Illinois Steel Corporation, and C. H. Flickinger described the methods employed by the Republic Steel Corporation for determining residual alloying elements. A more comprehensive paper on determining residual alloying elements by spectrographic methods was read by P. R. Irish. T. J. Crawford, M. J. Bradey and R. B. Sosman contributed to an open discussion on bath temperature measurement. G. Derge described a vacuum-fusion method for the rapid determination of oxygen in steel. T. S. Washburn presented a summary of the answers to a questionnaire on deoxidation practice.

The concluding session on basic open-hearth furnaces covered refractories and masonry. It began with an open discussion on the life of refractories, which was followed by the following papers: " Determining the Quality of Silica Brick " by H. M. Kraner, " Refractories for the Open-Hearth Roof " by R. E. Birch, " Behavior of Metal Plates in a Basic Refractory at a Temperature of 3000° F." by L. S. Longenecker, " American Suspended Silica Roof " by F. D. Hazen, " Gunmix " by R. A. Schoenlaub (this paper describes a gun and equipment for making patches on hot furnace walls), " Gun Mixtures " by I. A. Nicholas, " Basic Roofs—Current and Prospective Applications " by H. M. Griffith, " Installation of Basic Open-Hearth Roof " by H. S. Frank, and " Development of Basic Open-Hearth Roofs " by R. P. Heuer. The session concluded with open discussions on regenerators, special refractories and innovations in lining maintenance.

At the separate Acid Open-Hearth Technical Session F. H. Allison presented a review of papers on gases in steel, and the session closed with a paper on flaking in alloy steels by E. R. Johnson, S. W. Poole and J. A. Rosa.

The Conference concluded with the Annual Fellowship Dinner at which J. L. Perry delivered an address on " The Human Side of Engineering."

Bagnall-Bethel Patent Nozzle. (Iron and Coal Trades Review, 1944, vol. 149, Dec. 1, pp. 841–842: British Steelmaker, 1945, vol. 11, Jan., pp. 34–35). A description of the Bagnall-Bethel ladle nozzle with which the rate of teeming can be increased is given. It consists of a single- or multi-nozzle box fixed with a quick-acting bayonet joint below the ordinary nozzle. The orifice in the former is smaller than that in the latter, and when a suitable number of ingots have been poured through the smaller orifice the secondary nozzle box is removed with a rapid movement of a special spanner.

The Treatment of Cooling Water. W. Heckmann. (Stahl und Eisen, 1944, vol. 64, Oct. 26, pp. 682–684). The treatment of steelworks' cooling water, especially by inoculation with hydrochloric acid and by the base exchange process, is discussed.

Acid Electric Steel-Making Practice (Notes for Operators). C. C. Wissmann. (Metal Progress, 1944, vol. 46, Oct., pp. 723–726). Recommendations are made on the operation of acid-lined arc furnaces for steelmaking.

Contribution to the Investigation of the Life of Ingot Moulds. A. Keller. (Giesserei, 1944, vol. 31, Nov., pp. 168–172). The relative merits of two hematite irons for making ingot moulds were investigated. One iron contained no combined carbon, whilst the other contained 0·22%; their graphitic iron contents were 3·57% and 3·19%, respectively. Dilatometer readings were taken during the repeated heating and cooling of specimens up to 950° C. and the microstructures were examined. The hematite iron with a ferritic structure without any pearlite was much less liable to retain stresses after repeated heating and cooling than was the pearlitic iron, but as far as mechanical wear was concerned the pearlitic iron was superior. The cause of the difference in the life of ingot moulds made of irons of almost identical analyses is often found to be due to differences in the raw materials used for making the hematite iron.

"Lead Prints" for the Detection of the Presence and Segregation of Lead in Steels and Brasses. B. S. Evans. (Armament Research Department: Analyst, 1944, vol. 69, Dec., pp. 368–369). To detect the distribution of lead in bonded steels and brasses a printing process is used which relies on the formation of brown lead sulphide. This method is not suitable when the lead is evenly distributed and the contents low. In the process described the sulphide conversion is replaced by conversion to the scarlet dithizone compound; this colour is so distinctive that there is no possibility of confusion with other substances.

REHEATING FURNACES

(Continued from p. 12 A)

Rolling-Mill Furnaces. W. Offenberg. (Stahl und Eisen, 1944, vol. 64, Oct. 26, pp. 679–682). A detailed description is given of a gas-fired roller-hearth reheating furnace for a plate mill. The furnace was built in segments and assembled outside the mill in such a way that it could be erected in the mill and put in commission in 36 hr. The inside dimensions of the furnace were 10 m. × 2 m. It was built in four sections each 2·5 m. long carrying seven 26/14 chromium-nickel steel rolls 185 mm. outside dia., 155 mm. inside dia. and 2900 mm. long. The furnace has a throughput of 150 tons per 24 hr., the plates entering at 560° C. and leaving at 763° C. The design is such that its length can readily be altered by adding or removing segments.

An "In and Out" Plate Mill Furnace with Two Twenty-One-Foot Door Openings. L. S. Longenecker. (Blast-Furnace and Steel Plant, 1944, vol. 32, Oct., pp. 1208–1212). An illustrated description is given of the reconstruction of some slab and plate reheating furnaces at the works of the Carnegie-Illinois Steel Corporation. The new furnaces are in pairs built end-to-end, and are fired from the ends. The front wall of each furnace consists of three doors each 7 ft. long; these and the guides between them can be lifted together, thus permitting slabs up to 20 ft. in length to be put in and removed without difficulty.

Experience with a Top and Bottom Heated Pusher Ingot Furnace with Three Zones. F. Fechter. (Iron and Steel Institute, 1944, Translation

Series, No. 203). An English translation is presented of a paper which appeared in Stahl und Eisen, 1944, vol. 64, July 13, pp. 449–453 (*see* p. 12 A).

Full Details of a New Walking Beam Furnace. (Sheet Metal Industries, 1944, vol. 20, Dec., pp. 2099–2102). Descriptions are given of two gas-fired walking-beam reheating furnaces of recent design. One is 75 ft. long and 5 ft. 6 in. wide. In both cases the beam mechanism is very robust, the beams consisting of structural steel water-cooled members carrying particularly deep and heavy refractory keyed blocks arranged to permit breathing due to heating and cooling.

Three-Zoned Temperature Control on a Continuous Furnace. B. M. Putich. (Iron and Steel Engineer, 1944, vol. 21, Nov., pp. 35–41). An illustrated description is given of a reheating furnace for ingots and slabs. The hearth is 62 ft. long × 17 ft. 6 in. wide. The furnace is divided into three parts, the heating, the high-temperature and the soaking zones, each of which is equipped with eight long-flame combination natural-gas and fuel-oil burners. Each zone is provided with temperature-recording and control instruments, details of which are given.

Heating for Forging. J. Mueller. (Steel Processing, 1944, vol. 30, Oct., pp. 667–669). A rotary-hearth forging furnace for billets 19 in. long is described, and recommendations on heating rates and forging temperatures for different steels are made.

FORGING, STAMPING AND DRAWING

(Continued from p. 77 A)

Designing Tools for Forging. G. Espey and P. Culhane. (Steel, 1944, vol. 115, Nov. 13, pp. 108–110, 148–168). The design of dies and mandrels for forging shells 3 in. in dia. and up by the pierce, bottom and drawn method is discussed.

High Speed Steel Forgings for Cutting Tools. W. H. Wills. (Steel Processing, 1944, vol. 30, Oct., pp. 656–657,659). The production of forgings from rolled high-speed steel billets is described and illustrated.

One-Shot Forging. A. F. Macconochie. (Steel, 1944, vol. 115, Oct. 30, pp. 74–77, 126). An illustrated description is given of the process developed by the J. I. Case Co. of Racine, Wisconsin, for producing $6\frac{3}{4}$-in. dia. shells 26 in. long from $5\frac{1}{2}$-in. square billets $14\frac{3}{4}$ in. long. The billets are heated in a rotary-hearth furnace and are passed through sizing rolls; any remaining scale is removed from the billet, which is then lubricated with a mixture of salt, graphite and mica. The dies and punch of the forging press are designed so that the shell is completed in a single stroke.

Drop Forging. (Automobile Engineer, 1944, vol. 34, Nov., pp. 483–490). A detailed and illustrated description is given of the modern drop-forging methods practised by George Morgan, Ltd., of Birmingham. Die design is discussed and the methods employed in the die-sinking and forge departments are described. Details are also given of the way that scientific control is exercised throughout all stages of production by systematic chemical, metallurgical and dimensional examination.

The Wire Die Room. P. Grodzinski. (Industrial Diamond Review: Wire Industry, 1944, vol. 11, Dec., pp. 617, 624). A description is given of the layout and organisation of a servicing department for wire-drawing dies.

ROLLING-MILL PRACTICE

(Continued from p. 50 A)

The Applications of Electronic Control in the Steel Industry. E. H. Vedder. (American Society of Mechanical Engineers: Sheet Metal Industries, 1944, vol. 20, Aug., pp. 1362–1364; Sept., pp. 1549–1550). Circuit diagrams for electronic control apparatus are presented and their application for controlling the speed of motors in rolling mills is discussed.

The Measurement and Recording of Roll Pressures. N. de Ball. (Stahl und Eisen, 1944, vol. 64, Nov. 9, pp. 716–720). A detailed description is given of an apparatus for obtaining continuous records of the changes in roll pressure in rolling mills. The apparatus is attached to the housing, the deformation of which is measured by electrical means.

The Automatic Preset Screwdown. R. E. Marrs. (Blast Furnace and Steel Plant, 1944, vol. 32, Oct., pp. 1191–1197). The principles of an automatic control system developed by the General Electric Co. for presetting screwdowns to the positions desired for the next rolling on a reversing plate mill are explained with circuit diagrams and illustrations.

Roll Designing. T. Smith. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. 61–63). Experience gained and some of the methods employed in the design and operation of rolls for producing bars of many shapes and sizes at the rolling mill of the Great Lakes Steel Corporation are described.

Effects of Heating and Rolling on Blooms, Billets and Bars. R. Sergeson. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. 35–40). Illustrated descriptions are given of defects which occur in ingots, blooms, billets and bars with notes on the factors contributing to their cause.

The Merchant Mills at the Great Lakes Steel Corporation. C. Moon. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. 59–60). A description is given of the 10-in. and 14-in. merchant bar mills at the Ecorse Works of the Great Lakes Steel Corporation.

Heating and Rolling in the Bar Mill. J. L. McHugh. (Iron and Steel Engineer, 1944, vol. 21, Nov., pp. 44–46). The practical aspects of rolling bars are discussed.

Main Roll Drives for Merchant-Bar and Rod Mills. W. B. Snyder. (Iron and Steel Engineer, 1944, vol. 21, Nov., pp. 48–57). Some modern bar and rod mills in the United States are described, with particular reference to their layout, power requirements and drives.

Rolling Alloy Steels at Atlas Steels, Limited. C. P. Hammond and A. M. Cameron. (Iron and Steel Engineer, 1944, vol. 21, Nov., pp. 66–72). An illustrated description is given of the rolling-mill plant at the works of Atlas Steels, Ltd., Welland, Ontario, which comprises a 26-in. 2-high blooming mill and a 3-high 22-in. mill driven by a 1750 h.p. and a 1250 h.p. D.C. reversing motor respectively. Notes on the rolling practice for alloy steel billets and bars are included.

Effects of Heating and Rolling in the Strip Mill. W. H. Richey. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. 42–44). A brief account is given of experience gained in the heating and hot-rolling of low-carbon steel strip in the 96-in. mill of the Jones and Loughlin Steel Corporation. The account deals mainly with the effect of temperature and rolling speed on the grain size of the finished strip.

Improvements to Wide Strip Mills. H. Herzog. (Stahl und Eisen, 1944, vol. 64, Oct. 12, pp. 654–656). Improvements to the reheating furnaces, transport equipment and general layout of two German mills rolling steel strip up to 1250 mm. and 300 mm. wide, respectively, are briefly described.

New Type of Cold Reducing Mill Placed in Operation. (Blast Furnace and Steel Plant, 1944, vol. 32, Oct., pp. 1220–1224). A description is given of the principles of the Krause cold-reducing mill. Two successful commercial mills are now in operation on copper and brass, and tests indicate that the principle can be applied to rolling steel. In the Krause mill the material is held by a gripper at one end which keeps the material under tension during the working stroke of the mill. At the beginning of the working stroke the rolls are in a position slightly behind the previously rolled material; the rolls are backed up by cam plates which are held in the mill housing. The rolls are turned by friction between the material and the cam plates; the drive is applied to move the housing and cam plates forward. During the working stroke the sloping surfaces of the cam plates bear on the top of the upper roll and the bottom of the lower roll so that the material is squeezed between the rolls and is moved forward at the same time. Very heavy reductions per pass can be made in this mill.

Production of Tubes and Tubular Forgings. (Engineer, 1944, vol. 178, Dec. 8, pp. 451–452). A description is given of the plant and processes at the works of Tubes, Limited, where tubes are made in a pilger mill and by cold-drawing with subsequent forging of the ends, heat treatment, pickling and finishing.

The Production of Rolling Stock, Especially Tyres, Wheel Centres and Disc Wheels. O. H. Lehmann. (Stahl und Eisen, 1944, vol. 64, Sept. 14, pp. 597–603; Sept. 21, pp. 618–622; Sept. 28, pp. 632–640). A comprehensive account with numerous diagrams and illustrations is given of modern German machinery and processes, including the heat treatment, for producing railway tyres and wheels.

HEAT TREATMENT

(Continued from pp. 78 A–80 A)

The Direct Resistance Heating of Salt Baths for Patenting Steel Wire. W. Heidenhain. (Stahl und Eisen, 1944, vol. 64, Nov. 23, pp. 747–750). Many wire-patenting plants in Germany have been converted from the use of molten lead to molten sodium nitrate (*see* p. 51 A). In the present paper tests with external and internal electrical resistance heating for salt baths are reported. The former method was unsatisfactory whilst efficient heating was obtained using plate electrodes immersed in the salt at opposite sides of the bath. When the salt is solid the electrodes are connected with a steel strip which heats up when the current is switched on; this melts the salt round it and, when sufficient salt has been melted to form a path for the current, the strip is removed, and very little current is required to keep the bath molten.

Methods and Equipment for Induction Heating. C. C. Levy. (Iron and Steel Engineer, 1944, vol. 21, Oct., pp. 51–58). Equations for calculating the power input necessary for inductively heating the surface of bars of magnetic and non-magnetic materials are presented, and motor-generator sets for producing high-frequency currents are described and illustrated.

Induction Heating and Its Application to Annealing and Melting. J. W. Cable. (Electrochemical Society, Oct., 1944, Preprint No. 18). Several examples of the application of induction heating for local annealing and complete annealing of steel shells, cartridge cases, shafts and tubes, as well as for melting metals in crucibles are described.

Case-Hardening by Megacycle Induction Heating. V. W. Sherman. (Electrochemical Society, Oct., 1944, Preprint No. 22). The thin case-

hardening of small parts by induction heating with current of 15 megacycle frequency is described and its advantages discussed.

Surface Heating by Induction. H. F. Storm. (Electrical Engineering, 1944, vol. 63, Oct., pp. 749–754). The induction-heating effect is analysed by using concepts commonly applied to alternating-current engineering without involving differential equations. It was found that ordinary vector diagrams lead to quantitative expressions for the penetration of the magnetic field into the charge, for the distribution of eddy currents and generated heat, as well as for the impedance of the loaded inductor. The analysis is limited to the heating of a body of cylindrical shape surrounded by a uniform alternating magnetic field.

Induction Heating—Selection of Frequency. N. R. Stansel. (Electrical Engineering, 1944, vol. 63, Oct., pp. 755–759). Formulæ are developed for calculating the frequency required for a given induction-heating service on the basis of the relation between the frequency and the kilovolt-ampere capacity.

Nitrided Steels. E. Ineson and C. Petteford. (Iron and Steel, 1944, vol. 17, Nov., pp. 699–702 ; Dec., pp. 735–738). A comprehensive account is given of the theory and practice of nitriding. The extent of nitriding in dissociated ammonia is affected by the amount of dissociation, and this is in turn dependent on : (1) The temperature at which the process is carried out ; (2) the pressure in the nitriding chamber ; (3) the rate of flow of ammonia ; and (4) the catalytic effect of the muffle surface and the workpiece surface. The properties of nitrided surfaces are discussed and the equipment for commercial nitriding is described. Nitrided parts can be softened by heating in a bath of equal amounts of sodium and potassium chlorides.

Nitriding the Austenitic Steels Improves Wear Resistance. I. A. Binder. (Machinist, 1944, vol. 88, Dec. 30, pp. 102–104). A description is given of the nitriding process as applied to the stems of 14/14 chromium-nickel steel exhaust valves for internal-combustion engines. The case obtained is only 0·004–0·006 in. deep after 50 hr. treatment at 1000° F. in partially dissociated ammonia, but this hardened surface considerably reduces the wear by contact with the guide.

The Nitriding of High-Speed Steel. (Machinery, 1944, vol. 65, Dec. 7, pp. 627–630). Descriptions are given of two nitriding processes, the "Maxi" and the "Holden Hy-Speed Case" treatments. The former consists primarily of nitriding the surface of high-speed steel tools after they have been hardened, tempered and finish-ground. The salt used is a eutectic mixture of sodium and potassium cyanides which is applied to the tools at 1025–1050° F. The latter is a liquid-bath nitriding treatment similar to that applied by commercial liquid-salt baths.

Skin Recovery for Decarburized Steel Surfaces. O. E. Cullen. (Metals and Alloys, 1944, vol. 20, Oct., pp. 954–958). Some particulars are given of a process for recarburising case-hardened parts which have suffered from surface decarburisation. The treatment consists of heating the parts in a prepared atmosphere the carbon pressure of which is carefully controlled so that when equilibrium is reached between the steel and the atmosphere the carbon content at the surface of the steel is restored to the desired level. The composition of the atmosphere is not stated, but it is prepared by heating a mixture of air and hydrocarbon gases in the presence of a catalyst.

The Effect of Time and Temperature on the Relief of Residual Stresses in Low-Alloy Steels. J. K. McDowell and R. C. Cunnick. (Welding Journal, 1944, vol. 23, Oct., pp. 481-S–493-S). The effect of the time and temperature of heat treatment on the relief of residual stresses was determined for a manganese-silicon-chromium steel, a carbon-manganese steel, a copper-nickel-phosphorus steel and two low-alloy steels. Specimens

0·505 in. in dia. were stressed in a fixture to values just above the yield point. Both specimen and fixture were then subjected to a stress-relieving heat treatment and the amount of stress remaining in the specimen was determined by measuring with an electrical strain gauge the amount of relaxation taking place when the load was released. The residual stress after treatment of 1, 4 and 8 hr. at 900°, 1000°, 1100° and 1200° F. was determined for each steel. No two steels reacted exactly alike to the various treatments, some showing larger percentage increases in the amount of relief than others as the time or the temperature of the treatment was increased. With one exception, treatment for 1 hr. at 1200° F. lowered the residual stresses to 2500 lb. per sq. in. or less. The stresses were reduced to 10,000 lb. per sq. in. by heat treatments varying from 8 hr. at 1000° F. to 4 hr. at 1100° F.

The Further Development of High-Speed Steels Low in Scarce Alloying Elements and Their Optimum Heat Treatment. H. Schrader. (Stahl und Eisen, 1944, Oct. 12, pp. 645–654). The results are presented of machining tests with 84 high-speed steels, and the effects of various combinations of the alloying elements tungsten, vanadium, molybdenum, chromium, titanium and aluminium on their cutting capacity, as well as the influence of different hardening and tempering treatments, are discussed.

Heat Treatment of Medium Carbon Cast Steel in Moderately Heavy Sections. Part II. K. L. Clark, H. F. Bishop and H. F. Taylor. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 539–542). The investigation at the American Naval Research Laboratory previously reported (*see* Journ. I. and S.I., 1943, No. II., p. 159 A) has been continued by comparing the effects of normalising and of furnace-cooling cast low-alloy steel blocks. The results are presented in an extensive table, and these indicated that in most cases a single normalising and tempering treatment would produce adequate properties in the castings.

Heat Treatment of Moderately Heavy Cast Steel Sections. K. L. Clark, H. F. Bishop and H. F. Taylor. (Foundry, 1944, vol. 72, Nov., pp. 82–84).

Continuous Hardening of Engine Parts. W. Roth. (Steel Processing, 1944, vol. 30, Oct., pp. 663–666). Illustrated descriptions are given of the following types of heat-treatment furnaces: (*a*) Continuous furnaces for small parts through which the parts pass on a conveyor belt and fall off the end into a quenching tank; (*b*) pusher furnaces for which the parts are carried in trays; the trays are subsequently lowered into the quenching tank so that the direction of the cooling in each part is controlled; (*c*) furnaces for parts which are handled individually and then quenched in a fixture; and (*d*) rotary hearth furnaces.

"Interrupted" Quenching in Salt Baths. A. P. Seasholtz. (Metal Progress, 1944, vol. 46, Oct., pp. 730–738). Earlier papers on the isothermal quenching of steel are reviewed and the use of salt baths for quenching is discussed. An agitated salt bath held at 400° F. gives a cooling rate approximately equal to that obtained by quenching in a mineral oil bath (without agitation) at room temperature.

Salt Bath Quenching Processes. H. J. Babcock. (Metals and Alloys, 1944, vol. 20, Oct., pp. 964–972). The principles of the heat treatment of steel are explained with particular reference to the use of S-curves. Details of Martempering, austempering and other processes of stepped quenching are given with data on some of the results achieved by them.

Double Temper Mo-W High Speed at Slow Heating Rate. K. J. Trigger. (Metal Progress, 1944, vol. 46, Oct., pp. 743–745). Tests are described which proved the desirability of double tempering a high-speed steel containing tungsten 5·5%, molybdenum 5%, chromium 4%, vanadium 1·5% and carbon 0·8%. The heat treatment which gave the best results consisted of heating to 2230° F., holding for 4 min., quenching in oil at 80° F.

until cold, heating to 1025° F. in $1\frac{1}{2}$ hr., holding for 1 hr., cooling in air and then repeating this tempering treatment.

Sub-Zero Treatment of Steels. H. C. Amtsberg. (Steel, 1944, vol. 115, Oct. 23, pp. 78–85, 124–126; Machinery, 1945, vol. 66, Jan. 11, pp. 29–34). The theory and practice of the heat treatment of tool steels is discussed with special reference to the incorporation of cold treatment at down to $-100°$ F.

WELDING AND CUTTING

(Continued from pp. 80 A–81 A)

The Effect of Metallurgical Changes Due to Heat Treatment upon the Fatigue Strength of Carbon-Steel Plates. W. H. Bruckner and W. H. Munse. (Welding Journal, 1944, vol. 23, Oct., pp. 499-S–510-S). The object of the tests described was to determine the effect of metallurgical changes associated with welding, without actually depositing a weld, on the fatigue strength of specimens of 0·25%-carbon steel. The specimens consisted of plates measuring $18 \times 3 \times \frac{7}{8}$ in., the centre portion of which was machined down to $5 \times 2\frac{1}{2} \times \frac{3}{8}$ in. One series of specimens was tested in the rolled and machined state; a second series was held in a vacuum at 1150° F. for 30 min.; three more series were prepared using devices by which a limited area at the centre of the specimens was heated to various temperatures in the 1100–1750° F. range and then quenched in water. The test data indicated that: (1) The second series of specimens had the lowest resistance to fatigue; (2) specimens "spot-heated" at the centre to above the upper critical point and quenched had the highest resistance to fatigue; and (3) specimens with a 1-in. wide strip across the centre heated to above the upper critical point and quenched had their fatigue strength slightly increased.

Some Observations on the Welding of Manganese Steels. W. B. Brooks and A. G. Waggoner. (Welding Journal, 1944, vol. 23, Oct., pp. 511-S–523-S). The weldability of manganese steels was studied by means of end-quench hardenability tests and bend tests on a bead of weld metal deposited down the centre of a thick plate. Manganese steel plates of $\frac{3}{8}$ in. min. thickness containing over 0·14% of carbon and over 1·1% of manganese have a pronounced tendency to crack under normal welding conditions. The hardness and ductility of the heat-affected zones of the steels tested were unsatisfactory. The end-quench test was not found very useful as a means of measuring the weldability. To obtain good welds in any weather in thick plates the manganese should be well below 1%. Lower weld-hardenability can be obtained without undue sacrifice in tensile properties by replacing part of the manganese with silicon, and probably with nickel and copper, with an addition of titanium or vanadium.

The Effect of Postheat in Welding Medium-Alloy Steels. M. A. Pugacz, G. J. Siegel and J. O. Mack. (Welding Journal, 1944, vol. 23, Oct., pp. 536-S–544-S). A hydrogen flame treatment for locally preheating or post-heating welded structures of medium-alloy steels has been developed at some fabricating shops which has not been entirely satisfactory due to lack of adequate control. An investigation is reported the object of which was to determine the effect of local heat treatment and to establish a suitable technique to apply in the welding of these steels. The effective post-heating time and temperature for any steel which requires supplementary heating during fabrication can be determined by some modification of the longitudinal bead-bend test. Increasing the post-heat temperature does not

necessarily make the treatment more effective; neither does an increase in the post-heating time beyond that necessary to assure complete transformation of the austenite formed during welding.

Helium-Shielded Arc Welding of Stainless Steel Exhaust Collectors. F. H. Stevenson. (Welding Journal, 1944, vol. 23, Oct., pp. 873–876). A description is given of the helium-shielded arc welding process which was successfully applied to the welding of stainless steel parts for heat-exchanger and exhaust manifolds. A carbon electrode is made the negative pole and this is held in the torch to which helium is fed at the rate of about 5 litres per min. The filler metal consists of a stainless steel rod. The technique is very similar to that of oxy-acetylene welding. The results of tests on welded specimens are presented and discussed.

Arc Welding Practice in the Steel Foundry. F. Kiper and L. Gabes. (Welding Journal, 1944, vol. 23, Oct., pp. 877–881). A description is given of the welding practice for conditioning steel castings at the works of the Ohio Steel Foundry Company.

Arc Welded Cutting Alloys. E. C. Rollason and P. Harris. (Metallurgia, 1944, vol. 31, Nov., pp. 3–6). The application of Stellite, high-speed steel and special alloy electrodes for making lathe tools, milling cutters and other cutting tools is discussed. A new development is the use of mild steel core wire electrodes with a heavy metallic coating containing the alloying elements; by this means a variety of properties far in excess of that of ordinary high-speed steel deposits can be achieved, and care can be taken that the fundamental requirements of hot hardness and abrasion resistance are fulfilled.

Factors Controlling the Weldability of Steel. L. Reeve. (Welding, 1944, vol. 12, Nov., pp. 521–530; Dec., pp. 573–579). The author points out that the hardness of the heat-affected zone is not the only criterion of the weldability of steel. British investigations of the weldability of high-tensile steels and German and American work on low-alloy steels are reviewed. The conclusions from British reports are that weldability is judged largely (but not entirely) by the maximum welding hardness and that it may be gauged approximately from the analysis of the steel to be used, converted to its "carbon equivalent," from which the probable welding hardness can be predicted. The necessary size of weld to be deposited to avoid unsafe hardness can also be predicted approximately, taking into account the plate thickness and the preheating temperature. The possibility of cracking occurring throughout the deposit, particularly in butt welds, makes it necessary for the ultimate test of weldability to be an actual welding test, preferably of the restrained joint type.

The Bead-Weld Nick-Bend Test for Weldability. C. E. Jackson and G. G. Luther. (Welding Journal, 1944, vol. 23, Oct., pp. 523-S–535-S). The usefulness of various weldability tests is reviewed and details of the bead-weld nick-bend test for assessing the ductility of the heat-affected zone are given. In this test a single bead is deposited across the centre of a plate measuring 12 × 6 in. Two strips $1\frac{1}{2}$ in. wide are cut longitudinally from the plate and a nick is cut with a hack-saw or machined down the centre of the weld metal to a depth of $\frac{3}{64}$ in. below the plate surface. This nick does not enter the heat-affected zone, so that initial deformation in the subsequent bend test occurs in the weld metal and failure is caused by rupture of the low-ductility metal in the heat-affected zone. The results of many tests are presented and correlated with other test results. Special attention is drawn to the marked improvement in the ductility which was observed on prolonging the interval between welding and testing.

Fundamentals of Heavy Cutting. J. L. Walker and H. G. Hughey. (Welding Journal, 1944, vol. 23, Oct., pp. 934–942). Comprehensive data relating to the oxy-acetylene cutting of steel 12 in. thick and even heavier

are presented and discussed and recommendations on the technique are made.

MACHINING

(Continued from p. 54 A)

The Selection and Use of Cutting Fluids. H. W. Fowler, jun. (Steel, 1944, vol. 115, Nov. 13, pp. 122–126). A large number of ferrous and non-ferrous alloys are classified into six groups according to their machinability and recommendations are made as to the best cutting fluid to use when performing different machining operations on each group.

Evaluating the Quality of High-Speed Steels. W. Eilender and H. Arend. (Stahl und Eisen, 1944, vol. 64, Nov. 23, pp. 742–746). The manner in which the distribution and grain size of the carbides in 72 high-speed steels was affected by the hardening temperature and the formation of zones of secondary hardness was studied. The optimum hardening temperature increased with increasing fineness of the carbides and grain size. Both the width and the hardness of the zones of secondary hardness also increased with decreasing grain size.

Refrigeration in the Machining of Metals. G. Pahlitzsch. (Zeitschrift des Vereines Deutscher Ingenieure, 1944, vol. 88, July 8, pp. 365–371). The life of high-speed steel tools has been found to be enormously increased by using a refrigerated lubricant. The results of machining tests using lubricants cooled to within the 2–12° C. range are presented and discussed.

Measuring and Designating Surface Finish. J. A. Broadston. (Iron Age, 1944, vol. 154, Oct. 19, pp. 62–66; Oct. 26, pp. 76–81; Nov. 2, pp. 58–62). In the first part of this series of articles different types of surface finish on machined metals are described and definitions of the terms "flaw," "roughness," "waviness," "lay" and "*RMS* microinch value" are given. The *RMS* microinch value is the Root Mean Square of an infinite series of distances (in microinches) between the hypothetical nominal or mean surface line to the actual irregular surface contour. Symbols used by draughtsmen to indicate degrees and types of surface roughness are explained. In the second part the measurement of surface roughness by the profilometer and the Brush Analyser is described. In the concluding part methods of comparing surface finishes are given together with particulars of a new low-priced comparison roughness gauge made of a plastic material.

PROPERTIES AND TESTS

(Continued from pp. 84 A–88 A)

Non-Destructive Testing. (Midland Metallurgical Societies: Metal Industry, 1944, vol. 65, Dec. 22, pp. 386–388; Dec. 29, pp. 406–408). To commence an open discussion on the non-destructive testing of metals three short papers were read by H. M. Bigford, G. Biddle and S. H. Griffith respectively. In these and the discussion which followed, electrical and magnetic methods of testing and other methods employing the senses of sight, sound, and even smell, were dealt with.

Cupping Tests. S. Menghi. (Metallurgia Italiana: Metal Industry, 1944, vol. 65, Dec. 1, pp. 338–340; Dec. 8, pp. 354–357; Dec. 15, pp. 376–377). The Erichsen, K.W.I. (Kaiser Wilhelm Institut für Eisenforschung), Persoz and Jovignot deep-drawing tests for sheet metal are compared and the optimum conditions for their employment discussed.

Fatigue Resistance of NE Steel Shafts. O. J. Horger and T. V. Buckwalter. (Iron Age, 1944, vol. 154, Nov. 16, pp. 60–63). **Endurance of NE Steels in 1¾-In. Specimens.** O. J. Horger and T. V. Buckwalter. (Metal Progress, 1944, vol. 46, Oct., pp. 727–729). The results of fatigue tests on four low-alloy chromium-nickel-molybdenum steels are presented. The specimens were 1¾ in. in dia. and the object was to determine how changes in the fillet or notch radius affected the fatigue strength.

Shot Peening and the Fatigue of Metals. H. F. Moore. (Iron Age, 1944, vol. 154, Nov. 2, pp. 67–71, 136–142). Tests to determine the effect of shot-peening on the properties of metals are described and the results discussed. Shot-peening increases the tensile strength of a metal just below the surface and also sets up longitudinal compressive stress in the thin skin of peened metal. A limited amount of peening increases the fatigue strength, but too much peening can be applied and this reduces the fatigue resistance. The corrosion resistance of steel is increased by peening. Shot-peening can be applied to parts of irregular shape which might be distorted by heat treatment and local application is advantageous in such cases as the fillets of shafting where stress concentrations occur.

Fatigue Life of Stressed Parts Increased by Shot Peening. D. C. Turnbull. (Machinist, 1944, vol. 88, Dec. 16, pp. 83–86). Some examples are quoted which illustrate how shot-peening increases the fatigue strength of coil springs and gear teeth and a method of determining the intensity of peening is described. If a thin strip of hard steel is peened only on one side the surface fibres on that side are stretched and the strip becomes curved; this curvature increases with additional peening and a gauge reading in thousandths of an inch has been devised to measure this curvature and thus provide a rapid method of determining peening intensity.

Shot for Metal Peening. O. E. Harder and J. T. Gow. (American Society for Metals, Oct., 1944, Preprint No. 32). The increased use of shot-peening to improve the fatigue strength of metals has drawn attention to the selection of the most suitable shot. The properties and structure of chilled iron shot, malleablised chilled shot and of hardened and tempered iron shot are described and discussed.

Using the Jominy Test to Predict Physical Properties after Tempering. E. R. Weiher. (Steel, 1944, vol. 115, Oct. 23, pp. 90–94). It is shown how to correlate the results of Jominy end-quench tests with the hardness readings obtained at the surface, at half radius and at the centre of round bars of different diameters of the same steel and how to apply this correlation to determine the tempering temperatures required to obtain a given hardness.

Hardenability of Some Cast Steels. J. B. Caine. (Transactions of the American Foundrymen's Association, 1944, vol. 52, Dec., pp. 459–473). The results of end-quench hardenability tests on a number of fine-grained, aluminium-killed, plain carbon and low-alloy cast steels are presented and discussed. There is a significant difference between the experimentally determined hardenability curves for cast steel and those for hot-worked steel; the higher hardenability of the cast steel is caused by the slightly higher silicon and aluminium contents. The hardenability of cast steel can be calculated from the chemical composition and grain size.

On the Relationship between the Hydrogen Content of Iron, the Temperature and the Hydrogen Equilibrium Pressure. G. Phragmén. (Jernkontorets Annaler, 1944, vol. 128, No. 10, pp. 537–552). (In Swedish). An attempt is made to develop an equation for extrapolating the equilibrium pressure for hydrogen in iron to temperatures down to 200° C. and for exceedingly high pressures.

The Selection of Steels Used in the Rolled or Drawn Condition. Part I. Plates. J. H. Frye. (Metal Progress, 1944, vol. 46, Oct., pp. 705–712).

The degrees of importance attached to the strength, ductility, toughness, grain size, wear resistance, corrosion resistance and ability to withstand cold and hot work when specifying steel, particularly steel plates, are discussed.

Post-War Steel and Its Treatment. H. W. McQuaid. (Metal Progress, 1944, vol. 46, Nov., pp. 1067–1073). **Selecting Steels for Postwar Applications.** H. W. McQuaid. (Steel, 1944, vol. 115, Nov. 6, pp. 122, 168–174). Factors affecting the selection of steels for post-war automobile manufacture and other purposes are discussed. Most requirements can probably be met by securing the correct combination of manganese and the residual chromium and nickel which will be present in the steel as a result of melting down scrap arising from munitions and armaments.

Trends in Alloy Steels. A. B. Kinzel. (Metal Progress, 1944, vol. 46, Oct., pp. 689–692). Trends in the manufacture and application of alloy steels are discussed in the light of the war-time and expected post-war conditions of steelmaking in the United States.

Developments in Corrosion and Heat-Resisting Steels. L. Sanderson. (Engineering and Boiler House Review, 1944, vol. 58, Dec., pp. 308–310). An account is given of some war-time developments in corrosion- and heat-resisting steels. When stainless steels are alloyed with titanium, the nitrogen content must be kept as low as possible; if this is not done the prevention of intergranular corrosion may be impeded, the steels will be richer in non-metallic inclusions and much of the titanium will be robbed of its effectiveness. When columbium is added to 18/8 stainless steel its creep resistance at 600° C. is considerably increased, whilst titanium has little effect. Stainless steel electrodes have been successfully used for welding high-speed steel to carbon-steel shanks. The strength at high temperatures of fine-grained steels killed with silicon-aluminium is greater than that of steels killed by other methods.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 62 A–66 A)

The Microscope. (Automobile Engineer, 1944, vol. 34, Nov., pp. 475–482). The construction of metallurgical microscopes is described with notes on specimen preparation by polishing and etching, the mounting of sections and the interpretation of the patterns and structures revealed.

The Electron Microscope for Metals. R. G. Picard and P. C. Smith. (Metals and Alloys, 1944, vol. 20, Sept., pp. 636–641). A brief description of the electron microscope is given with explanations of the surface replica technique. Reproductions of pictures of metallic oxide smokes, pearlite and stainless steel at very high magnifications are presented.

A Comparison between the Back-Reflection and Replica Processes in the Electron-Microscopical Reproduction of Surfaces. E. Semmler-Alter and I. Ziesecke. (Zeitschrift für Metallkunde, 1944, vol. 36, May, pp. 115–120). The principles and characteristics of the back-reflection and replica processes of using the electron microscope are discussed and examples of the pictures obtained when examining samples of aluminium alloys, copper and steel are reproduced. When it is desired to study the amount of relief it is better to employ the back-reflection technique, but when the object is to examine very fine and closely packed structures the replica method gives much better definition.

X-Ray Diffraction. A. R. Weill. (Iron and Steel, 1944, vol. 17, Dec. 7, pp. 741–749). The theory of X-ray diffraction and its application to the

study of metals and alloys by the powder and back-reflection methods are explained in detail with many examples.

Preferred Orientations in Rolled Metals; Construction of Pole Figures. B. F. Decker. (Proceedings of the American Society for Testing Materials, 1943, vol. 43, pp. 785–802). The orientations of crystals in rolled sheet metals as revealed by X-rays are studied with the aid of a pole figure. Such a pole figure is a summary by stereographic projection of data for one set of crystal planes from a series of X-ray photographs taken at different angles to the sheet. The plotting of a pole-figure chart and the interpretation of pole figures is described.

Distribution of Carbon between Titanium and Iron in Steels. W. P. Fishel and B. Robertson. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1763: Metals Technology, 1944, vol. 11, Oct.). The distribution of carbon between titanium and iron was studied by measuring the relative amounts of iron carbide and titanium carbide present in a series of annealed steels in which the titanium/carbon ratio extended from 0·527 to 4·61. The formula of titanium carbide present in annealed steels is TiC. In titanium-bearing steels practically 100% of the titanium is present as TiC so long as the ratio Ti/C by weight in the alloy is below a value lying between 3·57 and 4; this means that titanium is a very strong carbide former. The possible existence of a double titanium-iron carbide was investigated; it was established that no such double or complex carbide exists as a true compound in annealed steels.

CORROSION OF IRON AND STEEL

(Continued from pp. 88 A–89 A)

A Generalized Theory of Stress Corrosion of Alloys. R. B. Mears, R. H. Brown and E. H. Dix, jun. (American Society for Testing Materials and American Institute of Mining and Metallurgical Engineers, Symposium on Stress-Corrosion Cracking, Nov., 1944, A.S.T.M. Preprint No. 20). The results of stress-corrosion tests, mostly with aluminium alloys, are presented and discussed. These indicated that: (1) Stress is necessary for cracking to occur; (2) the corrosion takes place along localised paths; (3) the corrosion occurs in most cases by an electrochemical mechanism, the localised path being the anodic area and the surrounding area being cathodic; and (4) once the crevices or localised paths are corroded, the stress opens the crevices further.

Elevated Temperature Tension Tests on Galvanized Steels. J. H. Craig. (American Society for Testing Materials and American Institute of Mining and Metallurgical Engineers, Symposium on Stress-Corrosion Cracking, Nov., 1944, A.S.T.M. Preprint No. 21). Tests are described the object of which was to determine at what temperature intergranular penetration of galvanised steel by zinc takes place under conditions of rapidly applied stress and heat. The tests were conducted in the range from room temperature to 1400° F. In this range, low-carbon and high-carbon steels did not suffer any adverse effects from the zinc coating. In fact, at 1400° F. the zinc coatings appeared to have a beneficial effect. With stainless steel, the coating had no effect at up to 1000° F. Above this temperature the zinc coating appeared to cause a marked loss in ductility and a less pronounced loss in strength; there was, however, no visible evidence of zinc penetrating the structure of the stainless steel.

Some Observations of Stress-Corrosion Cracking in Austenitic Stainless Alloys. M. A. Scheil. (American Society for Testing Materials and

American Institute of Mining and Metallurgical Engineers, Symposium on Stress-Corrosion Cracking, Nov., 1944, A.S.T.M. Preprint No. 22). Methods of testing the susceptibility of stainless steels to stress-corrosion cracking are described. These were developed by the A. O. Smith Corporation and consisted of suspending stressed specimens of the steel in jars containing a solution of magnesium chloride boiling at 309° F. and noting the time taken for the initiation of cracks and until fracture occurred. The results of tests on many steels containing up to 25% of chromium and up to 35% of nickel are presented. It was observed that austenitic stainless steels are susceptible to stress-corrosion cracking under certain conditions irrespective of their susceptibility to intergranular corrosion.

The Susceptibility of Austenitic Stainless Steels to Stress-Corrosion Cracking. R. Franks, W. O. Binder and C. M. Brown. (American Society for Testing Materials and American Institute of Mining and Metallurgical Engineers, Symposium on Stress-Corrosion Cracking, Nov., 1944, A.S.T.M. Preprint No. 23). Corroding media were examined as to their suitability for testing the susceptibility of annealed and cold-rolled austenitic chromium-nickel steels to stress-corrosion cracking. The media which most readily produced this cracking were boiling concentrated solutions of magnesium chloride, lithium chloride, zinc chloride, calcium chloride and ammonium chloride. Contact with any one of these solutions did not cause cracking unless the specimens were critically stressed beyond the yield strength at 0·2% offset. Cold-rolled austenitic stainless steels can be stressed to a very high value in a seaside atmosphere without danger of failure. Samples with simple bends, sheared edges, and impressions made by the Erichsen drawing test were tested in boiling 42% magnesium-chloride solution: Samples severely cold-worked by deep-drawing were more subject to stress-corrosion cracking than samples cold-worked by simple bending or rolling. The cantilever sample appeared to be suitable for determining the critical stress at which the steels would crack on the tension side in any of the corrosive solutions.

Some Examples of Stress Corrosion Cracking of Austenitic Stainless Steel. O. B. Ellis. (American Society for Testing Materials and American Institute of Mining and Metallurgical Engineers, Symposium on Stress-Corrosion Cracking, Nov., 1944, A.S.T.M. Preprint No. 24). Investigations of the failure by stress-corrosion cracking of stainless steel equipment are briefly described. The items in question included two coffee urns, two wool-conditioning units, a dyeing machine and a steam-jacketed pot, all of which failed in a short time in conditions of high humidity at temperatures below 212° F. The cracks were mainly transcrystalline and there were no indications of intergranular cracking, although it is possible that the initial attack was at the grain boundaries on the surface.

The Selection of Specimens for Testing the Resistance to Stress-Corrosion of Aluminium-Alloy and Steel Pressed and Rolled Products. F. Bollenrath, W. Bungardt and H. Cornelius. (Zeitschrift für Metallkunde, 1944, vol. 36, Apr., pp. 73–84). The resistance of aluminium alloys to stress-corrosion cracking was found to vary greatly with the position from which the specimens were cut and the grain direction. This led to extending the investigation to three low-carbon steels, one 3%-chromium steel and three chromium nickel steels (one containing tungsten). Fork-shaped specimens 40 × 14 × 6 mm. with a 6-mm. gap between the two arms were cut from different positions in billets; these were stressed by pulling the ends of the arms apart; they were then immersed in a hot aqueous 64%-$NaNO_3$ solution and the number of days which had elapsed on fracture occurring were recorded. The specimens of the carbon steels and the annealed nickel-chromium-tungsten steel which were cut from transverse sections of the billet were much more sensitive to stress-corrosion than those cut

longitudinally. A theory explaining this is put forward. The effect of the direction in which the specimen is cut upon its resistance to stress-corrosion is not so great that there is danger of transverse specimens failing when the steel is satisfactory in the longitudinal direction by reason of its composition and structure.

The Intercrystalline Corrosion-Cracking of Preheater Tubes. W. Mantel. (Zeitschrift des Vereines Deutscher Ingenieure, 1944, vol. 88, Apr. 29, pp. 238–240). An investigation of the cause of longitudinal and transverse cracks in copper-bearing steel preheater tubes is described. The tubes, which carried feed-water at 130–140° C. for a boiler installation, cracked on the inside and eventually leaked. It was found that the feed-water contained traces of oxygen which, under the plant conditions, oxidised the inner surface of the tube forming a protective oxide layer; this layer became firmly embedded in the steel of the tube. Temperature changes in the tube caused elongation and the brittle oxide cracked in a direction at right angles to the principal stresses. More oxygen penetrated the cracks to the metal of the tube and the process repeated itself until the tubes leaked.

Emulsions of Oil in Water as Corrosion Inhibitors. P. Hamer, L. Powell and E. W. Colbeck. (Iron and Steel Institute, 1945, this Journal, Section I). This paper describes an investigation which was undertaken with the object of preventing corrosion in recirculating cooling-water systems, and which was begun as a result of plant failures. The origin and object of the work are discussed, together with the factors influencing the choice of experimental method. The oil emulsions referred to are those produced by adding " soluble oils " to water. These resemble the cutting oils employed on machine tools. Attention has been chiefly directed to the prevention of attack on mild steel. Three types of water have been used under both static and flow conditions at room temperature, and at 60° and 90° C.

The investigation was limited to an attempt to answer certain questions which appeared to be of practical value. These questions were:

(*a*) What is the lowest concentration of emulsified oil required to prevent the corrosion of mild steel?

(*b*) Does pre-treatment with oil emulsion give protection in the absence of oil, and/or can a heavy initial dose be followed by a light dose?

(*c*) Can oil emulsion stop pitting or general corrosion which has already commenced?

(*d*) Do weak oil emulsions cause pitting or intensified local attack?

(*e*) Do oil emulsions prevent corrosion of non-ferrous metals alone or in contact with mild steel?

The work described in the paper has provided the following very general answers to these questions:

(*a*) Oil emulsions containing 0·5% or more of emulsifying oil reduce to negligible proportions the corrosion of mild steel under all conditions tested.

(*b*) The protection afforded by oil emulsion persists for some time after the specimen has been transferred to untreated water.

(*c*) There is evidence that oil emulsions soften preformed rust and tend to reduce the rate of attack.

(*d*) Localised attack can occur at parts of the metal which are inaccessible to the emulsion.

(*e*) Oil emulsions are effective inhibitors of the corrosion of the non-ferrous metals studied, but when in contact with mild steel some attack occurred at elevated temperatures in certain cases.

A suggested mechanism of the protective effects of oil emulsions is that they form a thin adsorbed film of oil over the whole metal surface, which may break down at specially susceptible spots. The film is then thickened due to coagulation of the emulsion by the ferrous salts formed at these anodic points, thus stifling the corrosion. The effect of the oil film in reducing heat transfer through clean metal surfaces was found to be negligible, but heat transfer was distinctly impaired when the metal surface was already covered with corrosion product. In hard waters the oil emulsions tend to produce scums, which may be due to the formation of insoluble soaps or to inversion of the emulsion to water-in-oil. The importance of these scums in practice cannot yet be assessed, but it may be necessary to adopt special measures to prevent the fouling of heating surfaces when the water is recirculated.

Nature of the Oxide Film of Fe–Cr–Al Solid Solution. I. I. Kornilov and I. I. Sidorishin. (Comptes Rendus (Doklady) de l'Académie des Sciences de l'URSS, 1944, vol. 42, No. 1, pp. 20–23). An investigation of the oxidation at temperatures in the 400–1000° C. range of iron-chromium-aluminium solid solutions by means of electron diffraction diagrams and chemical analysis is reported. At moderate temperatures an isomorphous mixture of oxides of the spinel type of the three metals with a cubic crystal lattice was formed on the surface of the alloy. With increasing temperatures the lattice constant of the solid solution of oxides decreased and it eventually approached that of the pure γ-Al_2O_3 lattice. The oxide film which formed on the surface of the alloy in the form of pure alumina crumbled easily and had no protective properties.

Theory of Oxidation of Iron-Chromium-Aluminium Solid Solutions. I. I. Kornilov. (Comptes Rendus (Doklady) de l'Académie des Sciences de l'URSS, 1944, vol. 42, No. 4, pp. 181–184). The manner in which the rate of oxidation of aluminium in iron-chromium-aluminium solid solutions is influenced by the amount of aluminium present was investigated. Evidence was obtained that the process takes place in two stages, first the oxidation of all three metals on the surface and secondly the diffusion of aluminium caused by its selective oxidation on the surface of the alloy. A relationship was established between the loss of aluminium by oxidation and the initial content of this element in the ternary solid solution.

Effects of Oxygen Exhaustion from Corrosive Solutions on High Nickel-Chromium-Molybdenum Alloy Steels. W. E. Pratt. (Electrochemical Society, Oct., 1944, Preprint No. 5). Two recent investigations of failures of stainless steel equipment based on laboratory tests and on many years of service have stressed the importance of determining in each case the oxidising capacity of a corrosive solution under operating conditions that is necessary to create a passive state and resistance to corrosion. In one case the evolution of hydrogen from pickled steel caused loss of oxidising capacity of weak hot sulphuric acid. In another case sludge acid had been boiled and was given little or no opportunity to absorb oxygen from the air before it was in contact with the alloy steel equipment. The procedures employed in discovering the causes of failure and the methods adopted for correcting the unusual corrosive conditions are described.

Diagnostic Methods in Problems Concerned with the Corrosion of Cans. W. B. Adam and D. Dickinson. (Chemistry and Industry, 1944, Dec. 16, pp. 426–428). Methods of tracing the cause of hydrogen swells in cans of fruit are discussed. A complete statistical solution is not offered but a statistical technique is proposed by which a mass of accumulated data can be compressed into such a form that it can be used conveniently when the time comes to make a general survey of the evidence.

Protection of Steel Parts for Overseas Shipment. H. A. Knight. (Metals and Alloys, 1944, vol. 20, Oct., pp. 934–940). Descriptions are given of

methods of rust-proofing and packing which have been developed in the United States for protecting military equipment which is sent overseas. These include: (1) A spray of a plastic compound based on "Vinylite" resins; this forms a flexible coating which can be quickly stripped from the part. (2) Wrapping in a sealed transparent film-like material. (3) Sealing the part with a dehydrating agent in a moisture-proof wrapper.

Fouling of Ships' Bottoms: Identification of Marine Growths. (Iron and Steel Institute, 1945, this Journal, Section I). This paper has been prepared by the Marine Corrosion Sub-Committee (a Sub-Committee of the Joint Corrosion Committee of The Iron and Steel Institute and The British Iron and Steel Federation) with the following objects: (1) To emphasise the deleterious effects of fouling on the operation of a ship; (2) to give a brief account of the settlement and growth of the marine organisms which are mainly responsible for fouling; (3) to discuss briefly the methods of anti-fouling research and the means of preventing fouling; and (4) to describe the characteristics of marine growths in sufficient detail so that they may be correctly identified when present on ships. The text is illustrated by 45 coloured and monochrome photographs and line drawings of the fouling organisms so that with a little experience the layman should be able to identify them with ease and certainty. A specimen of a dry-docking report form is reproduced and users of ships are asked to co-operate with the Marine Corrosion Sub-Committee by reporting on this form the occurrence, position, size and type of organism found when the ship is dry-docked, together with details of the voyages, painting, bottom corrosion, &c.

The Sub-Committee's investigations have shown that of the three most common poisons, mercury is more than twice as effective as copper, weight for weight, whereas arsenic *per se* is almost ineffective. Certain organic poisons may be used, but their successful incorporation into a paint presents a number of special problems.

ANALYSIS

(Continued from pp. 68 A–69 A)

A Method for Determining Low Carbon Contents in Steel. G. Ericsson. (Jernkontorets Annaler, 1944, vol. 128, No. 11, pp. 579–596). (In Swedish). An account is given of an electrical-conductivity method of determining low carbon contents in steel. An ordinary furnace for carbon determinations is used and the carbon dioxide is absorbed by a solution of $Ba(OH)_2$. The change in the electrical resistance of the solution caused by the formation of $BaCO_3$ is measured by means of a Wheatstone bridge consisting of a rheostat, a simple oscillator and a "Philiscop" containing a cathode-ray tube. A thermostat keeps the temperature of the solution constant. The absorption bulb is constructed according to the principles given by M. H. Kalina and T. L. Joseph (*see* Journ. I. and S.I., 1939, No. II., p. 113 A). The combustion tube of the furnace is fitted with a special seal to prevent air from entering the tube when open. The oxygen gas is washed free from carbon dioxide before it enters the combustion tube. The apparatus can be calibrated with air of known carbon-dioxide content. A small piece of marble weighed on a micro-balance may also be used for calibration purposes. Samples of 0·5–2·0 g. of steel containing less than 0·05% of carbon can be analysed by this method. With care, results differing from the mean value by a maximum of $\pm$ 0·0005% can be obtained.

Colorimetric Determination of Chromium in Steel. L. Singer and W. A. Chambers, jun. (Industrial and Engineering Chemistry, Analytical Edition, 1944, vol. 16, Aug., pp. 507–509). A colorimetric method of

determining chromium in steel when present in amounts up to 1% is described. The basis of the method, which is not subject to interference by iron or any alloying element usually present, lies in the fact that ferric perchlorate, which is itself colourless, intensifies the colour of the dichromate ion.

Spectrographic Determination of Small Amounts of Tungsten in Steel. P. Fischer, R. Spiers and P. Lisan. (Industrial and Engineering Chemistry, Analytical Edition, 1944, vol. 16, Oct., pp. 607–609). A modification of Knowles' chemical method of separating tungsten when making determinations of small amounts (0·01–0·25%) of tungsten in steel and a spectrographic procedure for completing the determination are described.

Nitrogen—Its Determination in Mild Steel by a Semi-Micro Method. S. D. Steele. (Iron and Steel, 1944, vol. 17, Nov., pp. 669–670). The method is described and the apparatus illustrated whereby nitrogen in mild steel may be determined colorimetrically. The method is superior to that of Allen in that determinations are made more rapidly and the results show greater consistency.

The Systematic Detection of Metals on the Small Scale. Christina C. Miller. (Metallurgia, 1944, vol. 31, Nov., pp. 39–42). An account is given of the application of small-scale methods to the systematic qualitative analysis of inorganic substances. Emphasis is laid on the employment of modern sensitive and selective reagents, especially organic substances for the identification of metals.

The Spectrograph as an Inspection Tool. J. D. Graham and H. F. Kincaid. (Metals and Alloys, 1944, vol. 20, Aug., pp. 355–358). The principles of the grating-type spectrograph are explained and the new spectrographic laboratory at the Farmwall Works of the International Harvester Company is described.

Standard Samples Issued or in Preparation by the National Bureau of Standards. (United States National Bureau of Standards, 1944, Supplement to Circular No. C398).

Fluorine in Coal. I. The Micro-Analytical Determination of Fluorine. H. E. Crossley. (Journal of the Society of Chemical Industry, 1944, vol. 63, Sept., pp. 280–284). Micro-analytical methods for the determination of fluorine are reviewed and an account is given of trials which led to the rejection of various methods. The method eventually selected and modified depends on the fading action of fluorides on zirconium-alizarin lake.

Fluorine in Coal. II. The Determination of Fluorine in Coal. H. E. Crossley. (Journal of the Society of Chemical Industry, 1944, vol. 63, Sept., pp. 284–288). A description is given of various methods in which it was sought to apply to coals the micro-analytical method selected (*see* preceding abstract). The low results given by the incineration method were corrected by fusing after incineration and removing silicates with zinc compounds before distillation.

Analyses of Virginia Coals. (United States Bureau of Mines, 1944, Technical Paper No. 656). A series of papers by various authors is presented in which the geological structure of the coalfields of Virginia, the mining conditions, the preparation of the coal, data on the coal industry, and coal analyses and properties are dealt with.

Determination of Hydrocyanic Acid, Especially in Coke-Oven Gas. J. A. Shaw, R. H. Hartigan and A. M. Coleman. (Industrial and Engineering Chemistry, Analytical Edition, 1944, vol. 16, Sept., pp. 550–553). A method is proposed for the rapid and accurate determination of hydrocyanic acid in coke-oven gas. It is based on the conversion of the cyanide or certain of its derivatives to cyanogen bromide, reduction of the latter with potassium iodide and titration of the liberated iodine with sodium thiosulfate solution.

MINERAL RESOURCES

(Continued from p. 39 A)

The Northumberland and Durham Coal Seams. J. H. Jones. (Institute of Fuel War-Time Bulletin, 1945, Feb., pp. 89–102). Descriptions are given of the geology, nomenclature and some of the properties of the coal seams of Northumberland and Durham. The information is based on a report resulting from the Fuel Research Survey carried out by the Department of Scientific and Industrial Research.

Vanadium on the West Coast of British Columbia. H. C. Gunning and D. Carlisle. (Transactions of the Canadian Institute of Mining and Metallurgy, 1944, vol. 47, pp. 415–423). A description is given of the geology of some small vanadium-bearing deposits on the west coast of British Columbia about 100 miles north of Vancouver.

Soviet Iron-Ore Supplies in Wartime. V. Rikman. (Iron and Coal Trades Review, 1944, vol. 149, Dec. 29, pp. 979–981). A brief account is given of the measures taken in the Soviet Union to ensure adequate supplies of iron ore in war-time when the Krivoi Rog basin was occupied by the enemy. Some particulars of the ores obtained in the Urals and in the Shorian highlands of western Siberia are given.

Cerro de Mercado—Mexico's Iron Mountain. A. T. Benitez. (Engineering and Mining Journal, 1944, vol. 145, Sept., pp. 88–89). A brief description of Mexico's Iron Mountain is given. This a hill 630 ft. high situated $1\frac{1}{4}$ miles north of Durango. The amount of iron ore is estimated at 76 million tons. From 1936 to 1942 the average annual output of ore was 128,000 tons.

REFRACTORY MATERIALS

(Continued from pp. 91 A–92 A)

The Durability of Refractories. W. J. Rees. (Journal of the Birmingham Metallurgical Society, 1944, vol. 24, June, pp. 102–117). Factors affecting the life of refractory bricks and cements are discussed. Firebricks which contain any free iron oxide disintegrate rapidly if used in a furnace where carbon monoxide is present. If the brick is burnt under reducing conditions at the peak temperature of the kiln the oxide combines with the clay substance to form a complex silicate which is unaffected by carbon monoxide. Furnace walls have failed recently when the fuel has been changed from ordinary fuel oil to crude pitch oil; this may be due to ash which is rather high in iron oxide settling on the wall or to faulty burner control causing the oil to be carried forward so as to burn on the face of the brick.

The Properties of Refractory Materials and Their Significance to Fuel Economy. A. T. Green and A. E. Dodd. (Journal of the Institute of Fuel, 1945, vol. 18, Feb., pp. 74–79, 90). An attempt is made to give a balanced view of the close relationship between the two technologies of fuel and refractories by presenting and discussing data on the thermal conductivity, heat-storage capacity, bulk-density and strength properties of refractory bricks.

Failures of the Refractory Linings of Blast-Furnaces and the Ramming of Carbon Aggregates. R. Klesper. (Stahl und Eisen, 1944, vol. 64,

Dec. 7, pp. 774–781). Some examples of the failure of blast-furnace linings are described and the use of carbon blocks and carbon ramming materials is discussed. When lead-bearing iron ores are smelted it has been said that the lead cannot penetrate a rammed carbon hearth and it remains an open question whether such a hearth can be used when there is lead in the burden. Crushed coke mixed with tar should be prepared for ramming by heating to 100° C. on an iron plate heated with steam or by a coke-oven gas flame. On drilling the shell plate round a carbon-lined hearth after two years working, a space 40 to 60 mm. wide was found between the back of the rammed lining and the plate, and the lining itself had vertical cracks in it which were full of loose carbon powder. It is probable that starting up the furnace too quickly was the initial cause of this. The correct preheating of the mix is very important because, if it is made too hot, the heavier hydrocarbons in the tar are cracked and its binding properties are destroyed. Tests were made by building up blocks from layers of ramming material with and without a coating of tar between the layers; after curing, the tar coating became a thin porous layer of loose carbon powder, whilst the blocks made without coatings of tar were hard and homogeneous throughout; the roughening of the top of one layer before applying the next layer, even after 24 hr. ageing, was quite sufficient to ensure complete fusion of the joint during curing. The requirements for the protective brickwork in the case of a rammed carbon hearth are much more exacting than when the hearth is made of carbon blocks; in the former case the wall should be at least 250 mm. thick.

Significance of Navy Spalling Test. W. T. Tredennick and J. F. Kelly. (Bulletin of the American Ceramic Society, 1944, vol. 23, Dec. 15, pp. 460 463). Details are given of a modification of the A.S.T.M. panel spalling test for refractories which has been developed by the United States Navy. As a result of the adoption of this test the spalling of refractories in naval boilers has been reduced from a major to a minor cause of failure.

Reheat Tests on Fire-Clay and Silica Refractories. E. C. Petrie and C. P. Walters. (Bulletin of the American Ceramic Society, 1944, vol. 23, Dec. 15, pp. 464–468). The changes in properties of fireclay and silica bricks fired at 1205°, 1260° and 1315° C. after reheating at 1400° C. for 5 hr. were studied. Of the bricks showing secondary expansion in the reheating test those made from a stiff mix and de-aired before firing had the greatest expansion, hand-made bricks somewhat less, and dry-pressed bricks the least expansion. The bricks fired at the highest temperature expanded the least. The reheating test does not always indicate which super-duty bricks showing no secondary expansion have been underfired and which have not.

The Flow of Heat through Granular Material. A. L. Waddams. (Journal of the Society of Chemical Industry, 1944, vol. 63, Nov., pp. 337–340). The method of measurement of the apparent thermal conductivity of granular materials, which involves unsteady heat flow from the surface to the axis of a cylinder of the material, is tested, and a modified experimental technique evolved. The purpose of the new modification is to provide a simple and reliable means of measuring the thermal conductivities of such materials. As examples, a number of experimental determinations are described and the results discussed in relation to the physical characteristics of the materials under test.

FUEL

(Continued from p. 93 A)

Induction Burners for Blast-Furnace Gas, Producer Gas and Coke-Oven Gas. G. Neumann. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, May–June, pp. 237–246). Induction gas burners have previously been described by W. Heiligenstaedt (*see* Journ. I. and S.I., 1943, No. II., p. 35 A). In the present mathematical treatise curves and equations are developed showing the relationship between gas and air pressures at the burner and the furnace pressure, and examples are given of their application to determine the dimensions of burners to meet certain requirements when the fuel consists of blast-furnace gas, producer gas or coke-oven gas mixed with cold air.

Flocculation and Flotation Principles in the Recovery of Low-Grade Fuels. J. O. Samuels. (Institute of Fuel War-Time Bulletin, 1945, Feb., pp. 103–111). An important problem in the fuel economy of Great Britain is the utilisation of between 8 and 10 million tons of fine coal separated from the larger coal. The main methods of attack since about 1935 (apart from dry screening) have been flocculation and flotation. A comprehensive review of these two systems is presented.

Coal Preparation Plant at Waterloo Main Colliery. A. Holdsworth. (National Association of Colliery Managers : Iron and Coal Trades Review, 1945, vol. 150, Jan. 19, pp. 87–90 ; Jan. 26, pp. 125–128 ; Feb. 2, pp. 166–169). A description is given of the coal screening and washing plant, incorporating a Rhéolaveur washer, which was completed in 1942 at the Waterloo Main Colliery on the South Yorkshire coalfield.

The Selection of Coals for Carbonization. B. P. Mulcahy. (American Institute of Mining and Metallurgical Engineers : Blast Furnace and Steel Plant, 1944, vol. 32, Oct., pp. 1198–1204 ; Nov., pp. 1322–1327). Data showing the effect of fineness of grinding, bulk density, volatile matter content, coking time, oven temperature and degree of oxidation of the coking coal on the properties of the coke produced are presented and discussed. Coal low in volatile matter oxidises during storage much more rapidly than one high in volatile matter and the oxidation is very detrimental to the quality of the coke produced.

The Molecular Nature of Coking Coal Bitumens. H. E. Blayden, J. Gibson and H. L. Riley. (Institute of Fuel War-Time Bulletin, 1945, Feb., pp. 117–129). Earlier X-ray investigations have shown that bituminous coals can be conceived as being made of an intimate mixture of two types of molecular structure, namely : (1) The ulmin, humin or residual coal which is insoluble in organic solvents and on carbonisation gives products containing crystallites the *c* dimension (a co-ordinate used in the measurement of crystals) of which shows little or no variation over a wide range of carbonising temperature ; and (2) the soluble fraction, which is responsible for a pronounced increase in the *c* dimension as the carbonisation temperature is increased up to about 550° C. Evidence obtained by X-rays is put forward which supports the authors' views on the ulmin-bitumen character of bituminous coal.

Cleaning Producer Gas. G. Dougill. (Institution of Chemical Engineers : Iron and Coal Trades Review, 1945, vol. 150, Jan. 26, pp. 123–124). An account is given of difficulties which arose through the accumulation of deposit in the regenerators in which producer gas for heating coke-ovens was preheated, and how they were overcome. The solids were found to be coming from chlorides in the coke, and not from the scrubber water ; they also contained pitchy matter and ash. Experiments on a

small scale with an electrostatic precipitator led to the design of a full-scale precipitator for cleaning the gas before it entered the regenerators.

A Comparison of Different Designs of Cyclone. W. vom Hofé. (Stahl und Eisen, 1944, vol. 64, Dec. 21, pp. 798–801). The results of some full-scale trials in which three designs of cyclone dust-catchers were tested in blast-furnace-gas mains are presented and discussed. All three designs gave good results. The cleaning efficiency depended on the dust content of the dirty gas and the cyclone design. In all the tests the dust content of the cleaned gas was between 1 and 3 g. per cu. m. at N.T.P. and was independent of the initial dust content. There was least pressure loss with the long and narrow type of cyclone.

Purging Operations and Controls. H. E. Ferguson. (Blast Furnace and Coke Association of the Chicago District: Blast Furnace and Steel Plant, 1944, vol. 32, Nov., pp. 1334–1340, 1373). Factors affecting the speed and efficiency with which gas-holders and pipe-lines can be purged are discussed and a detailed description is given of the purging of a large water-sealed gas-holder and of a blast-furnace-gas system for heating coke-ovens.

PRODUCTION OF IRON

(Continued from pp. 93 A–94 A)

Swedish Iron Production after the First World War and Now. (Teknisk Tidskrift, 1945, vol. 75, Jan. 13, pp. 39–40). The commercial aspects of the Swedish iron and steel industry in the 1920 and 1930 decades are compared and the problems with which it is likely to be faced at the end of the present war are discussed.

Modern Trends in Blast Furnace Design. F. Yanecek. (Iron and Steel Engineer, 1944, vol. 21, Dec. pp. 37–40). The decrease in the number of blast-furnaces in the U.S.A. and the increase in their size and production during the last 25 years are described. Several recent developments in blast-furnace construction, such as electrically driven clay guns, and automatic stock-line recorders, are mentioned.

Miniature Smelting: A Working Model Blast-Furnace. G. H. Goodwin. (Iron and Steel, 1945, vol. 18, Jan., pp. 6, 17). A description is given of a miniature blast-furnace which was constructed for research purposes for the Iron and Steel Industrial Research Council of the British Iron and Steel Federation. The furnace hearth is 21 in. in dia. and the height is 10 ft. It was charged with Northamptonshire ore, and while in use for 13 weeks produced 60 tons of iron, consuming 36–45 cwt. of coke per ton.

Blast-Furnace Tuyeres. J. B. Fortune. (Iron and Steel, 1944, vol. 17, Dec., pp. 704–708; 1945, vol. 18, Jan., pp. 7–9). The design of tuyeres for blast-furnaces, causes of failure and methods of obtaining the maximum service life are discussed.

Investigations on the Recovery of Vanadium from Slags. J. Klärding. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Jan.–Feb., pp. 153–157). The influence of the form of pretreatment and of the chemical composition of vanadium-bearing slags on the facility with which the vanadium can be leached out with acids was investigated. The amount leached out was increased by raising the temperature of the roasting pretreatment and by increasing the proportion of silica in the slag mix before roasting.

The Relationship between the Production of Iron Powders and the Properties of the Compacts. W. Dawihl and U. Schmidt. (Stahl und Eisen, 1945, vol. 65, Jan. 4, pp. 9–14). The results of investigations

of the porosity, strength and other properties of powder-metallurgy products made of iron powders produced by several different processes are discussed and compared. The porosity and structure were examined by micrographs at high magnification, and microhardness tests were made. The effects of the manufacturing process are still quite marked, even after sintering at 1450° C. The pores in parts made from powders of iron that has been melted are uniformly distributed, whilst the distribution is not uniform if the iron has not been melted (*e.g.*, powder from sponge iron). If the powder has been subjected to severe deformation (*e.g.*, when made by the cold-stamping of sheet iron) this has a detrimental effect on the sintering properties and on the growth of the ferrite. The micro-hardness readings indicated that the uniting of the crystals gave rise to additional stresses called "sintering stresses;" these stresses were probably the cause of the changes in microhardness with the method of producing the powder. Additional time and power consumption in making the powder can in some cases be compensated by simplification in the finishing processes; for instance, when a compact is made from a powder produced by directing two streams of particles one against the other, sintering only is required to produce a soft and easily deformed part, whereas, with powders made by other methods, forging and recrystallisation are necessary to produce the same deformability.

Powder Metallurgy. E. E. Schumacher and A. G. Souden. (Metals and Alloys, 1944, vol. 20, Nov., pp. 1327–1339). A comprehensive review of the powder-metallurgy process is presented, with its advantages and limitations. A bibliography of 68 references is appended.

Iron in Powder Metallurgy. R. J. Traill. (Transactions of the Canadian Institute of Mining and Metallurgy, 1944, vol. 47, Dec., pp. 490–500). The economic aspects of iron powder metallurgy are reviewed. The advantages and disadvantages of using powders prepared from electrolytic iron, by the gaseous reduction of iron oxide and from mill-scale, are discussed. The results of tests on bars containing 0·5% of carbon, all pressed at 60,000 lb. per sq. in., indicate that those made from electrolytic iron have greater tensile strength, elongation and density than those made from hydrogen-reduced iron.

FOUNDRY PRACTICE

(Continued from pp. 94 A–98 A)

Cast Iron Metallurgy. J. E. Hurst. (Staffordshire Iron and Steel Institute: Iron and Steel, 1945, vol. 18, Jan., pp. 14–17; Feb., pp. 53–54). **Research Methods in Cast Iron Metallurgy.** J. E. Hurst. (Metal Treatment, 1944, vol. 11, Winter Issue, pp. 219–228). **Progress in Research Methods in the Metallurgy of Cast Iron with Special Reference to Gases in Cast Iron.** J. E. Hurst. (Metallurgia, 1944, vol. 31, Dec., pp. 92–97). A foundry producing continuously a particular casting by uniform mechanised methods involving modern methods of sand preparation and melting control experienced epidemics of porosity at irregular intervals. An investigation of the causes is described. Evidence was produced in support of the following conclusions: (1) The differences in the characteristics of the castings were associated with the dissolved gas content; and (2) the differences in the characteristics of pig-irons associated with the dissolved gas content survive the remelting and casting procedure in the foundry.

Dry Foundry Ladles through Eliminating Boiling Produce More Salable Castings. C. E. Bales and F. McCarthy. (American Foundryman, 1944,

vol. 6, Nov., pp. 2–5). The advisability of thoroughly drying and pre-heating foundry ladles is stressed and some designs of gas and oil burners for this purpose are described.

The Making of High-Duty Iron Castings to Specification. E. Hunter. (Institution of Engineering Inspection: Foundry Trade Journal, 1945, vol. 75, Feb. 1, pp. 95–97). Recent progress in iron-founding has satisfied new standard specifications for cast-iron. The organisation and operation of technical control are described and their importance is emphasised. Some modern applications in which irons have replaced alloy steels are also mentioned.

Moulding Sands of South Australia. H. A. Stephens. (Australian Institute of Metals: Australasian Engineer Science Sheet, 1944, Nov. 7, pp. 22–29). Various features of South Australian moulding sands and moulding practice are discussed. South Australian moulding sands are found to be poor, but moulding practice can be improved by the wider use of synthetic sands and by increased attention to the preparation of the sand.

Dust in Steel Foundries. (Ministry of Labour and National Service: H.M. Stationery Office, 1944). This pamphlet contains the First Report of the Committee on Dust in Steel Foundries, which was appointed by H.M. Chief Inspector of Factories. The various operations in a foundry which give rise to silica dust are described, and recommendations are made on means of protecting foundry workers from the risk of silicosis. Particular attention is given to sand-blasting.

Patternmaking. W. C. Perry. (Institute of British Foundrymen: Foundry Trade Journal, 1945, vol. 75, Feb. 1, pp. 85–90). An illustrated description is given of modern patterns which had been mass-produced during the war to satisfy the greatly increased demand. A discussion on contraction allowances shows that little uniformity has been attained throughout the country. It has also been found impossible so far to establish a standard taper for patterns. Metal spraying of patterns proved very useful with damaged or over-sized patterns.

The Production of Builders' Castings. C. Gillespie. (Institute of British Foundrymen: Foundry Trade Journal, 1945, vol. 75, Jan. 25, pp. 65–68). The author deals with plant for the production of general repetition castings, explaining the best use of suitable pattern equipment, and the most advantageous placing of runners. Special attention is paid to moulding boxes and moulding sand, and the importance of moulding machines to increase production and ease the burden of the worker is pointed out.

Runners and Risers. A. Roberts. (Institute of Australian Foundrymen: Australasian Engineer Science Sheet, 1944, Dec. 7, pp. 2–5). Recommendations are made on the dimensions of sprues, runners, gates and risers and on their best positions relative to the casting.

Precision Casting by the Lost-Wax Process. A. Dunlop. (Institute of British Foundrymen: Foundry Trade Journal, 1945, vol. 75, Feb. 8, pp. 107–116, 118; Metal Treatment, 1944, vol. 11, Winter Issue, pp. 247–255). The technique of mould preparation and casting known as the lost-wax process is described and many examples of small articles made by it of steel, high-nickel alloys, bronze and brass are illustrated. The process was discussed in a recent paper on precision casting (*see* Journ. I. and S.I., 1944, No. I., p. 193 A).

Methods and Problems Indigenous to a General Engineering Iron Foundry. W. Montgomery and J. Doig, jun. (Institute of British Foundrymen: Foundry Trade Journal, 1945, vol. 75, Jan. 11, pp. 25–29, 33; Jan. 18, pp. 49–53; Jan. 25, pp. 71–73). Moulding work in a jobbing foundry is described with numerous examples of the techniques employed

to avoid the necessity of making special patterns for single castings. These techniques include: (*a*) Using "shell" patterns for special pipe bends; (*b*) making pipe core irons so that they can be easily removed from cores in pipes which have branches and bends; (*c*) making loam patterns; (*d*) sweeping green sand moulds; and (*e*) "salt box" casting in green sand. Particulars of the melting practice with two cupolas in conjunction with receivers are given.

Unique Centrifugal Steel Casting Method. G. E. Stedman. (Metals and Alloys, 1944, vol. 20, Nov., pp. 1311–1315). A modern centrifugal steel casting plant is described which produces castings spun on a vertical axis, employing mainly the dry-sand practice. A description of the centrifugal machines used and of the safety measures applied is given. The advantages of centrifugal casting of sheaves as compared with casting by the static method are pointed out. Data on spinning speeds and practice are also given.

Plant Layout for High-Frequency Melting. G. F. Applegate. (Foundry, 1944, vol. 72, Dec., pp. 84–86, 202–206). High-frequency furnace installations and plant lay-out for melting ferrous and non-ferrous metals in foundries are discussed and described with numerous illustrations and plans.

Redesigning a Foundry for Maximum Efficiency. R. W. Bierwagen. (Foundry, 1944, vol. 72, Dec., pp. 74–78, 106). The mechanisation of the large jobbing foundry of the Westinghouse Electric and Manufacturing Co., Pittsburgh, is described and illustrated.

A Graphic Method for Determining Rates of Solidification. M. Hampl and V. Vodička. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Jan.–Feb., pp. 185–191). A rather complicated equation has been developed by Gröber for calculating solidification rates and the depth to which a substance will solidify in a given time under given conditions. In this paper curves are presented which enable the equation to be solved in a few minutes, and examples are given of their application to determine the freezing rates of water and of liquid steel, as well as the progress of solidification in steel when in contact with a wall of infinite thickness of either cast iron or fireclay.

How Fast Do Metals Freeze? H. A. Schwartz. (Foundry, 1944, vol. 72, Dec., pp. 80–81, 236–244). Previous papers on the rate of solidification of castings are referred to, particularly that of N. Chvorinov (*see* Journ. I. and S.I., 1939, No. II., p. 75 A), and modifications of Chvorinov's formula are presented. The differences in the rates of freezing noted by investigators are probably due to changes in the apparent thermal conductivity of the moulding sand caused by radiation from grain to grain.

PRODUCTION OF STEEL

(Continued from pp. 98 A–101 A)

Slag Systems. Viscosity Determinations of the System MnO–Al_2O_3–SiO_2. Helen Towers and J. M. Gworek. (Journal of the West of Scotland Iron and Steel Institute, 1943–44, vol. 51, pp. 123–132). An apparatus for determining the viscosity of acid open-hearth slags is described. The first series of experiments was carried out on the binary system MnO–SiO_2. A procedure was adopted to keep the MnO/SiO_2 ratio constant and add increasing amounts of Al_2O_3. All the evidence pointed to the fact that there is a range of liquid immiscibility in the MnO–Al_2O_3–SiO_2 system near the rhodonite composition. In this region viscosity determinations have

no real meaning. The effect of Al_2O_3 is to cause an initial increase in viscosity with additions up to 10%, but with further increases up to 15% the viscosity is reduced to the order of that given by slags without alumina additions. A number of viscosity-temperature curves for different slags are presented.

New Crane Hoist System Provides Maximum Speed Combined with Maximum Safety. M. A. Whiting. (Blast Furnace and Steel Plant, 1944, vol. 32, Nov., pp. 1317–1321, 1327). A design for an electric crane motor and control is described which embodies an unusual form of the Ward-Leonard system. A motor-generator is provided of a capacity required by the hoist motor only. Where the incoming power is alternating current, the motor-generator consists of a standard squirrel-cage induction motor, a D.-C. generator of typical design, a standard constant-voltage exciter and a small machine known as a cross-flux exciter.

Steel Ingots. E. Barber. (British Steelmaker, 1945, vol. 11, Jan., pp. 14–18; Feb., pp. 76–80). The pouring, cooling and solidification of steel ingots are described. It is shown how to obtain a sound ingot with a minimum of wastage by preventing excessive segregation and by using a "dozzle" to prevent pipe. The essential factors in sound ingots are: Proper deoxidation, well-designed moulds, and cooling in a vertical position.

The Influence of Various Metallurgical Factors on Ingot-Mould Life. B. Körös. (Iron and Steel Institute, 1945, Translation Series, No. 211). An English translation is presented of a paper which appeared in Stahl und Eisen, 1944, vol. 64, Mar. 9, pp. 159–164 (*see* Journ. I. and S.I., 1944, No. II., p. 148 A).

FORGING, STAMPING AND DRAWING

(Continued from p. 102 A

Forging. (British Steelmaker, 1945, vol. 11, Jan., pp. 22–27; Feb. pp. 68–73). A detailed description is given of structural and other changes in steel which take place during forging. The effects of overheating and burning are pointed out, and also the importance of the fibre of a forging. Several instructive experiments are mentioned to show the results of up-ending a slug. A discussion of the mechanical properties of forgings follows, with special reference to impact tests.

Practical Problems of Light Presswork Production. J. A. Grainger. (Sheet Metal Industries, 1944, vol. 20, Aug., pp. 1367–1370; Sept., pp. 1553–1560; Oct., pp. 1735–1738; Nov., pp. 1921–1928; Dec., pp. 2105–2112; 1945, vol. 21, Jan., pp. 81–84, 88). The design of dies and machines for the mass production of parts out of light-gauge sheet metal is discussed with examples of some of the calculations involved and descriptions of how some practical problems have been solved.

Chain-making Equipment: Development and Processes Described. (Wire Industry, 1945, vol. 12, Jan., pp. 27–29). An illustrated description is given of a number of machines for the bending, lacing, welding and twisting of chains between $\frac{3}{16}$ and $\frac{1}{2}$ in. diameter. The material used is mainly mild steel and Armco wire.

Deep-Drawing Oil Sump Pans. G. W. Birdsall. (Steel, 1944, vol. 115, Nov. 27, pp. 74–77, 110–111). The method employed in producing an oil sump pan $5\frac{1}{2}$ in. deep is described. It was found possible by improving the die design and the lubricants to draw the $5\frac{1}{2}$-in. deep pan in a single operation. The operations necessary to produce the finished article, *i.e.*, welding, lead dipping, trimming, &c., are also described.

ROLLING-MILL PRACTICE

(Continued from pp. 103 A–104 A)

The Metadyne Electrical Control System Applied to Speed Control on Rolling Mills. (Sheet Metal Industries, 1945, vol. 21, Jan., pp. 72–74). The "Metadyne" is a D.C. generator of special design developed by the Metropolitan-Vickers Electrical Co. A distinctive feature is that it has an extra set of brushes which are short-circuited. Its principles and application for control purposes in rolling mills are described.

Power Circuit Breakers. H. F. Hentschel and E. W. Boehne. (Iron and Steel Engineer, 1944, vol. 21, Dec., pp. 74–82). Power air circuit-breakers are now being used on high-voltage circuits, and they offer to the steel industry a safe method of current interruption functioning in a manner which solves some problems of maintenance. The change-over from oil to air as the insulating medium is discussed and several types of installation for iron and steel works are described.

Lubrication with Engine-Oil Emulsions, High-Pressure-Compressor-Oil Emulsions, Cylinder-Oil Emulsions and Emulsified Greases. W. Reuschle. (Stahl und Eisen, 1945, vol. 65, Jan. 4, pp. 14–19). In order to economise in the consumption of lubricating oil, the use of 50/50 oil-water emulsions has been developed considerably in Germany. The preparation of these emulsions for the lubrication of various types of heavy machinery and their reconditioning after use are described.

The Manufacture of Grease. L. P. Lochridge. (Iron and Steel Engineer, 1944, vol. 21, Dec., pp. 86–89). The composition and manufacture of lubricating greases are described. Before a grease is considered acceptable it has to be tested thoroughly as to its chemical and physical properties and also as to its mechanical properties in actual service. Apparatus and methods for testing are shown.

Water and Composition Roll Neck Bearings. J. P. Queeney. (Iron and Steel Engineer, 1944, vol. 21, Dec., pp. 83–85). The effects of the temperature, degree of purity and amount of the lubricating water used for phenolic plastic bearings for rolling-mills are discussed. The results of tests to study these effects are presented. The following conclusions were reached: (1) The water must be clean, cool and free from salts and acids likely to cause corrosion; (2) an ample flow must be maintained; a minimum of 0·6 gal. per min. per sq. ft. of bearing surface is suggested; (3) the water must be applied so as to ensure the maximum area of contact; (4) the temperature of the water should not exceed 45° C.; (5) the roll necks, including the fillets, must be ground smooth; (6) raising the temperature of the water decreases its viscosity and considerably reduces the coefficient of friction, but above a certain temperature the water film breaks down and hot spots develop in the bearing; and (7) injecting a small amount of grease into the bearing at infrequent intervals tends to stabilise the coefficient of friction.

The Technique of Section Rolling. B. D. N. Tate. (Sheet Metal Industries, 1945, vol. 21, Jan., pp. 108–110). A description is given of a tube-rolling machine which produces spiral coils of 3-in. dia. stainless steel tubing from flat 20-gauge strip. The coil has a mean diameter of 41½ in., and from this rings can be cut to make exhaust manifolds for aeroplane engines.

Developments in the Lay-out of Modern Wire Mills. (Wire Industry, 1945, vol. 12, Jan., pp. 35–36). The lay-out of a modern wire mill, consisting of a five-stand continuous 13-in. roughing group, a five-stand 13-in. zig-zag train, an eight-stand and an 11-in. continuous finishing group is

described. Starting with billets of 2–2·5 in. square, wires of various diameter are produced, the finished products having a diameter ranging from 0·2 to 0·5 in.

Methods of Increasing the Production and Improving the Operation of a Cold-Rolling Mill. T. Thiemann. (Stahl und Eisen, 1944, vol. 64, Dec. 7, pp. 763–774). Numerous improvements to the equipment of a mill for cold-rolling, annealing and pickling coils of strip are described and illustrated.

Shearing Flat Rolled Steel. F. E. Flynn and D. A. MacArthur. (Iron and Steel Engineer, 1944, vol. 21, Dec., pp. 41–68). A comprehensive discussion of the theory and practice of shearing, with particular reference to strip mills, is presented. Formulæ are developed for calculating the power requirements of shears for cutting plates and sheets and for slitting and trimming strip, and the results of many shearing tests in which the current consumption was measured are given. Many modern slitting and trimming machines are described and illustrated.

HEAT TREATMENT

(Continued from pp. 104 A–107 A)

Spheroidising. J. G. Ritchie. (Australian Institute of Metals: Australasian Engineer Science Sheet, 1944, vol. 44, Nov. 7, pp. 25–35). A short explanation of the term spheroidising is given and the means by which a spheroidised structure can be obtained are pointed out. These include: (1) Sub-critical annealing, which is largely a slow diffusion process, is carried out to increase the ductility of medium carbon steels. The time required is extremely long, but can be shortened by a hardening pretreatment. Spheroidisation is also accelerated by previous cold-work. (2) Spheroidising by transformation from austentite, which is based on the direct transformation of austenite to spheroidite. Low-carbon steels show a tendency to transform to pearlite rather than to spheroidite if the austenitising temperature is raised too far above the Ac_1 point. Slow cooling cycles for plain carbon tool steels are described in detail. (3) Spheroidising by isothermal transformation, where the sub-critical temperature must be not more than 45° C. below the Ac_1 point in order to obtain a spheroidal product. These methods can also be applied successfully to alloy steels. The elimination of decarburisation constitutes a difficult problem in this treatment. A certain amount of graphitisation of cementite often occurs. Spheroidising is carried out to improve cold-working properties, machinability and the structure after a final hardening. Some of the faulty structures likely to be produced are also discussed. A bibliography of 34 references is appended.

The Isothermal Treatment of Alloy Tool Steel. A. N. Alimov, N. N. Lipchin and N. F. Sivkov. (Iron and Steel Institute, 1945, Translation Series, No. 208). A translation is presented of a paper which appeared in Katshestvennaia Stal, 1937, No. 2, pp. 37–40; this is an account of tests made on alloy steels for forging into tools with a view to reducing the time required for heat-treatment. The five steels used were: (1) A 12%-chromium steel; (2) a low-alloy chromium-nickel-molybdenum steel; (3) a 1·20%-chromium 1·70%-tungsten steel; (4) an 8·40%-tungsten 2·53%-chromium 0·33%-vanadium steel; and (5) a 17·5%-tungsten 3·90%-chromium steel. Satisfactory heat-treatments were developed which involved holding at 860–900° C. for 1–1½ hr. followed by holding at a subcritical temperature for not more than 4 hr. The total

heat-treatment time was reduced by about 60% as compared with the former methods.

A Producer Gas Fired Furnace Installation for the Heat Treatment of Alloy and Special Steel Bars. (Wire Industry, 1944, vol. 11, Dec., pp. 611–612). **Producer Gas Fired Furnace Installation for Heat-Treating Alloy-Steel Bars.** (Machinery, 1945, vol. 66, Jan. 18, pp. 73–74). **A Producer-Gas Fired Heat Treatment Furnace Installation for Alloy and Special Steel Bars.** (Sheet Metal Industries, 1945, vol. 21, Jan., pp. 126–128). **Producer Gas Fired Installation for Heat-Treating Bars and Sections.** (Iron and Steel, 1945, vol. 18, Jan., pp. 33–35). **Producer Gas Fired Furnaces.** (Automobile Engineer, 1944, vol. 34, Dec., pp. 539–540). A description is given of the heat-treatment plant of the Hallamshire Steel and File Co., Ltd., Sheffield (*see* p. 78 A).

Electric Induction Heating. (Steel, 1944, vol. 115, Dec., pp. 116–118). A recently built laboratory at the works of the Ohio Crankshaft Co., devoted solely to the advancement of metallurgical applications of induction heating, is described.

Magnetic Devices Control Speed of Roller-Hearth Furnaces. G. W. Heumann. (Machinist, 1945, vol. 88, Jan. 27, pp. 106–109). Descriptions are given of magnetic devices which can be fitted to roller-hearth heat-treatment furnaces so as to provide fully automatic operation of the vestibule doors at the charging end, the roller-hearth drive and the discharging door.

The Structure of the Nitride Case on an Austenitic Steel. W. Betteridge. (Journal of Scientific Instruments, 1945, vol. 22, Feb., pp. 28–29). The nitriding of a steel containing nickel 12%, manganese 6%, chromium 4·5% and vanadium 0·3% was not successful by ordinary methods, as the hardened surface was spoiled by flaking of the outer layers of the case. Crystallographic examination showed that this was due to the presence of Fe_4N, which, in addition to flaking away, was softer than the underlying layers of γ solid solution. If the case could be obtained with a nitrogen content at the surface just too small to allow the formation of Fe_4N, flaking would be avoided and the hardness would still be adequate. This was achieved by holding the steel parts at 600° C. for 6 hr. after cutting off the flow of ammonia to the nitriding chamber, thus giving the nitrogen time to diffuse inwards and reduce the concentration at the surface.

Diagram of Transformation during Continuous Cooling of Steel. C. A. Liedholm. (Metal Progress, 1944, vol. 46, Nov., pp. 1097–1101). An explanation is given of the method of applying the continuous cooling transformation diagram referred to in a previous paper (*see* Journ. I. and S.I., 1944, No. II., p. 47 A) for controlling the die-quenching of hollow steel propeller blades.

The Selection of a Quenching Oil for Hardening Steel. H. Krainer and K. Swoboda. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Jan.–Feb., pp. 163–169). A. Rose has previously constructed characteristic curves for different quenching media which depict the changes in cooling velocity at the centre of a silver ball when quenched from 800° C. (*see* Journ. I. and S.I., 1939, No. II., p. 272 A). In the present paper experiments are reported in which the cooling rates at the centre of steel plates from 25 to 200 mm. thick when quenched in different oils and in water were first calculated using Rose's curves, and then checked by making hardness determinations and examining the microstructure of the quenched steels. The steels used included a chromium-molybdenum steel, a chromium-vanadium steel and two chromium-manganese steels. The calculated and experimental results were in good agreement.

WELDING AND CUTTING

(Continued from pp. 107 A–109 A)

Fabrication and Reclamation in Steel Mill Maintenance. E. W. Gruber. (Welding Journal, 1944, vol. 23, Nov., pp. 1031–1040, 1047). The maintenance department at the works of the Wheeling Steel Corporation is described. The application of welding has been so much developed in this department that practically all the works' repairs and replacements can be carried out by it. Several examples of the work done are described and illustrated.

Salvaging of Large Cast Iron Castings. H. O. Quartz. (Iron Age, 1944, vol. 154, Nov. 30, pp. 52–56). The technique of repairing large iron castings is described and some examples are illustrated. In some cases the method is to chip out porous areas and screw a number of studs about 1 in. apart into the hollow thus formed; the recess is then filled up by welding with Monel metal electrodes which are woven between the studs. The centre of the weld is always kept higher than the outside edges; after each layer is deposited it is lightly peened and brushed clean.

Shrinkage Distortion in Welding. W. Spraragen and M. A. Cordovi. (Welding Journal, 1944, vol. 23, Nov., pp. 545-S–559-S). A comprehensive review of the literature on the distortion of welded components is presented; it covers United States publications for the seven years ending December 31, 1943, and those of other countries up to December 31, 1940.

Electrode Coatings. Rapid Determination of Moisture Content. G. Haim. (Welding, 1945, vol. 13, Jan., pp. 600–604). A close control of the moisture content of welding-electrode coatings during electrode manufacture is essential. A description is given of the Brabender semi-automatic moisture tester with which the determinations can be made in 15–35 min. after taking the samples.

Proposed Recommended Practices for Resistance Welding. 1. Recommended Practice for the Spot, Seam and Pulsation Welding of Low-Carbon Steel (Including Coated Steels). 2. Recommended Practice for the Spot and Seam Welding of Stainless Steel. 3. Recommended Practice for the Flash-Butt Welding of Low & Medium Forging Strength Steels. (Welding Journal, 1944, vol. 23, Aug., pp. 713–717; Sept., pp. 812–817). The welding standards presented in these papers have been prepared by the Resistance Welding Committee of the American Welding Society; they are not yet approved, but are published so as to elicit comments.

The Flame Cutting of Steel. C. A. E. Wilkins and W. J. Currie. (Metal Treatment, 1944, vol. 11, Winter Issue, pp. 259–265, 272). The theory of oxy-acetylene cutting, the technique and the metallurgical changes in the steel during the process are reviewed. The metallurgical phase formed finally is dependent on the quality of the steel and the cutting conditions. If high-carbon and alloy steels are to be crack-free and machinable after cutting, preheating or annealing are required. Air-hardening steels require both a preheating and an annealing operation.

CLEANING AND PICKLING OF METALS

(Continued from pp. 81 A–82 A)

The Pickling of Steels. E. W. Mulcahy. (Sheet Metal Industries, 1944, vol. 20, Nov., pp. 1901–1904, 1917; Dec., pp. 2085–2093; 1945, vol. 21, Jan., pp. 61–64, 67). The factors involved in pickling-plant

design are discussed and the importance of a thorough knowledge of the properties of the materials used in plant construction is stressed. Particulars of the properties of lead, timber, rubber, bitumastic compounds, phosphor bronze, ferro-silicon alloys, Monel metal, siliceous cement and vitrified blue brickwork as materials for tanks and other pickling equipment are given. A special section deals with equipment for fume extraction.

Acid Cleaning and Quality Production. (Wire Industry, 1944, vol. 11, Nov., pp. 553–554). Brief descriptions are given of various methods of cleaning wire, including liming practice, the use of nitric acid and inhibitors, the submerged combustion process, the nascent metallic sodium process and cleaning with angular steel grit.

Liquid Honing. (Steel, 1944, vol. 115, Nov. 27, p. 100). A process of finishing and polishing metal surfaces known as "liquid honing" or "vapour blasting" is briefly described. It consists of directing a stream of a fine abrasive mixed with a chemical emulsion at high pressure against the metal surface. The mixture can be discharged from a conventional sand-blast nozzle by compressed air at a velocity of 3200 ft. per sec.

COATING OF METALS

(Continued from pp. 82 A–84 A)

Metal Finishing. H. Silman. (Sheet Metal Industries, 1944, vol. 20, July, pp. 1205–1214; Aug., pp. 1397–1402; Sept., pp. 1590, 1597–1600; Oct., pp. 1781–1785; Nov., pp. 1965–1970; Dec., pp. 2125–2135). Continuation of a series of articles (*see* Journ. I. and S.I., 1944, No. II., p. 83 A). The principles of electroplating are explained and details are given of the equipment for producing the plating current and of methods of heating, stirring and filtering the plating solution. Descriptions are given of the lay-outs of automatic plating plants and the processes for the deposition of silver, copper and nickel.

Protective Coatings on Metals. W. J. Blanch. (Australian Institute of Metals: Australasian Engineer Science Sheet, 1944, Nov. 7, pp. 11–16). Descriptions are given of methods of plating steel and cast iron with cadmium, zinc, tin, nickel and chromium, and of spraying with aluminium, zinc and tin. Some weathering and corrosion-resistance tests are also discussed.

"Ecco" Electrolytic Wire Plating Plant. (Wire Industry, 1944, vol. 11, Nov., pp. 559–561). A description is given of the "Ecco" electroplating plant for coating fine wire with nickel, copper, tin or silver. This consists of three main tanks for cleaning, etching and plating respectively. The wire enters the first tank over insulating rollers, passes to the bottom of the tank and then passes in an ascending coil round eight insulating and two conducting rollers on vertical spindles, the bearings for which are above the surface of the solution. The individual turns of the helix thus formed are kept in the correct position by combs made from glass rods. The mechanism is the same for each of the tanks, and all the rollers are positively driven by gearing on a single shaft driven by a motor at one end.

Electro-Plating on Wire. J. Kronsbein and A. Smart. (Journal of the Electrodepositors' Technical Society, 1945, vol. 20, pp. 31–38). The limitations of the original design of electroplating equipment for wire are pointed out and an illustrated description is given of the "Ecco" plant (*see* preceding abstract).

Jessop's Stainless-Clad Steel Economical for Many Purposes. W. M. Crouch, jun. (Steel Processing, 1944, vol. 30, Oct., pp. 652–655). Seven different methods of making mild-steel plates clad with stainless steel are briefly described. These are : (1) A pair of stainless steel plates, separated by an inert compound and arc-welded peripherally, is centred in an ingot mould ; mild steel is then cast round it. The composite ingot is heated to 2300° F. and rolled to twice the desired thickness. The welded edges are sheared off, and two clad plates are thus produced. (2) Pure iron is electro-deposited on the outside surfaces of two stainless-steel sheets ; the plates are placed face to face with an inert compound between them, and the pair is then sandwiched between two thick mild-steel plates ; the edges are sealed and the assembly is heated, rolled and sheared as in the first process. (3) A sheet of stainless steel is spot-welded on to a mild-steel plate with welds spaced about ¾ in. apart. (4) This is similar to (3), except that the spot welds overlap, forming continuous rows. (5) A refractory jacket is built round one face of a mild-steel slab, and stainless steel is introduced into the space by a combination of arc welding and electric-furnace melting ; an intermelting between the surface of the slab and stainless steel takes place and the resulting slab is rolled to the required thickness. (6) A mixture of ferro-alloys, nickel, iron and a suitable slag, all in granular form, is spread over one surface of a mild-steel slab ; an arc is passed over the surface, causing a fusion of the essential elements ; the composite slab is then rolled in the usual manner. (7) Overlapping beads of stainless steel are deposited over the surface of a mild-steel slab by one of the welding processes. The surface is ground and the composite slab is heated and rolled.

Practical Aspects of Hard Chrome Plating. J. L. Vaughan and I. A. Usher. (Canadian Metals and Metallurgical Industries, 1944, vol. 7, Nov., pp. 20–30, 53). A detailed description is given of the chromium-plating process used by John Inglis Co., Ltd., of Toronto, and of its application for the plating of machine-gun parts.

Electro-Plating Nickel from Low *p*H Electrolytes. E. E. Halls. (Metal Treatment, 1944, vol. 11, Winter Issue, pp. 235–243). Experience has shown that low-*p*H nickel-plating baths require a higher degree of technical control than normal baths and that they are admirable for plating wire and strip, but not so obviously advantageous for small components. A brief account of the latest practices for nickel-plating components and wire is given.

Increasing the Pliability of Hot-Dip Zinc Coatings. W. Püngel and R. Stenkhoff. (Stahl und Eisen, 1944, vol. 64, Nov. 9, pp. 720–725). Recent German investigations of methods of improving the pliability of hot-dip zinc coatings on steel sheet and wire are reported. Thick and pliable zinc coatings can be obtained on sheets if the surface is roughened by mechanical or chemical means before coating ; this can be done with emery or sand, or by pickling for a short time in nitric acid. Wire can also be given a pliable coating if all pickling salts are removed by sand-blasting before the coating process.

Continuous Electro-Zinc Plating. J. R. Erbe. (Iron Age, 1944, vol. 154, Nov. 16, pp. 70–72). An illustrated description is given of a continuous electroplating plant for steel strip which was originally designed for tinning but was altered for zinc plating. Strip up to 38 in. wide can be zinc-plated at a speed of 160 ft. per min. with a coating weight in the range 0·1–0·2 oz. of zinc per sq. ft.

Electrotinning of Strip in the Alkaline Bath. T. G. Timby. (Steel, 1944, vol. 115, Nov. 6, pp. 124–128, 176). An illustrated description is given of the continuous electrolytic tinning plant for steel strip at the works of the Jones and Laughlin Steel Corporation.

Canmakers' Wartime Problems. J. A. Stewart and R. W. Pilcher. (Chemical and Engineering News, 1944, vol. 22, Mar. 25, pp. 422–434). A review is presented of changes made in the American can-making industry in the period 1940–1943, with special reference to the preservation of tin. Several tables are presented in which the losses of vacuum of cans containing various foods after storage for long periods at 70° and 100° F. are compared so as to demonstrate the relative merits of hot-dip and electrolytic coatings with and without an enamel finish on the inside.

An Investigation on the Silver Plating of Steel. J. M. Sprague. (Journal of the Electrodepositors' Technical Society, 1945, vol. 20, pp. 39–46). An investigation was carried out on methods of obtaining silver plate up to 0·002 in. thick on a chromium-molybdenum steel, the coating to withstand a temperature of 400° C. for at least 1 hr. without blistering or apparent decrease in adhesion. Direct deposition of silver on this steel, or silver plating on an undercoat of cadmium or of nickel, was not found to be successful. The most satisfactory procedure consists essentially of anodic etching in sulphuric acid and of the subsequent provision of undercoats of nickel followed by copper before striking in silver and silver-plating.

Lead Coatings on Steel. H. A. Knight. (Metals and Alloys, 1944, vol. 20, Nov., pp. 1296–1301). The application of lead coatings on steel has found a wide field in recent years. Pure lead is sometimes used, but more often it is alloyed with small amounts of tin, antimony, zinc or silver. The main defects in lead coatings are pinholes, but it is found that a thickness of at least 0·0005 in. assures adequate corrosion-resistance in spite of pinholes, because of the self-sealing qualities of lead. Fluxes and the soldering of lead-coated steels and some fabrication factors are also described, and an account is given of electrolytic methods of depositing lead.

New Lead Plating Process. A. G. Gray. (Steel, 1944, vol. 115, Nov. 27, pp. 78–80, 114–120). Lead forms a very suitable protective coating to steel, it is cheap, easily applied, and supplies of lead are less restricted than those of most other plating metals. A new lead-plating process is described; it is based on commercially produced sulphamic acid. An aqueous solution of lead sulphamate is used in the plating process. Operating conditions and applications are discussed.

Mechanics of Enamel Adherence: XVI. Influence of Manganese Dioxide on Metal Precipitation at Ground-Coat/Iron Interface. R. M. King. (Journal of the American Ceramic Society, 1944, vol. 27, Nov., pp. 350–351). Continuation of a series of articles (*see* Journ. I. and S.I., 1944, No. I., p. 99 A). Experiments are described which show the influence of manganese dioxide on the phenomenon of metal precipitation in ground coats of enamel on sheet iron. The presence of manganese oxide alone in ground coats did not bring about a detectable precipitation of metal even when available in double the minimum effective percentages of cobalt and nickel oxides.

Tests Concerning the Metal and the Preparation of Metal Surfaces for Porcelain Enameling. W. W. Higgins. (Bulletin of the American Ceramic Society, 1944, vol. 23, Dec. 15, pp. 473–475). A review of the literature on the preparation of metal surfaces for porcelain enamelling is presented.

Review of Tests for Fineness and Consistency of Enamel Slips. C. M. Andrews. (Bulletin of the American Ceramic Society, 1944, vol. 23, Dec. 15, pp. 475–477). Methods of testing the fineness and consistency of enamel slips are reviewed and their utility for plant control is discussed.

Review of the Test Methods Used to Determine Some of the Physical Characteristics of Porcelain Enamels when Applied to Iron and Steel. F. A. Petersen. (Bulletin of the American Ceramic Society, 1944, vol. 23,

Dec. 15, pp. 477–480). Adherence, abrasion, impact and acid-resistance tests for enamel coats on iron and steel are reviewed.

Test Methods Used in Studying the Properties of Enamel Frits. R. L. Cook. (Bulletin of the American Ceramic Society, 1944, vol. 23, Dec. 15, pp. 480–482). Tests for determining the softening point, viscosity, surface tension, strength and elasticity of enamel frits are described.

Dip Coating Made to Produce More Uniform Results by Electrostatic Detearing. (Steel, 1944, vol. 115, Oct. 23, pp. 74–76, 122). A brief description is given of the equipment and processes developed for coating 75-mm. steel cartridge cases with a synthetic-resin varnish. The equipment includes a cleaning, dipping and electrostatic "detearing" plant. Detearing is the removal of drops of surplus varnish from the lowest point of the case (*see* Journ. I. and S.I., 1944, No. I., p. 100 A).

Plastic Coatings for Metals. C. D. Townsend. (Iron Age, 1944, vol. 154, Nov. 2, pp. 56–57). A brief survey is made of the properties of the alkyd, melamine, acrylic and phenolic resins which can be used for coating metals.

Painting Steel. W. C. Porter. (National Bureau of Standards, 1944, Building Materials and Structures, Report BMS 102). Extensive laboratory and field-exposure tests of protective painting schemes by the National Bureau of Standards have been completed. (For a previous report on this work see Journ. I. and S.I., 1940, No. II., p. 191 A). The laboratory tests were of three types: (1) Continuous exposure to a fine spray of 20%-sodium-chloride solution; (2) continuous exposure to ultra-violet light, but wetted and chilled by a spray of tap water for about 2 min. of the 20-min. cycle; and (3) tests in which condensation of moisture on the specimens was induced at intervals by artificial means. The exposure tests were made by placing the panels on racks at 45° to the horizontal on a roof in Washington. All the tests were made on 6 × 3 in. steel panels, some plain and some galvanised, the paints being applied after various surface treatments. Various combinations of more than sixty priming paints and fifteen top-coat paints were tested. Chromate pigments, particularly zinc chromate, preponderated in the priming paints used. The main results of the investigation are: (1) Careful cleaning and preparation of the steel surface are considered more important than quality of the paint, but price should not be the deciding factor in selecting materials, because high initial cost may in the end be more economical. (2) The value of chemical methods of pretreating both plain and galvanised steel surfaces was demonstrated; phosphating processes are especially valuable. (3) On plain steel, whether phosphate-treated or not, primers pigmented with zinc chromate, alone or in combination with red oxide or aluminium, gave good results; the media used were chiefly of the alkyd or phenolic resin type. (4) On untreated hot-galvanised steel, primers pigmented with mixtures of zinc dust and zinc oxide such as those covered by the United States Federal Specification TT-P-641 alone gave good results. (5) The list of primers giving good results was greatly increased (numerically from 4 to 17) when the material was phosphated before painting. (6) Where the maximum protection is desired and decoration is of minor importance, aluminium, iron oxide, olive drab and black paints should prove most suitable as top coats because they have greatest resistance to the effects of sunlight and outdoor weathering; under conditions of high humidity, aluminium paints should be best for this purpose. (7) A system composed of phosphate-treated galvanised steel, zinc-chromate priming and aluminium top-coat paint combines to a high degree the essential requirements for withstanding corrosive atmospheres.

New Cycle Testing System Provides Means for Quickly Checking Paint Film Characteristics. K. N. Kathju. (Steel, 1944, vol. 115, Nov., pp.

100–101, 110). Descriptions are given of devices which were developed for the rapid testing of the hardness, adhesion and elasticity of paint films. The "Microknife" is an instrument for drawing a diamond point across a painted surface under controlled conditions; the load applied and the number of strokes are a measure of the scratch hardness. Adhesion is tested with the same instrument by making a series of parallel cuts through the paint, the distance between the scratches being successively decreased until the film between them is dislodged. The degree of adhesion is measured by the distance between the scratches when this occurs. A modified Erichsen machine is used to test the elongation. The machine is arranged to thrust a $\frac{3}{16}$-in. mandrel upwards against the underside of the painted test panel. Water is applied to the painted surface through a fine copper tube. The tube and the panel are connected to a galvanometer; as long as the paint stretches with the sheet metal there is no electrical contact, but when the film is ruptured the bimetal couple causes a deflection of the galvanometer needle, and the depth to which the metal has been deformed when the paint film breaks is the criterion of its elasticity.

PROPERTIES AND TESTS

(Continued from pp. 109 A–111 A)

Measurement of Dynamic Stress and Strain in Tensile Test Specimens. R. O. Fehr, E. R. Parker and D. J. DeMichael. (Journal of Applied Mechanics, 1944, vol. 11, June, pp. A-65–A-71). In the investigation described the tensile and yield strengths and the breakage energy of specimens of cold-rolled steel and duralumin were measured while the specimens were being broken by a force applied at a high rate of speed in a commercial high-velocity impact-testing machine. The dynamic tensile and yield strengths and the dynamic breakage energy were found to be higher than the static values up to the maximum impact velocities applied. The technique used is explained and some test results are presented and analysed.

The Acoustic Strain Gauge. R. S. Jerrett. (Journal of Scientific Instruments, 1945, vol. 22, Feb., pp. 29–34). An acoustic strain gauge and its application to the measurements of surface strains produced both by static and dynamic loading are described. The apparatus consists of a test gauge, a reference gauge and a control set. The note from a vibrating wire in the test gauge is matched against the note from a similar wire vibrating in the reference gauge. The method of measurement is very sensitive, and under normal conditions strains of the order of 1×10^{-4} in. can be recorded. The test gauge may be used in remote positions and controlled from a distance.

Rosette Strain Calculator. T. A. Hewson. (Tech. Engineering News, 1944, vol. 26, Oct., pp. 38–39, 54–56). Electrical strain gauges are now constructed in units holding three or four gauges set at different angles; these are known as "rosette" strain gauges. The computation of the principal stresses from the readings obtained is tedious work. In this paper a mechanical calculator for the rapid computation of these stresses is described.

Conditions of Fracture of Steel. J. H. Hollomon and C. Zener. (Transactions of the American Institute of Mining and Metallurgical Engineers, 1944, vol. 158, pp. 283–296). The effects of temperature and of strain rate on the behaviour of metals are discussed, and examples are given of the dependence of flow stress and stress at fracture on these two vari-

ables. Tests are also described in which the flow stress and fracture strength were obtained over wide ranges of temperature and strain rate for a pearlitic steel and a tempered martensitic steel at two stress levels. The results indicated that deformation had a negligible effect on the fracture strength of the martensitic steel. The anomalous effect of deformation on the fracture strength of a pearlitic structure is due to a reorientation of the carbide lamellæ. The phenomenon of fracture is readily understood in terms of the concept of a flow stress and a fracture stress introduced by Ludwik. In the yield-strength ranges of the steels examined, only the pearlitic steels were brittle at high strain rates and at low temperatures.

Torsion Testing Machines. (Engineer, 1945, vol. 179, Feb. 9, p. 120). Descriptions are given of two new designs of torsion-testing machines, one motor driven and the other hand driven.

Measurement of the Damping of Engineering Materials during Flexual Vibration at Elevated Temperatures. C. Schabtach and R. O. Fehr. (Journal of Applied Mechanics, 1944, vol. 11, June, pp. A-86–A-92). The method and equipment developed and used by the authors for measuring the damping of materials are described. A tuning-fork specimen is vibrated by jerking a spreader from the gap between the ends of the tines. The damping is expressed in terms of the logarithmic decrement of the decaying vibration, which is measured and recorded by means of a magnetic oscillograph, amplifiers and a resistance-type electric strain gauge cemented to the specimen. The steels and non-ferrous alloys tested differed widely in damping and in their variation of damping with temperature.

Methods for Computing Springback when Bending Stainless Steel. C. M. Brown, W. O. Binder and R. Franks. (Metal Progress, 1944, vol. 46, Nov., pp. 1077–1080). The mechanical properties of 18/8 and 17/7 stainless-steel sheet in the annealed state and after various degrees of reduction by cold-rolling are presented and discussed, and it is shown how to use the data to calculate the amount of "spring-back" which will occur on releasing the sheet after bending in a press.

Recrystallisation after Hot-Deformation with Special Reference to Deformations in the α-Iron Range. H. Kornfeld and G. Hartleif. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Mar.–Apr., pp. 221–226). An investigation is described in which specimens of Armco iron (measuring $10 \times 10 \times 55$ mm.) were deformed at different temperatures by a single blow from a falling weight having a sloping face, so that the specimen was subjected to different degrees of deformation along its length. The object was to determine the effect of the initial grain size and the temperature on the final grain size. The grain size after deformation at above the Ac_3 point was independent of the initial grain size, and was determined by the temperature and degree of deformation. The grain size after deformation in the α-phase at up to the critical forging temperature varied directly with the initial grain size, except after high degrees of deformation, which refined the grain considerably. The effect of increasing the time between two blows each producing 10% reductions at 800° C. on the grain size was also investigated. With an interval of 2 sec. the final grain size was the same as that after 20% reduction in a single blow. With intervals increasing up to 7 min. the final grain size increased rapidly; with longer intervals it increased quite gradually to a maximum size at 5 hr.

Deformation of Metals. H. W. Swift. (Manchester Metallurgical Society: Metallurgia, 1944, vol. 31, Dec., pp. 53–62). An engineer's assessment of the problems involved in plastic flow is made by defining the range within which the conditions of plastic flow operate and by

attempting to discover the laws of plastic flow within that range. There are three hypotheses for the inception of plastic flow which survive direct experimental test, namely, maximum shear stress, shear resilience and total resilience. There appears to be sufficient evidence for preferring shear resilience as the criterion for the inception of plastic strain in a statistically isotropic material. Some characteristics of plastic deformation are: (1) The strains in plastic flow are large compared with elastic strains; (2) plastic strains take time to develop; (3) the work absorbed in the development of plastic strain reappears largely in the form of sensible heat, but about 10% of it remains in some potential form in the material; (4) plastic strain may be regarded either as the rate of straining or creep in relation to time, or as the resultant amount of distortion necessary to restore equilibrium between the intrinsic resistance of the material and the applied stress; (5) the directions of plastic strain are related to the directions of the applied stress system; and (6) plastic strain is distortional involving a change in shape but not in volume.

The true stress-strain curve is discussed. Its slope does not represent the strain-hardening rate, and the area below it does not represent the specific strain-work, but if a curve is drawn in which the true stress is plotted against the logarithm of the initial area divided by the reduced area, the slope at every point is directly proportional to the rate of strain-hardening, and the area under the curve is a direct measure of the strain work. It is suggested that the rather elusive Bauschinger effect is fundamentally due to the non-uniform distribution of internal stress in an over-strained material and is not necessarily a function of strain-hardening at all.

The Wedge-Drawing Test. E. M. Loxley and H. W. Swift. (Institution of Automobile Engineers: Engineering, 1945, vol. 159, Jan. 12, pp. 38–40; Jan. 26, pp. 77–80; Feb. 16, pp. 136–138). The development of the wedge-drawing test for sheet metal is discussed, and a description is given of the shape of specimen and the design of the clamps which were adopted for an investigation of the drawing properties of a 0·07%-carbon steel strip 0·036 in. thick. Tests were made to compare the effects of lubrication with tallow, with Aquadag and of no lubrication. The breaking loads in the first two cases were very consistent at 1·52 tons. With no lubrication the consistency was less perfect, fracture occurring in the range of 1·50–1·58 tons. Critical and experimental examination of wedge-drawing as a basis for the commercial testing of deep-drawing properties points to the following conclusions: (1) In theory, wedge-drawing is a unidirectional simulation of cup-drawing in a simplified form which eliminates any effects of die or punch profiles; (2) for determining the limiting drawing ratio, it is necessary to carry out a series of convergent bracketing tests, but for proof testing or acceptance purposes, one test at a single specified drawing ratio would suffice; (3) because of the shape of the wedge specimen, it would be difficult to carry out tests on transverse specimens taken from narrow strip; (4) a wedge-drawing test could not be carried out in a machine of normal design, but would demand one with a specially long straining range; (5) the wedge-drawing test is, on the whole, more readily adaptable than the cup-drawing test for specimens of different thickness; (6) wedge specimens are more expensive to prepare than circular blanks; (7) the drawing ratio obtained in practice from the wedge-drawing test is substantially lower than that obtained in the cup-drawing test; and (8) the marginal range of uncertainty between success and failure is considerably greater in the wedge test than in cup-drawing.

Shrink-Fit Stresses and Deformations. A. W. Rankine. (Journal of Applied Mechanics, 1944, vol. 11, June, pp. A-77–A-85). Equations are

derived for the stresses in a solid cylinder of infinite length caused by the action of an applied radial stress distributed uniformly over a single circumferential ring of finite length on the surface of the cylinder. The practical value of the results presented lies in their approximate applications to shrink fits. Test measurements of shaft deformations due to shrink pressures are included to illustrate the application of the theoretical results.

Magnetic Crack-Detection Equipment. (Engineering, 1945, vol. 159, Feb. 2, pp. 86–87). Two forms of apparatus for the magnetic-powder testing of steel bars are described; one is for detecting longitudinal cracks and the other for transverse cracks.

Magnetic Crack Detection and Its Application in Rod-Drawing Shops. A. Weisselberg. (Stahl und Eisen, 1944, vol. 64, Dec. 21, pp. 791–797). The latest types of apparatus for the magnetic powder testing of steel bars in general, and of aero-engine parts in particular, are described and illustrated with notes on introducing this method of testing in the rolling mill.

The Effect of the Steel on the Life of Gudgeon Pins. E. Mickel and P. Sommer. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Mar.–Apr., pp. 227–234). A statistical analysis of the causes of the failure of gudgeon pins is presented and their relative importance is discussed. An examination of over 100 failures and the results of fatigue tests in compression indicated that the life of gudgeon pins made of many different alloy steels is approximately the same. The two main causes of failure are slag inclusions and faulty heat-treatment. Sand-blasting or grinding the inner surface of a pin increases the fatigue strength. Whilst the magnetic powder method is an excellent way of examining the outer surface, there is no really suitable method of testing the inner surface.

Fatigue-Testing Methods and Equipment. H. W. Foster and V. Seliger. (Mechanical Engineering, 1944, vol. 66, Nov., pp. 719–725). The more important considerations in the general design of fatigue-testing apparatus, and the methods and equipment developed and now in use at Lockheed Aircraft Corporation are discussed in detail.

Temper Brittleness and Heat Embrittlement of Alloy Steels. P. B. Michailow-Michejew. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Jan.–Feb., pp. 177–184). An investigation is reported on the influence of the hardening and tempering temperatures, the tempering time, the cooling rate after tempering, a prolonged second tempering, cold deformation and the chemical composition on the temper brittleness and heat embrittlement of low-alloy steels. The brittleness was determined by notched-bar impact tests at room temperature as well as at between $-200°$ and $+500°$ C. The specimens were cut from chromium-nickel and chromium-nickel-molybdenum steels. Temper brittleness and heat embrittlement are phenomena of similar nature and can be regarded as related to the embrittlement which occurs at low temperatures. This brittleness results from the steel remaining for a considerable time in a critical temperature range the extent of which depends on the tempering time and the chemical composition; the phenomenon often occurs at between 400° and 550° C. It can be prevented by alloying the steel with elements which impede the shifting of the brittle range to higher temperatures. An addition of molybdenum to chromium-nickel steels acts in this way. The susceptibility of chromium-nickel steels to heat embrittlement can be reduced by holding them at a high tempering temperature (up to 650° C.) for at least 24 hr., provided that this treatment does not raise the cold-embrittlement temperature range to room temperature.

The Creep of Heat-Resisting Steels at Temperatures of 800° to 1200° C. E. Siebel and G. Hahn. (Archiv für das Eisenhüttenwesen, 1944, vol. 17,

Mar.–Apr., pp. 211–220). An investigation of the creep strength of heat-resisting steels is described. The steels tested included three chromium steels (with 6%, 16% and 23% of chromium respectively), a 23/18 chromium-nickel steel, a steel containing 5% of aluminium, 9% of chromium and 1% of titanium, and one containing 18% of manganese, 9% of chromium and 1% of nickel. A special test apparatus was constructed in which hollow specimens with a gauge length of 80 mm., an outside dia. of 8·7 mm. and an inside dia. of 3·5 mm. were heated by their own resistance to a heavy current to temperatures up to 1200° C. for periods up to 1650 hr. A comparison of the loads required to cause a total creep of 1% at 1000° C. in the ferritic steels after 1000 hr. showed the great superiority of the chromium-aluminium steel; at 900° C the manganese-chromium steel gave the highest value. Previous treatment of the 16%-chromium steel at 1000° C. for 500 hr. increased the creep strength at 900° C. by 250%. The fractures in the austenitic steels were intercrystalline, whatever the temperature and load, whilst those of the ferritic steels were mostly transcrystalline. There was much more deformation before fracture in the case of the ferritic steels. The direct resistance heating of the specimens did not appear to affect either the deformation resistance or the structure of the specimens.

Some Thoughts on Medium and High Tensile Wrought Alloy Steels, Present and Future. J. H. G. Monypenny. (Metallurgia, 1944, vol. 31, Dec., pp. 79–82). The need of economy of certain alloys used in low alloy steels resulted in narrow limits of standardisation during war-time, although scrap-recovery somewhat eased the position. It is suggested that these restrictions should be eliminated in the post-war years.

Vanadium in Gray Iron Castings. R. G. McElwee and T. E. Barlow. (Foundry, 1944, vol. 72, Dec., pp. 88–89, 246–252). Recent developments in the use of grey cast-iron alloyed with vanadium are reviewed. Vanadium improves the properties of the iron in the heat-treated state, especially the resistance to wear, deformation, hot-cracking and distortion at high temperatures; it also increases the density.

This Heat-Treatment Business. A. J. K. Honeyman. (Metal Treatment, 1944, vol. 11, Winter Issue, pp. 207–212). The effects of adding boron to steel and the merits of some different heat-treating processes are discussed. Boron increases the hardenability of steel provided it is added to a previously fully-killed fine-grain steel. An improved yield ratio and impact strength are obtained provided full advantage is taken of the increased hardenability and the steel is fully hardened before tempering; with partial hardening boron is beneficial provided no transformation takes place in the pearlitic range. Some results obtained by austempering carbon steels and by the treatment of tool steels at sub-zero temperatures before tempering are quoted from the literature.

The Influence of Nitrogen on the Properties of Rustless Steels. A. Samarin, A. Yaskevich and I. Paisov. (Iron and Steel Institute, 1945, Translation Series, No. 215). An English translation is presented of a paper which appeared in the Bulletin de l'Académie des Sciences de l'U.R.S.S., 1943, No. 5–6, pp. 71–77. It contains an account of investigations of methods of alloying steel with nitrogen, and of the effect of nitrogen on the properties of stainless steels. Heat-treated cold-rolled steel containing 17–19% of chromium, 5% of nickel and 0·15–0·20% of nitrogen is equivalent to 18/8 stainless steel in its mechanical properties and corrosion resistance. The porosity of ingots increases with increase in the nitrogen/chromium ratio in the steel; if this ratio is greater than 0·01 there will be blowholes in the steel.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 111 A–112 A)

A Vacuum Microbalance for the Study of Chemical Reactions on Metals. E. A. Gulbransen. (Review of Scientific Instruments, 1944, vol. 15, Aug., pp. 201–204). The construction of a vacuum microbalance having a sensitivity of $0 \cdot 3 \times 10^{-6}$ g. is described. The technique of measurement is briefly discussed, and the results of some experiments in which it was used to study the formation of oxide films on iron at 25° and 300° C. are presented.

The Electron Microscope in Metallurgical Research. C. S. Barrett. (American Physical Society: Journal of Applied Optics, 1944, vol. 15, Oct., pp. 691–696). Some examples of electron micrographs obtained by the polystyrene-silica surface replica technique are reproduced and discussed. Experience suggests that new metallographic polishing and etching techniques are needed. Standard methods of etching often produce pitting which is too coarse; the newer electrolytic polishing sometimes gives better results than mechanical polishing. It is best to take electron micrographs at a low magnification and enlarge them by printing, and a final magnification of 2500, 5000 or 10,000 is generally preferable to higher magnifications that give very small fields of view.

The Microscopical Examination of Samples of Lead-Bearing and Lead-Free Steels and Ingot Irons. T. H. Schofield. (Iron and Steel Institute, 1945, this Journal, Section I). Microscopical examination of samples of lead-bearing steels and ingot irons has revealed small characteristic inclusions which are not present in similar lead-free materials. Using a modified polishing technique, these inclusions appear white, and an etching test indicates that they consist of or contain lead. The inclusions are removed by mercury at 100° C.

Mode of Occurrence of Lead in Lead-Bearing Steels and the Mechanism of the Exudation Test. W. E. Bardgett and R. E. Lismer. (Iron and Steel Institute, 1945, this Journal, Section I). As relatively little is known about the mode of occurrence and distribution of lead in lead-bearing steel, an extensive examination has been carried out on high-sulphur and manganese-molybdenum wrought steels and on 0·25% carbon, 1% manganese steel ingots. Normal methods of microscopic examination failed to reveal the presence of lead as discrete particles, but an electrolytic method revealed particles believed to be lead, not visible in unetched specimens. A new electrographic method is described which produces a clear pattern of the lead or lead-bearing particles in the ingot section. Interesting features were brought to light by microscopical examination of samples during heating.

Examination of Two Ingots of Free-Cutting Steel, One Containing Lead and the Other Lead-Free. C. S. Graham. (Iron and Steel Institute, 1945, this Journal, Section I). The segregation in an ingot of leaded steel has been compared with that in a non-leaded ingot from the same cast. Little difference in the chemical compositions at the Heterogeneity of Steel Ingots Committee's standard positions was found, with the exception of a small, but possibly significant, reduction in the oxygen content in the lead-bearing ingot. The lead itself was evenly distributed, except at the extreme bottom of the ingot. On the examination of an axial section, the leaded ingot showed less segregation of sulphur and a thinner columnar zone.

Bibliography of Stress Analysis by Means of X-Ray Back-Reflection Method. H. R. Isenburger. (Welding Journal, 1944, vol. 23, Nov.,

pp. 571-S-572-S). A bibliography of 97 references to the literature on stress analysis by the X-ray back-reflection method is presented.

The Systematics of the Binary Systems of Iron. H. Prediger. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Mar.–Apr., pp. 235–236). A complete explanation of the effect of alloying elements in steel on the limits of the phases has not yet been found. A theory is put forward which accounts for the phenomena observed in binary systems of iron. This is based on the formation of two types of solid solution, namely, substitution solid solutions, in which the atoms of the alloying element take the place of the iron atoms in the iron lattice, and interstitial solid solutions, in which the atoms of the alloying element occupy the spaces between the iron atoms. When a melt begins to crystallise the size of the lattice of the first crystals of solid solution to form is large enough, because of the high temperature of the γ-phase and the mobility of the atoms, for atoms of elements the atoms of which are of small volume to fill up the free space of the γ-cube and this will presumably expand the γ-lattice. As there is only space for one atom in the γ-cube there will be an average of four iron atoms to one dissolved atom in the case of a lattice in which all the γ-cube spaces are taken up. Any additional dissolved atoms are therefore compelled to take the place of iron atoms in the γ-lattice; in other words the formation of substitution solid solution begins. With further cooling and complete solidification the stresses set up by the cooling of the expanded γ-lattice forces more and more of the interstitial atoms to replace iron atoms so that finally a pure substitution solid solution, as observed at room temperature, is formed. This theory of the process of crystallisation accounts for the fact that when the atoms of an alloying element have a volume below a certain limit, that element will extend the γ-range, but elements having larger atoms will either narrow the γ-range or be insoluble. The anomalous behaviour of boron is explained by the fact that this unites with iron to form the compound Fe_2B. For beryllium a larger atomic volume than the accepted one must be assumed.

Transformation in Ferrous Alloys. C. J. Osborn. (Australian Institute of Metals: Australasian Engineer, 1944, vol. 44, Dec. 7, pp. 26–29). Transformations in ferrous alloys and the effects of different alloying elements are explained by reference to equilibrium diagrams and transformation-temperature/time curves. Additions of all alloying elements, except cobalt, retard the decomposition of austenite, and this is due partly to the necessity for diffusion of the slower-diffusing alloying element. The only other major effect is that of the carbide-forming elements in increasing the reaction time in the range of about 500° to 600° C. The explanation of this phenomenon is not known, but it is probably related to the stability of the carbides of the alloying elements.

The Iron/Iron-Sulphide/Iron-Boride System. R. Vogel and T. Heumann. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, May–June, pp. 271–274). The Fe–FeS–FeB system was studied by metallographic and X-ray methods. Between FeB and Fe_2B on the one hand and FeS on the other there is complete insolubility in the liquid state. The ternary miscibility gap extending from this region covers the whole of the ternary system right up to a small region of solubility in the iron corner of the equilibrium diagram. Two sections of the diagram are reproduced, one at 12% of sulphur and the other at an iron/boron ratio of 97·5/2·5. It is evident that boron is one of the elements very small quantities of which cause iron/iron-sulphide melts to separate into two components.

CORROSION OF IRON AND STEEL

(Continued from pp. 112 A–116 A)

On the Kinetics of the Formation of Oxide Films on the Surface of Metals. J. Frenkel. (Academy of Sciences of the U.S.S.R., Journal of Physics, 1944, vol. 8, No. 4, pp. 225–229). A mathematical study of the dissociation of adsorbed oxygen molecules into atoms and the rate of penetration of the atoms into the surface of metals is presented.

Stress-Corrosion Tests of Bridge-Cable Wire. R. E. Pollard. (Journal of Research of the National Bureau of Standards, 1944, vol. 33, Sept., pp. 201–211). Stress-corrosion cracks were produced in several statically stressed specimens of cold-drawn high-carbon steel wire from the Portsmouth Bridge (U.S.A.) and in one specimen from the Mount Hope Bridge by immersion in dilute nitrate solutions. No cracks were produced in the cold-drawn Mount Hope replacement wire after long exposure. No cracks were produced by immersing similar specimens in distilled water, dilute ammonium sulphate or in inhibitive solutions such as dilute ammonium nitrate or dilute sodium hydroxide. In this respect the results conform to the theory of selective corrosion by intercrystalline attack.

Soil-Corrosion Studies, 1941: Ferrous and Non-Ferrous Corrosion-Resistant Materials and Non-Bituminous Coatings. K. H. Logan and M. Romanoff. (Journal of Research of the National Bureau of Standards, 1944, vol. 33, Sept., pp. 145–198). A report is presented on the resistance of a variety of ferrous and non-ferrous materials with metallic and organic coatings to corrosion by fourteen different soils in periods of two to nine years. The addition of small percentages of certain alloying elements to some ferrous alloys improved their corrosion resistance by an insignificant amount. Ferrous materials must contain considerable quantities of nickel and/or chromium if they are to resist the attack of very corrosive soils. A coating of 3 oz. of zinc per sq. ft. added about three years to the life of steel exposed to some of the most corrosive soils. Lead coatings appeared to be inadequate for severe conditions. Air-dried Bakelite coatings blistered within four years, but a baked-on Bakelite coating had only a few blisters after four years; pitting occurred under some of the blisters.

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REFRACTORY MATERIALS

(Continued from pp. 118 A–119 A)

Investigation of Refractory Material for Teeming Equipment. L. O. Uhrus and K. Öberg. (Jernkontorets Annaler, 1944, vol. 128, No. 12, pp. 597–609). (In Swedish). An investigation of the erosion of different ladle nozzle materials by different steels is described. Nozzles 33 mm. in dia. were used, some of hand-pressed and others of machine-pressed aggregates, and the effect of de-airing them was tested. The nozzle wear was measured and given as the increase in diameter in millimetres per 1000 kg. of steel teemed. The nozzle of hand-pressed de-aired material showed the least wear with all three qualities of steel teemed. The steel containing carbon 0·15% and manganese 1·20% caused much more wear than steels containing carbon 1% and either 1·10% or 0·35% of manganese. Some of the steel was teemed into water to form granules, and these were sectioned and examined for inclusions; the steel with 1% of carbon and 0·35% manganese had the least inclusions. The nature of the inclusions was studied in the light of the SiO_2–Al_2O_3–MnO equilibrium diagram, and the distribution of inclusions in ingots was also examined.

Converter Refractories. W. C. Cress. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 854–869). The properties required for refractory lining materials for Bessemer converters are discussed and the preparation of a mix that will have maximum bond and contain a minimum of alumina is described. The results of tests on many mixtures are given. The longest life of monolithic linings in 3-ton converters and in ladles was obtained using a mixture of ⅜-in. ganister 50%, ⅛-in. ganister 30%, 140-mesh silica flour 14% and bentonite 6%.

Insulation of Open-Hearth Furnaces and Blast Furnaces. J. M. Ferguson. (Institute of Fuel: Iron and Coal Trades Review, 1945, vol.150, Feb. 16, pp. 245–247; Feb. 23, pp. 280–282, 289). Data on the heat losses and wear of the refractory walls and surfaces of an old and a modern 90-ton basic open-hearth furnace are presented and discussed. The improvement in the newer furnace is due to better placing of the thermal resistances at the points of maximum wear. Considerable advantages accrue from the judicious use of water cooling at vital points but compromise is necessary. Insulation on the old furnace to the same extent as on the new one would have saved 0·56 cwt. of producer fuel per ton of steel. Special insulation should not be applied to open-hearth roofs, but appreciable thermal saving can be achieved by increasing the thickness of the roof. Where dolomite bricks are used in the hearth, insulation should not be applied and the maximum thickness of the firebrick shaping courses should be carefully calculated. Hearth insulation imposes increased thermal pressure upon the refractories, and the risks involved make it inadvisable to insulate the hearths of either of the two furnaces. Insulating the gas main of the newer furnace reduced the fuel consumed to make good the heat lost from 6¼% to 2% of the fuel charged into the producer. Insulation of blast-furnace stacks is not recommended but important economies have been obtained by insulating hot-blast mains and bustle pipes.

High Temperature Heat Insulation. G. W. Paterson. (Canadian Ceramic Society: Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 927–944). *See* Journ. I. and S.I., 1943, No. II., p. 2A.

FUEL

(Continued from pp. 120 A–121 A)

The Theoretical Quantities of Air and Products of Combustion When Using Gaseous Fuels. H. A. Lundberg. (Jernkontorets Annaler, 1944, vol. 128, No. 12, pp. 617–620). (In Swedish). Equations are developed for calculating the quantities of air and the products of combustion when burning gaseous fuels. They are based on data obtained when using producer gas, blast-furnace gas and gas from electric smelting furnaces.

Why Destroy Heat by Excess Air, Why not Conserve Heat by Regenerative Heating ? G. Neumann. (Stahl und Eisen, 1945, vol. 65, Jan. 18, pp. 43–45). Data are presented and applied to calculate the heat losses and the gas consumption of furnaces with and without regenerators. The increase in gas consumption in the latter case may amount to as much as 140% with core-drying stoves and up to 35% with sand-drying stoves.

The Flow of Coal-Ash Slag on Furnace Walls. P. Cohen and W. T. Reid. (United States Bureau of Mines, 1944, Technical Paper No. 663). The development of pulverised-coal furnaces with provision for tapping off the slag from the ash has led to studies of the flow properties of the slags and their effect on the rate of heat transfer. An equation is derived for computing the thickness of slag deposits under various furnace conditions. For fixed furnace conditions, the viscosity of the slag and the temperature of critical viscosity (the temperature at which the cooling slag passes from the viscous to the plastic state) are equally important in controlling the thickness of the deposited layer of slag ; this thickness increases with an increase in either of the above factors.

Carbonizing Properties of Western Region Interior Province Coals and Certain Blends of These Coals. J. D. Davis, D. A. Reynolds, J. L. Elder, W. H. Ode, C. R. Holmes and J. T. McCartney. (United States Bureau of Mines, 1944, Technical Paper No. 667).

Carbonizing Properties of Pocahontas No. 3-Bed Coal from Kimball, McDowell County, W.Va., and the Effect of Blending this Coal with Pittsburgh-Bed Coal. J. D. Davis, D. A. Reynolds, W. H. Ode and C. R. Holmes. (United States Bureau of Mines, 1944, Technical Paper No. 670).

Plastic Characteristics of Coal. R. E. Brewer. (Industrial and Engineering Chemistry, Industrial Edition, 1944, vol. 36, Dec., pp. 1165–1168). A study is made of the chemical and petrographic compositions and the physical properties of nineteen American low-, medium- and high-volatile bituminous coals and of the relationship between these properties and the maximum fluidity as measured by the Gieseler test.

Production of Sulphate by the Crystallization Method. J. B. Hazel. (Eastern States Blast Furnace and Coke Oven Association and the Blast Furnace and Coke Association of the Chicago District : Blast Furnace and Steel Plant, 1944, vol. 32, Dec., pp. 1445–1450). Experience gained in the operation of the low-differential large-crystal ammonium sulphate process described by W. Tiddy is discussed (*see* Journ. I. and S.I., 1942, No. I., p. 215A).

Utilising Coal to Best Advantage. W. H. Brook. (B. H. P. Review, 1944, vol. 22, Dec., pp. 1–4). An account is given of the improvements in the methods of utilising coke-oven gas and blast-furnace gas which have been developed since 1920 at the Newcastle Steelworks of the Broken Hill Proprietary Co., Ltd.

Bureau of Mines Research on the Hydrogenation and Liquefaction of Coal and Lignite. A. C. Fieldner, H. H. Storch and L. L. Hirst. (United States Bureau of Mines, 1944, Technical Paper No. 666). A brief review of

the research work done by the Bureau of Mines on the direct hydrogenation and liquefaction of American coals is presented.

FOUNDRY PRACTICE

(Continued from pp. 122 A–124 A)

What the Engineer Expects from the Foundry. W. P. Eastwood. (Institute of British Foundrymen : Foundry Trade Journal, 1945, vol. 75, Mar. 1, pp. 167–170). The requirements of castings for large and small machine tools are discussed, and some defects in the organisation of jobbing foundries are pointed out.

Alterations in Cast Iron Properties Accompanying Use of a Strong Inoculant of the Silicon-Manganese-Zirconium Type. C. O. Burgess and R. W. Bishop. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 671–705). Tests are described on the effect of ladle inoculation on a series of irons covering a wide range of composition using a silicon-manganese-zirconium alloy as inoculant. The total carbon and the silicon contents of the irons used covered the 2·75–3·40% and the 1·0–2·5% ranges respectively. The normal chilling tendencies of cast irons can be materially reduced by ladle additions of a strong inoculant. The best properties were obtained by inoculation with the above alloy when the total carbon content of the iron was as low as possible, *i.e.*, below 3·05%. Variation in the silicon content between 1% and 2% had practically no effect on the tensile strength, but keeping it high limited the chill depth, increased the fluidity and improved the general founding characteristics.

Problems Encountered in Making and Heat-Treating Castings of an Air-Hardening Steel. G. W. Johnson. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 814–817). For castings which must be hardened throughout with a minimum of distortion after machining, the author recommends an air-hardening steel to the following analysis : Carbon 0·55%, manganese 0·70%, chromium 0·75%, nickel 1·50%, molybdenum 0·50% and silicon about 0·40%. The heat-treatment of castings of this steel is described in detail.

Brazil Produces Steel Castings to Replace Forgings and Fabricated Parts. H. Zimnawoda. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 871–882). War-time conditions in Brazil have enforced the development of steel castings to replace forged and rolled products. The moulding, pouring and heat-treatment techniques for making cast steel furnace roof rings 9 ft. in dia. and large connecting rods, crankshafts and columns at a foundry in San Paulo are described. The steel is melted in a 3-ton basic-lined Héroult furnace.

Reproducibility of Elevated Temperature Sand Test Results. J. A. Rassenfoss. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 711–732). An account is given of a study of the accuracy with which the results of elevated temperature compression strength tests can be reproduced at a single laboratory and between laboratories at different plants.

Foundation of Sand Rammer. (American Foundryman, 1944, vol. 6, Dec., pp. 11–16). An account is given of the work done by a sub-committee of the American Foundrymen's Association which was appointed to " investigate and standardise the method of ramming test specimens and of checking and standardising tube and stripping post equipment." The values obtained for the density, permeability and strength of rammed specimens of foundry sand held by rigid and non-rigid supports differ even when all the other conditions are kept the same. Non-rigid supports, such

as wooden tables and tables on wooden floors, should not be used. Steel, concrete or wooden columns securely anchored, with the rammer fastened tightly to the top, are all satisfactory.

Mold Surface Gas Pressure. H. W. Dietert, R. L. Doelman and R. W. Bennett. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 733–749). An apparatus is described with which the permeability to gas flow of small rammed samples of foundry sand at high temperatures was measured. Four mixtures of an Ottawa sand with bentonite containing different amounts of cereal, silica flour and moisture were tested, and determinations of the pressure of the gases generated by the hot sand were made.

Gating and Feeding Malleable Pipe Fittings. M. L. Hawkins. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 593–603). Detailed descriptions are given of the moulding practice at an American foundry for the production of sand-cast malleable iron pipe fittings.

Heading and Gating of Malleable Iron Castings. A. J. Klimek. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 604–609). Particulars are given of methods of controlling the flow of metal in the production of malleable iron castings ranging from 0·1 to 300 lb. The positions of the runners and risers are shown in several illustrations.

Gating and Heading. C. C. Lawson. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 612–622). The positioning of feeder heads and the use of "skim cores," combs and other devices in gates and runners to eliminate slag and dirt in small iron castings are explained.

Gating and Feeding of Malleable Iron Castings. A. T. Jeffery. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 623–628). Some examples are given of the positioning of risers for medium-weight malleable iron castings which were difficult to feed.

Gating Malleable Castings. D. I. Dobson. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 632–640). Examples are given of the gating and risering of heavy automobile malleable iron castings.

Heading and Gating Malleable Castings. E. F. Waterbor. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 643–646). The moulding practice for casting automobile differential cases in malleable iron is described.

Industrial Heating. Advantages of Town Gas for Mould Drying, Smith's Work and Pipe Bending. (North British Association of Gas Managers: Iron and Steel, 1945, vol. 18, Feb., pp. 57–58). Some experiments carried out by the Aberdeen City Gas Department on the use of gas in foundries and the blacksmith's shop are described. In a test on drying 7½ tons of wet sand in a stove 10 × 10 × 9 ft. at 400° F. for 16 hr. with 13 changes of air per hr. the surface of the mould was well dried, but the centre was too damp for a cast to be made. A second test was made at 360° F. with seven changes of air per hr., thus keeping the atmosphere more humid. In this case the drying was highly satisfactory and the gas consumption was reduced from 7·7 to 4·25 therms per 1000 lb. of sand.

Work on the Spindle. J. H. List. (Iron and Steel, 1945, vol. 18, Feb., pp. 51–52). A detailed description is given of the preparation of a mould for a balanced flywheel 2 ft. 7½ in. in dia. in green sand using a spindle and strickle boards.

Improvements in Pressure Ferrous Castings Influencing their Future Use. E. C. Jeter. (Society of Automotive Engineers: American Foundryman, 1944, vol. 6, Dec., pp. 5–9). Examples of steel parts for automobiles and aeroplanes made by centrifugal casting are described and illustrated.

Mention is made of casting to very close dimensions by applying air under pressure through the top of a riser while solidification takes place.

Maintenance of Equipment. F. J. Dost. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 784–788). The organisation of maintenance work in foundries is discussed.

The Elimination of Micro-Shrinkage by Accurate Thermal Dissipation. H. F. Hagemeyer. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 758–766). The thermodynamic considerations involved in the pouring of metal into moulds and in the subsequent solidification and cooling are discussed in detail and the properties of an ideal mould material are pointed out. The ideal pouring condition would be one in which the liquid metal enters the cavity at the bottom without turbulence in such a manner that the oxide skin is not ruptured until the entire mould cavity is filled.

Heat-Flow Problems in Foundry Work. V. Paschkis. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 649–670). The theory of the heat-flow analyser developed at Columbia University has been explained by V. Paschkis and H. D. Baker (*see* Journ. I. and S. I., 1942, No. I., p. 133 A). This electrical apparatus consists of a large number of resistors and condensers by means of which a great variety of circuits can be produced for the study of heat-flow problems by analogy. In the present paper the application of this apparatus to study the freezing of iron in sand moulds is described. The simple case selected was that of a cylinder 10 in. in dia. × 15 in. high with a feederhead above it 8 in. in dia. × 6 in. high; between the two there was a core or gate with an opening 2 in. in dia. in one test and 3 in. in dia. in a second test. Cooling curves for various points in the casting and in the mould are presented. The size of the hole in the core had very little effect on the cooling rate of the casting. Those tests were only intended to indicate the possibilities of the heat-flow analyser.

Special Low Carbon Steel for Castings. C. G. Lutts and J. P. Hickey. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 904–911). A description is given of a defect which frequently occurred in the bore of heavy cast-steel pipe-fittings. This took the form of cracks in the plane of the flange; they were detected by the magnetic powder process. The fittings were normally made of 0·20–0·30%-carbon steel, but as this steel was known to be less ductile at temperatures a little below the solidification point than steel with 0·12% max. of carbon, the American Navy Yard concerned developed a special steel with about 0·10% of carbon, 1·10% of manganese and 0·50% of molybdenum. The occurrence of cracks in flanges of this steel was very low, but the cost of making the steel was considerably higher.

Some Aspects of Casting Inspection as Seen by a Purchasing Agent. R. V. Elms. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 750–757). Specifications for castings for a company making a variety of instruments for the United States Government and the difficulty in obtaining satisfactory agreement between the purchaser and the foundry are discussed.

PRODUCTION OF STEEL

(Continued from pp. 124 A–125 A)

Presidential Address. O. J. Thomas. (Swansea and District Metallurgical Society, Oct., 1944). The effect of war-time conditions on the steel industry of South Wales is reviewed and some suggestions for the

improvement of older plants are discussed. Several advantages resulting from the use of clean gas instead of solid fuel for furnaces in the tinplate industry are pointed out, with particular reference to gas made from Welsh anthracite and to the Lee Wilson annealing furnace. To obtain maximum ductility in sheets it is essential to control the rate of cooling to produce the desired grain size with a good distribution of pearlite and to cool as slowly as possible from 550° C. to room temperature.

The Metallurgy of Side-Blown Converter Practice. H. Wentrup and O. Reif. (Iron and Steel Institute, 1945, Translation Series, No. 212). An English translation is presented of a paper which appeared in Stahl und Eisen, 1944, vol. 64, June 1, pp. 349–358 (*see* p. 9 A).

Physical Control Methods in the Steel Industry. W. Barr and T. F. Pearson. (Journal of Scientific Instruments, 1945, vol. 22, Jan., pp. 1–5; Engineering, 1945, vol. 159, Feb. 9, pp. 104–105; Mar. 2, pp. 165–166). Some methods of control (involving physical principles) which are used in large steelworks are reviewed; their significance and some of the difficulties still outstanding are considered. Methods of determining the temperature of liquid steel and of furnace flames, furnace pressure regulators, fuel-gas flow meters, controllers for fuel-gas/air ratio control and rapid methods of determining the carbon in steel are described.

Working a Heat of Acid Electric Steel (Notes for Operators). C. C. Wissmann. (Metal Progress, 1944, vol. 46, Dec., pp. 1277–1284). Recommendations are made on the melting practice for acid-lined electric furnaces. The ideal condition is for the boil to have boiled itself out when the heat is hot enough to tap. The steel should be boiled for at least 10 min., and the slag, when the steel is tapped, should be getting lumpy and be a pea-green colour.

The Constitution of Basic Steel Furnace Slags. J. R. Rait and H. J. Goldschmidt. (Iron and Steel Institute, 1945, this Journal, Section I). Systems of phase assemblages were deduced from the available phase-diagram data for basic electric reducing and oxidising slags and basic open-hearth slags. Accordingly, the constitutions of these slags were calculated and compared with the constitutions determined by the X-ray powder method. The agreement between the theoretical and observed results indicates that the systems of phase assemblages are essentially correct. Further confirmation was obtained by correlating the constitutions of a number of basic open-hearth slags as observed by Mason (Journ. I. and S. I., 1944, No. II., p. 69 P) with the theoretical results deduced from his analysis. The slags are classified according to their constitutions, the chief characteristics discussed and the various phases described.

The occurrence of merwinite as the main phase in certain basic electric reducing slags and the mode of occurrence of fluorine and TiO_2 was also established. An X-ray method for the quantitative estimation of the amounts of phases was developed for reducing slags.

In electric oxidising slags the (Fe,Mn,Mg)O solid solution, the composition and amount of which vary from slag to slag, is a major phase. The lattice parameters of the RO phases were measured and correlated with their compositions, and the change of spacing during the oxidising stage was observed. An interesting feature is the appearance of a high-temperature form of $2CaO.SiO_2$, as distinct from the β form (denoted α'-$2CaO.SiO_2$), and of merwinite liable to contain manganese-merwinite in solution.

Basic open-hearth finishing slags from a number of steelworks were examined and their constitutions established. The chief phosphate phase is nagelschmidtite or fluorapatite, depending on whether fluorspar had been added or not. The solid solution (Fe,Mn,Mg)O occurs in all the slags and, in the more basic slags, co-exists with free (Ca,Mn)O. The co-existence of these two RO phases established by the X-ray method is believed to be a

new discovery of considerable practical importance. The amount of the (Ca,Mn)O phase decreases with decreasing basicity until a limiting value is reached at which all the lime is in combination and only one RO phase (Fe,Mn,Mg)O remains. A practical X-ray method of slag control based upon the estimation of these two RO phases is proposed.

A Study of the Basic Open-Hearth Process, with Particular Reference to Slag Constitution. A. H. Jay. (Iron and Steel Institute, 1945, this Journal, Section I). An account is given of experiments carried out to investigate the constitution of basic open-hearth slags in relation to the mechanism of dephosphorisation in the furnace and rephosphorisation in the casting pit.

The work described involved the X-ray examination and chemical analysis of slag samples taken during the refining period and the casting of the ingots. Differences in constitution were found, depending on the rate of cooling of the slag samples. Changes in bath composition have been studied by means of metal samples taken during the same period. The X-ray work has been supplemented by the examination of oxide mixtures prepared in the laboratory, the results of which have greatly contributed to the interpretation of the slag constitution. A hypothesis has been evolved to relate rephosphorisation in the pit to the composition of the tapping slag. A suggestion is made to explain the mechanism of dephosphorisation in the furnace.

The Phosphorus Reaction in Basic Open-Hearth Practice. Y. K. Zea. (Iron and Steel Institute, 1945, this Journal, Section I). The practical applications of the equations for calculating the phosphorus content of the metal on the basis of the slag composition and the bath temperature, established by four investigators, are examined in relation to new temperature data and slag and metal analyses obtained from fifteen casts of basic open-hearth steel. It is found that Schenck and Riess' method gives the best results. The reversion of phosphorus from slag to metal during casting was studied on twenty-five casts of basic open-hearth steel, first by graphical examination of the slag and metal analyses together with the bath temperature, and secondly by the application of Schenck's equilibrium diagrams. It is found that for a given set of working conditions, the rephosphorisation likely to occur in the ladle may be estimated quantitatively by the use of Schenck's equilibrium diagrams to establish a ratio of free lime to free iron oxide concentration in the final furnace slag from its composition and the bath temperature. The results of this investigation show that rephosphorisation during casting is due to a change in the composition of the slag by enrichment of the silica content resulting from its reaction with the fireclay lining of the ladle. It is found that rephosphorisation does not occur when the ladle is lined with basic material; and that the degree of rephosphorisation, when a fireclay-brick-lined ladle is used, may be controlled to a large extent by keeping the temperature of the slag and metal in the ladle as low as possible, and by adjusting the slag composition so that, in spite of the reaction with the ladle lining, the basicity of the slag will not fall below a certain minimum necessary for holding phosphorus. The latter aim can be achieved, when circumstances permit, by increasing the lime content of the slag.

The Movement of Slag Particles in Liquid Steel. F. Hartmann. (Stahl und Eisen, 1945, vol. 65, Jan. 18, pp. 29–36). The movement of solid and liquid inclusions in steel was studied and subjected to a mathematical analysis. The different types of movement were defined and the rates at which the particles rise in liquid steel were calculated using Stoke's law. This rate depends on their size and specific gravity and on the viscosity of the steel, which in turn depends on its temperature and composition. An example is given of the application of such data to the determination of the

size of particle which would have time to rise to the surface of steel in a ladle filled to a given height in a given time; the effect of cooling, which increases the viscosity of the steel, was also calculated. The movement of slag particles in steel in horizontal and in vertical turbulent motion is discussed, and it is shown that it is much more difficult to remove solid particles from steel than liquid particles of the same size.

Carbon and Graphite Mold Plugs and Stool Inserts. V. N. Nolan. (Steel, 1944, vol. 115, Dec. 25, pp. 94–96, 120). Data on the properties of carbon (prepared with pitch and baked in a gas-fired furnace) and graphite are presented and the use of carbon inserts in the bottom of ingot moulds is discussed. The thermal conductivity and coefficient of expansion of carbon are lower than those of cast iron. Discs of carbon 3 in. or 4 in. thick were inserted in ingot-mould stools, and the erosion caused by the stream of steel when pouring was much less in the carbon than that in the cast-iron stool. The mould was much more easily stripped from the ingot when there was a carbon insert in the bottom. In one case a stool was machined to take a carbon block 17 × 17 × 4 in. The distance from the ladle nozzle to the stool was 21 ft., and seven ingots were cast before the block was worn too thin to be of further use.

Baling Sheet Scrap. G. W. Birdsall. (Steel, 1944, vol. 115, Dec. 18, pp. 92–93, 142–148). The increase in the quantity of sheet-metal scrap produced at many works has led to improvements in the design of baling presses. Some presses of recent design are described and illustrated.

The Supply of Water for Iron and Steel Works. K. Guthman. (Iron and Steel Institute, 1945, Translation Series, No. 209). An English translation is presented of a paper which appeared in Stahl und Eisen, 1944, vol. 64, May 11, pp. 301–310 (*see* p. 9 A).

Steelworks Lighting. W. R. Stevens. (Iron and Steel, 1945, vol. 18, Feb., pp. 38–44). Data are presented on the relation between lighting and production in different industries, and a standard of lighting suitable for the interior of steelworks buildings is suggested.

FORGING, STAMPING AND DRAWING

(Continued from p. 125 A)

Dodge Chicago Plant Makes Own Forgings for Army B-29 Super-Fortress Plane Engines. (Industrial Heating, 1944, vol. 11, Oct., pp. 1581–1594). A general description is given of the new forging plant at the works of the Chrysler Corporation, Chicago. Steam hammers with capacities ranging from 2000 to 35,000 lb. are installed, and the characteristics and operation of the largest hammers are discussed.

The Pressing of Sheet Metal. A. W. Swift. (Journal of the Junior Institution of Engineers, 1945, vol. 55, Feb., pp. 109–135). A comprehensive review of the theory and practice of pressing sheet metal is presented. Mechanical and hydraulic presses and their tools and equipment are described, and separate sections are devoted to cupping and shallow-drawing, deep-drawing, forming, trimming and finishing, and press capacity.

Deep Drawing Steel. G. E. Stedman. (Steel, 1944, vol. 115, Dec. 7, pp. 90–91, 132–140). An illustrated description is given of the sequence of processes for the mass production of 3-in. dia. steel cartridges by cupping and deep-drawing blanks 7·8 in. in dia. and 0·550 in. thick, which were developed by the Norris Stamping and Manufacturing Co., Los Angeles.

Forging Tube Forms. (Steel, 1945, vol. 116, Jan. 8, pp. 92–96). Descriptions are given of some presses which were altered so that they could be used for shaping the nose and tail on lengths of seamless steel tubing so as to speed up the production of bomb cases.

Roll Forging and Twisting Process Developed for Production of Shankless Twist Drills. (Steel, 1944, vol. 115, Dec. 25, pp. 82–86, 119). A new process of making twist drills of 6/6/2 tungsten-molybdenum-vanadium high-speed steel is described. Mill lengths of the grooved steel are cut into lengths which are a multiple of that of the individual drills required. These are heated in electric furnaces to a forging temperature of about 2000° F. They are then put through a roll-forging machine, which produces the fluted contour at a rate of 700 ft. of bar per min. The bars, while still hot, pass next to the hydraulic twisting machine; with this machine the final helix angle can be controlled to within 0·5°. The next operation is that of pressing the cutting edge to the proper dimensions and straightening. The drills are then cut off to the finished lengths and ground on special flute-grinding machines to remove the decarburised surface on the under side of the cutting edge. Finally they are hardened and tempered.

Gap-Mill Forging. G. W. Birdsall. (Steel, 1944, vol. 115, Dec. 11, pp. 116–119, 166–172). An illustrated description is given of the gap-mill process of forging; in this the forging is done between rolls, each set of rolls being cut with a series of impressions, into which the work is placed in sequence. The rolls are not complete cylinders, as about a quarter of the periphery is cut away to permit entry of the work. The impressions diminish in size down to the desired final shape. The billet or bar is moved by the operator from one pair of impressions to the next smaller size, and is rotated about 90° each time, so that the flash is squeezed into the bar in each roll-forging operation. The part is finally rounded-up and straightened in dies under a steam hammer. The process lends itself to mass production, as about 725 forgings can be made in an 8-hr. shift.

Service Failure of Forging Die-Shanks. J. Vanas. (Steel Processing, 1944, vol. 30, Nov., pp. 718–720). The failure of die-blocks as a result of cracks from fillet to fillet across the die-shank is described. This was not due to heat-treatment. It was prevented by changing the design so that the wings of the die rested on the bolster, leaving a clearance of 0·008 in. between the die-shank and the bottom of the recess in the bolster into which the shank fits.

Alignment Charts for Bending Dies for 18-8 and 17-7. C. M. Brown, W. O. Binder and R. Franks. (Metal Progress, 1944, vol. 46, Dec., pp. 1268–1272). Methods of calculating the "spring-back" when bending stainless-steel sheet were explained in a previous paper (*see* p. 135 A). Five nomographs are now presented which enable the correct die angle for producing any desired bend in the following materials over a wide range of thicknesses to be read off: (1) Annealed 18/8 stainless steel with a tensile strength of 36,000 lb. per sq. in.; (2) cold-rolled 18/8 steel with a yield strength of 115,000 lb. per sq. in.; (3) cold-rolled 18/8 steel with a yield strength of 126,000 lb. per sq. in.; (4) and (5) cold-rolled 17/7 steel with yield strengths of 124,000 and 151,000 lb. per sq. in. respectively.

Reactive Wire Drawing. K. B. Lewis. (Wire Industry, 1945, vol. 12, Feb., pp. 79–81). Formulæ are developed for calculating the stresses and forces involved when drawing wire with an applied back-pull (or "reactive drawing"), and the advantages of the process are pointed out.

PYROMETRY

The Practical Side of Fine Temperature Control. R. Barrington Brock and L. T. Townson. (Chemistry and Industry, 1944, Oct. 14, pp. 358–360). The problems of thermostatic control are dealt with under four main headings: stirring, heating, temperature control and heat losses. Particular reference is made to the control of baths, the temperature of which must be maintained with a high degree of accuracy.

Steelworks Pyrometry. A. Linford. (Iron and Steel, 1944, vol. 17, Oct., pp. 632–636; Nov., pp. 671–675). The theory of radiation is discussed with explanations of Planck's formula and Wien's displacement law, and it is shown how these are applied to the design of total-radiation and optical pyrometers.

Improved Control through Use of Transparent Thermocouple Protecting Tubes. J. P. Vollrath. (Industrial Heating, 1944, vol. 11, Oct., pp. 1619–1620, 1708). Tests are described in which a small electric heat-treatment furnace was maintained at 1200° and then at 1600° F. by an "on-off" control connected to a thermocouple; the object was to determine the rapidity of response when the thermocouple was protected by "Vycor" (a transparent glass), by stainless steel and by a sillimanite tube. The response with Vycor was at about three-quarters of the speed obtained with no protection; stainless steel came next, and sillimanite caused the slowest response.

HEAT TREATMENT

(Continued from pp. 127 A–128 A)

Wartime Developments in the Heat-Treatment of Steel and Their Effect on the Design of Automotive Equipment. H. W. McQuaid. (S.A.E. Journal, 1944, vol. 52, Dec., pp. 598–608). War-time improvements in induction-hardening, flame-hardening, isothermal heat-treatment, quenching, shot-blasting and controlled atmosphere carburising are reviewed. These indicate the possibility of improving the design of many mechanisms on the basis of aircraft engine experience and the necessity for much closer co-operation between the designing engineer, the metallurgical engineer and the production engineer during the initial development of any important design.

How Long is "Holding Time"? W. Trinks. (Industrial Heating 1944, vol. 11, Oct., pp. 1626–1632). It is suggested that the holding times, usually stated in heat-treatment specifications include such a large safety factor that owners of modern and efficient heat-treatment furnaces in which parts are heated uniformly on all sides are unnecessarily penalised, because they are prevented from using their equipment at the maximum rate of throughput.

Cold Working and Forming of Silicon-Manganese Spring Steel. R. G. Sartorius. (Iron Age, 1944, vol. 154, Dec. 14, pp. 50–51). Difficulties experienced in the coiling, shearing and forming of a steel containing carbon 0·57%, manganese 0·72%, silicon 1·89%, nickel 0·20% and chromium 0·11% are described. Three alternative forms of heat-treatment were developed which overcame these troubles. One was to raise the hot-rolled shapes to 130–180° F. in steam-heated chambers; the second was heating to 1100–1300° F. and quenching in water or cooling in air, and the third was a spheroidising treatment consisting of holding at 1325° F. for 3

hr., then at 1450° F. for 15 min., cooling fast to 1325° F., holding at 1325° F. F. for 2 hr., cooling fast to 1275° F., holding at 1275° F. for 1 hr. and cooling in air to room temperature.

Heat-Treatment of Tank-Track Spindles. A. R. Page. (Machinist, 1945, vol. 88, Feb. 10, pp. 281–283; Feb. 17, pp. 287–288). A detailed description is given of the plant and processes used for case-hardening tank-track link spindles. Except for the larger sizes, all the spindles are hardened in salt baths. When cutting lengths off the bar-stock a knob is machined at one end of each piece. A short wire is twisted round each knob, and this enables a large number of spindles to be suspended from a single horizontal bar which is carried by a hoist running on a monorail. The spindles can thus be lowered into the cyanide bath. The organisation of the inspection and testing departments is also described.

Three Million Steel Cartridge Cases Successfully Heat-Treated by Electric Induction Heating. F. M. Arnold. (Steel, 1944, vol. 115, Dec. 25, pp. 102–107, 122). An illustrated description is given of a Tocco induction-heating unit for the heating and quenching of 3-in.-dia. steel cartridge cases.

Fundamental Principles and Applications of Induction Heating. (Sheet Metal Industries, 1945, vol. 21, Jan., pp. 111–115; Feb., pp. 293–297, 310). Continuation of a series of articles (*see* p. 79 A). The application of induction heating for the hardening of razor blades, drill chucks, gears for the automobile industry, tank track pins, gauges, camshafts and crankshafts is described.

Some Needed Precautions When Induction and Flame Hardening. J. O. Almen. (Metal Progress, 1944, vol. 46, Dec., pp. 1263–1267). The residual stresses after carburising, flame-hardening and induction hardening and their effects on the fatigue strength of the parts are pointed out. The net residual stress in the surface of a part hardened by quenching without carburising is equal to the compressive stress produced by the transformation to martensite minus the tensile stress produced by the thermal contraction; net compressive stresses of 30,000–40,000 lb. per sq. in. have been measured. The layer under the surface is not transformed, and is stressed in tension; if the part is subjected to external tensile stresses this second layer is very highly stressed, and may lead to failure, especially where it extends to the surface of the part, as at the end of a hardened zone. The danger of this "second zone fatigue vulnerability" is not confined to gear-teeth and crankshafts, but is present in all highly loaded machine parts where the bending or torsional stresses are high and the hardened layer is relatively shallow. Some inferences on the selection of induction-hardening equipment are drawn from the above observations.

Special Loading Fixtures Add to Furnace Capacity and Operating Economy. P. Goetcheus. (Industrial Heating, 1944, vol. 11, Oct., pp. 1639–1646). A number of plates, grids, suspension brackets and stands of heat-resisting steel for carrying charges in heat-treatment furnaces are described and illustrated.

The Design, Selection and Application of Alloy Pots for Liquid Heat-Treating Media. H. I. Dixon. (Industrial Heating, 1944, vol. 11, Oct., pp. 1648–1658). Recommendations are made on the design and material for salt-bath case-hardening pots. The nickel-chromium-iron alloys suggested have nickel and chromium contents ranging from 3% to 70% and from 10% to 30% respectively.

The Principles and Practice of Lithium Heat-Treating Atmospheres. Part II. Gas Carburising. C. E. Thomas. (Industrial Heating, 1944, vol. 11, Nov., pp. 1810–1818). Continuation of a series of articles (*see* p. 78 A). When mixtures of air and propane are used for carburising, heavy soot deposits occur with 35% of propane; if more air is added to decrease the

soot formation, the rate of carbon penetration decreases sharply. With lithium vapour in the atmosphere a high air/propane ratio can be used without reducing the carburising rate and without trouble from soot. A carburising furnace designed for a lithiated atmosphere is described.

Sub-Zero Treatment Improves Tool Life of High-Speed Steels. T. M. Snyder. (Machinist, 1945, vol. 88, Mar. 3, pp. 91–93). A comparison is made of the service lives of tools of two types of high-speed steel (18/4/1 and 5·5/4/1·5 tungsten-chromium-vanadium high-speed steels, the latter with 3% of molybdenum) which had been subjected to three different forms of heat-treatment. These treatments consisted of: (*a*) Quenching in oil from the hardening temperature to 800° F., cooling in air to 200° F. and double tempering at 1050° F. for 2 hr.; (*b*) quenching in oil and cooling in air as in (*a*), then tempering at 1050° F. for 2 hr., cooling down to −110° F., warming to room temperature and tempering at 1050° F.; and (*c*) quenching in molten salt to 450° F., holding for 4 min., cooling in air and then to −110° F., holding for 1 hr. and tempering at 1050° F. The treatments (*b*) and (*c*) increased the tool life by about 60% as compared with treatment (*a*), and the results with (*b*) were slightly better than those after (*c*).

Sub-Zero Treatment of Steel. H. C. Amtsberg. (Iron Age, 1945, vol. 155, Jan. 11, pp. 50–53: Steel Processing, 1944, vol. 30, Nov., pp. 721–725). The theory and practice of heat-treatment cycles which include treatment at sub-zero temperatures are discussed and recommendations on the cycles to apply to high-carbon steels, low- and high-alloy steels and to high-speed steels are made.

Employing the Corrosion Resistance of Stainless Steel for Surface Hardening. R. C. Cunningham. (Steel, 1944, vol. 115, Dec. 25, pp. 74–75, 116–118). A process of nitriding chromium and chromium-nickel corrosion-resisting steels is described. Ammonia gas is passed through a scrubber and a dehydrating tower into a cracking unit; the cracked gas is then forced through a container held at 700° F., in which the parts to be nitrided are packed; this stage of the process takes 3 hr. In the second stage the ammonia is by-passed away from the cracking unit and put through a chamber, in which it is ionised by a corona discharge; the ionised gases are passed into the container, the temperature of which is raised to 975–1025° F. The duration of this operation depends on the depth of case required. A case depth of 0·010–0·012 in. is obtained in 12–14%-chromium steel in a total processing time of 30 hr. With 18/8 stainless steel the case depth obtained in the same time is 0·006–0·008 in. The resistance of the case to corrosion by high-temperature steam, petrol and fuel oil is as good as that of the untreated steel, but its resistance to attack by halogen acids and concentrated nitric acid is less than that of the steel before the nitriding treatment.

MACHINING

(Continued from p. 109 A)

Basic Mechanics of the Metal-Cutting Process. M. E. Merchant. (Journal of Applied Mechanics, 1944, vol. 11, Sept., pp. A-168–A-175). A mathematical analysis of the geometry and mechanics of machining processes is presented. Two cases are covered, that of a straight-edged cutting tool moving against the work in a direction perpendicular to its cutting edge, and that of a similar tool set with its cutting edge oblique to the direction of relative motion of tool and work.

The Measurement of Surface Finish. H. P. Jost. (Transactions of the Liverpool Engineering Society, 1944, vol. 65, pp. 49–70). Surface finish can

be defined as the science of the profiles of surfaces as formed by the surface deviations. Surface finish metrology deals with the theory of surface finish and surface finish inspection. The terms and units used in surface finish metrology are explained, and some instruments for measuring surface finish are described; the latter include: (1) The Tomlinson surface finish recorder; (2) the Abbott profilometer; (3) the Brush surface analyser; (4) the Taylor Hobson Talysurf; and (5) photomicroscopes and comparison microscopes for quantitative measurements.

The Reclamation of Swarf. (Machinery, 1945, vol. 66, Jan. 18, pp. 57–60). A description is given of the system developed at one of the Rolls-Royce aero-engine factories for collecting aluminium and steel turnings, recovering the oil from them and packing the cleaned and sorted metal for despatch to the remelting works.

PROPERTIES AND TESTS

(Continued from pp. 134 A–138 A)

Load Calibration Equipment for Testing Machines. G. L. Brown and H. T. Loxton. (Journal of the Institution of Engineers, Australia, 1944, vol. 16, Dec., pp. 235–239). Methods used for the load-calibration of all types of testing machines and their suitability and accuracy are discussed. A brief description is given of standard types of equipment, including a new design of loop dynamometer and the Amsler standardising box.

Effect of Length on Tensile Strength. C. Gurney. (Nature, 1945, vol. 155, Mar. 3, pp. 273–274). The results of a large number of tension tests on nominally identical test-pieces is conveniently represented by frequency curves. Equations and curves are developed for deriving the distribution of the strength of rods having lengths equal to a multiple of that of the rods tested.

Symposium on Cohesive Strength. (American Institute of Mining and Metallurgical Engineers, Technical Publication No. 1782: Metals Technology, 1944, vol. 11, Dec.). This symposium on the technical cohesive strength of metals was held at Chicago in October 1943.

A brief summary of the papers comprising the symposium was made by M. Gensamer and, after some introductory remarks by the chairman, the following papers were presented:

> **The Technical Cohesive Strength of Metals in Terms of the Principal Stresses.** D. J. McAdam, jun. The technical cohesive strength is a function of all three principal stresses, and must therefore be represented diagrammatically by a surface in a three-dimensional figure. An attempt is made to delineate the cohesion limit in diagrams based on the author's own experiments as well as on researches by Kuntze, Bridgman and Tuckerman. The cohesion limit depends on the previous mechanical history of the material. The evidence indicates that the difference between the technical cohesive strength of a metal and its resistance to plastic deformation is much less than has been generally supposed; after even moderate plastic deformation the resistance to further plastic deformation is only slightly less than the technical cohesive strength. The additional plastic deformation required to overcome this slight difference varies greatly, however, with the metal and the conditions of stressing.
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> **Flow and Fracture.** P. W. Bridgman. The conditions governing the flow and fracture of metals in tests such as the tensile test are

discussed. The author believes that the longitudinal and radial stresses in a notched tensile specimen depart much more from constancy than McAdam assumes, and that this departure is important, particularly in determining fracture. Fracture starts because of the presence of a hydrostatic tension superimposed on the uniform tension; the general effect of a hydrostatic tension is to make the substance more brittle. The natural strain at which tensile fracture occurs is a linear function of the hydrostatic pressure in the medium in which the experiment is conducted.

Conditions of Fracture of Steel. J. H. Hollomon and C. Zener. *See* p. 134 A.

Some Speculations Regarding the Plastic Flow and Rupture of Metals under Complex Stresses. L. R. Jackson. Equations representing complex three-stress systems and their effect on plastic flow are developed, and the triaxial ratio is defined as the mean value of the three principal stresses divided by the absolute magnitude of the largest principal stress. Methods of calculating the distribution of stresses in complex stress systems are explained and some experimental results showing the effect of different types of complex stress on the flow capacity of various metals are given.

Mechanical Failures in Aero-Engines. W. N. Twelvetrees and K. W. Walpole. (Aircraft Engineering, 1944, vol. 16, Sept., pp. 268–274). Several examples of the mechanical failure of aero-engine parts are described and illustrated. In order of average occurrence the following are the basic factors causing failure under war-time conditions: (1) Operational overloading and abuse; (2) shock loading; (3) enemy action; (4) "hydraulicing" (a coined word used to describe the results of hydraulic pressure in the combustion chamber); (5) defects in design; and (6) defects in material.

Investigation of Graphitization of Piping. H. N. Boetcher. (A.S.T.M. Bulletin, 1944, Oct., pp. 37–40). A preliminary report is presented of investigations of graphitisation which occurred in molybdenum-steel tubes in high-pressure high-temperature steam installations. Sections were cut from tubes after 29,000-hr. service at 900° F. and even higher temperatures. Two samples with a total carbon content of 0·118% were found to have 0·024% and 0·029% of graphite respectively. The steels with the most graphitisation contained relatively large amounts of metallic aluminium (above 0·003%), but not much aluminium oxide. Photomicrographs indicated that graphitisation takes place only in coarsely spheroidised steel.

Magnetic Powder Inspection of Castings. C. L. Frear. (A.S.T.M. Bulletin, 1944, Dec., pp. 12–14). The procedure developed by the United States Bureau of Ships for the magnetic powder testing of steel castings is outlined. The Bureau of Ships has interpreted a very large number of radiographs of castings as well as magnetic powder indications, and has come to the following conclusions: (1) Magnetic powder inspection is specially suitable for surface tears and cracks which are easily identified as such; (2) the correct interpretation of internal defects such as blowholes, sand, slag and shrinkage is difficult and in many cases impossible by magnetic powder inspection, but it is easy to detect such defects when they are severe enough to endanger the life of the casting; (3) magnetic powder inspection is considerably faster than radiographic examination; and (4) the two methods are complementary.

Magnetic Particle Inspection of Forgings. C. J. Boyle. (A.S.T.M. Bulletin, 1944, Dec., pp. 14–15). The results obtained in the magnetic powder inspection of turbine rotor forgings, known to contain defects, using wet and dry powders and different magnetising currents are compared. The following conclusions were drawn: (1) Inspection under a circum-

ferential magnetic field was the most effective method; (2) direct current was more effective than alternating, and the results were equally good with direct current from generators, copper-oxide rectifiers or storage batteries; (3) direct residual currents were slightly less effective than continuous currents; and (4) the wet method was more sensitive than the dry method.

Miscellaneous Applications of the Magnetic Particle Test. E. A. Snader. (A.S.T.M. Bulletin, 1944, Dec., pp. 15–16). The equipment and technique for testing turbine blades by the magnetic powder method are briefly described, with notes on the testing of other parts, such as gear teeth, milling hobs, bolts and studs.

Testing the Quality of Steel by Magnetic Methods. H. Sjövall. (Jernkontorets Annaler, 1944, vol. 128, No. 12, pp. 610–616). (In Swedish). Some magnetic methods of testing ball-bearing steels developed by Svenska Kullagerfabriken are described. They are based on measurements of the induced current when the part or sample is placed in a magnetic field. Several instruments for detecting differences in the hardness of ball races are illustrated, and some circuit diagrams are presented.

Fatigue Testing Apparatus. (Aircraft Engineering, 1944, vol. 16, Nov., pp. 334–335). A number of fatigue-testing machines made by W. and T. Avery, Ltd., are described and illustrated.

Wear Phenomena with Dry Sliding Friction. E. Siebel. (Iron and Steel Institute, 1945, Translation Series, No. 217, Part I.). An English translation is presented of a paper which appeared in Zeitschrift des Vereines deutscher Ingenieure, 1942, vol. 86, Mar. 7, p. 157. The results of wear tests are discussed in which annular specimens of various metals were in sliding friction with a hard steel without lubrication. Three types of wear were noted, namely, metallic wear with and without deformation and oxidic wear.

The Behaviour of Hard Surface Layers under Operational Stress. H. Wiegand. (Iron and Steel Institute, 1945, Translation Series, No. 217, Part II.). The advantages and disadvantages of hard surfaces on steel by chromium-plating, case-hardening and nitriding as compared with those of hardening through to the centre are discussed with special reference to their effect on the fatigue strength. The original paper appeared in Zeitschrift des Vereines deutscher Ingenieure, 1943, vol. 87, Mar. 6, pp. 137–138.

Preferred Orientation—An Asset and a Liability. J. K. Stanley. (Metal Progress, 1944, vol. 46, Dec., pp. 1254–1258). While the high magnetic permeabilities obtainable as a result of the recrystallisation texture obtained by annealing are an asset to the electrical industry, and while some mechanical benefits have been derived from the same phenomenon, preferred orientation is recognised as very harmful in a piece of sheet or strip which is to be deep-drawn or cupped. Curves are presented showing the changes in the magnetic and tensile properties of copper and a low-carbon steel in relation to the direction of rolling.

The Magnetic Properties and the Constitutional Diagram of Iron-Cobalt Alloys. M. Fallot. (Métaux, Corrosion–Usure, 1943, vol. 18, Dec., pp. 214–219). The results of tests on the magnetic properties of a series of iron-cobalt alloys of high purity are presented in tables giving the Curie points and magnetic susceptibilities at different temperatures. A curve is drawn from the data showing the sharp rise in the Curie point which occurs as the cobalt is increased from 76% to 100%.

A Study of Several Kinds of High Strength Plate Steel. G. F. Comstock. (Metal Progress, 1944, vol. 46, Dec., pp. 1248–1253). The results of tensile, impact, hardenability and weldability tests on specimens from laboratory heats of 19 nickel-chromium-manganese-molybdenum steel with and without titanium additions are presented and discussed. In general, additions of up to 0·074% of titanium are quite as beneficial as additions of

the more expensive vanadium for improving the properties of high-strength steel plate.

Attempts to Develop a Structural Steel with a Minimum Tensile Strength of 60 kg/mm². H. Neumeister and H.-J. Wiester. (Stahl und Eisen, 1945, vol. 65, Jan. 4, pp. 1–9; Jan. 18, pp. 36–43; Feb. 1, pp. 68–77). An extensive investigation is described the object of which was to produce an easily weldable structural steel which in the as-rolled or normalised state would have an elastic limit of at least 40 kg. per sq. mm., a tensile strength of 60–70 kg. per sq. mm., good ductility and a higher fatigue strength than steel St 52 (a manganese steel with a tensile strength of 52 kg. per sq. mm.). The effects of alloying steel St 52 with silicon, copper, chromium, molybdenum, vanadium and titanium and of various heat-treatments were examined. A slight increase in the alloying elements in steel St 52 was sufficient to ensure, when the steel was cooled in air, that transformation to pearlite was not complete, but that intermediate structures were formed. The latter caused an increase in the tensile strength; the elastic limit, however, was not increased in the same degree, and it was even lowered in some cases; an explanation of this is suggested. The results of tests indicated that increasing the silicon and manganese did not have such a good effect as additions of chromium, copper or molybdenum, but with the latter elements there was an increase in hardness, so that the desired elastic limit was seldom obtained without a dangerous increase in the tensile strength. By adding 0·1–0·2% of vanadium or titanium it was possible to produce a steel which, after normalising, had the desired tensile properties combined with toughness. These results were confirmed by tests on manganese-vanadium and manganese-titanium steel plates from 10 to 70 mm. thick. The former had the required properties in either the as-rolled or the normalised state, but the latter had to be normalised to ensure success. Both types were readily weldable.

Physical Properties of Some NE Steels. W. C. Stewart and R. E. Wiley. (American Society of Naval Engineers: Iron Age, 1944, vol. 154, Dec. 14, pp. 59–74; Dec. 21, pp. 54–57). *See* p. 87 A.

18-8 Stainless Modified for Formability. W. G. Hubbell. (Iron Age, 1944, vol. 154, Dec. 7, pp. 78–82). An investigation of the formability of 18/8 stainless steel sheet is described; its purpose was to determine the best type of sheet for making aero-engine exhaust manifolds. The amount of free ferrite in the steel is inversely proportional to the nickel content, and the tensile strength is in direct proportion to the amount of free ferrite. The forming properties improve with decreasing ferrite and with increasing nickel content. An 18/10 stainless steel can be more easily worked than a fully annealed 18/8 steel. The creep resistance at 1100° F. and the ductility of columbium-stabilised stainless steel are greater than that of steel stabilised with titanium.

One-Coat Enameling Steel is Developed with Exceptional Cold-Drawing Qualities. (Steel, 1944, vol. 115, Dec. 18, pp. 100, 151). A titanium-bearing steel has been developed which will take a single coat of vitreous enamel in white or other light colour and be free from black specks, pits or blisters, thus obviating the necessity of two coats. The steel has good drawing properties.

METALLOGRAPHY AND CONSTITUTION

(Continued from pp. 139 A–140 A)

Macro-Examination. D. Taylor. (Automobile Engineer, 1945, vol. 35, Feb., pp. 43–50). Information is given on the simple equipment

required for examining steel for defects in the various stages of manufacture and on the interpretation of the observations made by etching and sulphur printing.

Automatic Electronic Exposure Control Provides Uniform X-Ray Exposures. H. D. Moreland. (Steel, 1945, vol. 116, Jan. 8, pp. 98–100, 132). A brief description is given of an automatic photo-electric timer for ensuring correct exposures in X-ray work. X-ray radiation passes through an object, and strikes a fluorescent screen, producing a visible skiagraph. A section of the screen is scanned by a photo-electric cell which measures the light leaving the screen. When sufficient light has left the screen to expose the film for the correct time, the cell operates a relay which breaks the circuit to the X-ray tube.

CORROSION OF IRON AND STEEL

(Continued from p. 141 A)

Special Problems Connected with Pumping Water Containing Acid. J. W. Pirie. (National Association of Colliery Managers: Iron and Coal Trades Review, 1945, vol. 150, Feb. 23, pp. 275–277). Some experiences of the corrosion of pump parts and valves by water containing sulphuric acid at a colliery in Scotland are recounted. Phosphor-bronze pump neck-rings and a 3%-nickel steel shaft were severely attacked, so also was a 16/2 chromium-nickel steel. Stainless steel and an alloy called "Acidure" developed by the Fife Coal Co. gave excellent service.

Protective Painting of Structural Steelwork. J. C. Hudson. (Journal of the Oil & Colour Chemists' Association, 1945, vol. 28, Feb., pp. 27–37). The paper on the protective painting of structural steel by the Protective Coatings Sub-Committee of the Corrosion Committee (*see* Journ. I. and S.I., 1942, No. I., p. 87 P) is reviewed and amplified by more detailed references to the experimental work of the Corrosion Committee of The Iron and Steel Institute.

ANALYSIS

(Continued from pp. 116 A–117 A)

Carbon and Sulphur Determinators. (Steel, 1944, vol. 115, Nov. 27, p. 93). Two newly designed apparatus for the determination of carbon and sulphur in steel, iron, non-ferrous metals, coal and cóke are described. Rapid and accurate results can be obtained with them.

Further Advances in the Determination of Phosphorus in Iron and Steel by Means of the Spekker Photo-Electric Absorptiometer. An "Arsenic-Free" Method. T. S. Harrison. (Journal of the Society of Chemical Industry, 1944, vol. 63, Nov., pp. 350–351). When the more selective mercury-vapour light source is used for the absorptiometric determination of phosphorus in iron and steel by the phosphovanadomolybdate method, the relationship between absorption and phosphorus content is linear. The omission of permanganate as the oxidising agent and the removal of graphite by filtration prior to the development of colour are improvements justified by results. The control of the acid concentration is essential for the reproducibility and stability of the colour.

The Determination of Tin in Steels. A. Robertshaw and G. C. Bromfield. (Analyst, 1944, vol. 69, Nov. pp. 240–343). A volumetric method for the determination of tin in steels for amounts less that 0·5% is described. The only elements likely to interfere in steel are tungsten and vanadium;

tungsten can be dealt with by a simple modification, while the error due to vanadium, when present up to 1%, is too small to be of importance.

Determination of Thickness and Composition of Tin-Lead Alloy Coatings on Steel (Terne Plate). J. W. Price. (Analyst, 1945, vol. 70, Jan., pp. 10–14). A volumetric method for the rapid determination of the thickness and composition of tin-lead alloy coatings on steel is described. It was found that 0·4 g. or more of precipitated antimony in the titration solution led to erratic results. Some other applications of the method are also mentioned, *e.g.*, the analysis of electro-deposited tin-lead alloys on steel and the determination of tin in tin-lead solders.

A Combined Method of Chemical Analysis for Cast Iron, Malleable Iron and Steel. W. B. Sobers. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 913–926). *See* Journ. I. and S.I., 1944, No. II., p. 65 A.

The Determination of Hydrogen and Oxygen in Liquid Steel. W. Geller and Tak-Ho-Sun. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Jan.–Feb., pp. 159–162). A description is given of a new apparatus and procedure for determining the hydrogen and oxygen in liquid steel. The description is mainly concerned with the vacuum apparatus for drawing the liquid steel into a quartz glass pipette and collecting the gases evolved during cooling with complete exclusion of oxygen from the air.

The Determination of Hydrogen in Solid Steel. Tak-Ho-Sun, F. Willems and W. Geller. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, Mar.–Apr., pp. 207–210). The following methods of determining the hydrogen in solid steel are discussed and their reliability compared : (*a*) Ageing the specimens at room temperature ; (*b*) hot-extraction of hydrogen from solid specimens at temperatures in the 400–1100° C. range ; (*c*) hot-extraction of hydrogen in a tube furnace or a graphite spiral furnace at temperatures exceeding the melting point of the steel ; and (*d*) hot-extraction of hydrogen from specimens melted with tin in a quartz tube at 1100–1200° C. Tests made with specimens of the same steel by methods (*b*), (*c*) and (*d*) gave different results, the values increasing with the method in that order. Extraction at 1650° C. in the graphite spiral furnace gave the same results as when the tube furnace was used. Extraction could not be carried out at 1150° C. by melting with tin in the graphite spiral furnace. The tests confirmed the opinion of Bennek and Klotzbach that method (*d*) is the most reliable ; it is, however, time-consuming and is very exacting on the apparatus. Additional multi-stage tests at between 400° and 1100° C. indicated the advisability of extraction at 600° C. instead of at 400° C. to obtain the total hydrogen.

The Effect of Annealing on the Determination of Nitrogen in Alloy Steels and Pig-Irons. C. Holthaus and W. Holtmann. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, May–June, pp. 247–260). Factors affecting the acid-solubility of nitrides in steels were studied. The nitrogen compounds, insoluble in hydrochloric acid, which are present in chromium- and vanadium-bearing steels and pig-irons can be converted into soluble forms by rapid quenching from high temperatures, but tempering at just under the Ac_1 point causes them to revert for the most part to their original insolubility in hydrochloric acid. The insolubility of the nitrogen compounds in high-chromium steels is, however, unaffected by heat-treatment. The total nitrogen content of most steels and pig-irons can be determined by ordinary chemical methods as no nitrogen compounds insoluble in hydrochloric acid are found when iron is alloyed with manganese, silicon, molybdenum and nickel. It is only when the nitrogen content is exceptionally high that some remains undissolved after the heat-treatment. Titanium nitride remains insoluble in hydrochloric acid, and this is unaffected by heat-treating the steel. The nitrogen compounds in aluminium-bearing steels are insoluble

only when the aluminium content exceeds 0·075%, but in such cases they can be converted into soluble compounds by treatment with an alkali, and the determination of nitrogen can then be made in the usual way. The combination of different forms of heat-treatment with methods of determining nitrogen provides a useful technique for the study of nitrogen compounds in iron and steel.

The Solubility in Acids and the Electrolytic Decomposition of the Nitrides Occurring in Alloy and Plain Carbon Steels. H. Kempf and K. Abresch. (Archiv für das Eisenhüttenwesen, 1944, vol. 17, May–June, pp. 261–270). Experiments on the effect of heat-treatment on the determination of nitrogen in steel are described. When plain carbon and alloy steels are tempered at below 700° C. a large proportion of the nitrides are converted to forms insoluble in acid and insoluble during electrolysis. By quenching the steel from a temperature in the 700–1400° C. range, the actual temperature depending on the composition of the steel, these nitrides are converted back to the soluble form.

Nitrogen in Steel. (Iron and Steel, 1945, vol. 18, Feb., p. 60). An improved apparatus for determining nitrogen in steel by the Kempf-Abresch method (*see* p. 68 A) is described.

The Rapid Chemical Determination of Iron, Manganese, Silica and Lime in Basic Open-Hearth Slags to Control the Melting Process. E. Stengel. (Stahl und Eisen, 1944, vol. 64, Dec. 21, pp. 802–806). Recommendations are made on methods of taking samples of open-hearth slag and rapidly preparing them for chemical analysis. Methods of determining the total iron, ferrous oxide, manganese, silica and lime which are sufficiently rapid and accurate to be used for controlling the heat are described.

An Introduction to Metallurgical Spectrographic Analysis. D. M. Smith. (Institute of British Foundrymen : Foundry Trade Journal, 1945, vol. 75, Jan. 18, pp. 55–56). A brief outline of the principles employed in metallurgical spectrographic analysis is given.

The Uses and Limitations of the Spectrograph. J. F. McNeil. (Australasian Engineer, Science Sheet, 1944, Dec. 7, pp. 10–14). A spectrograph is described and its principles are explained. Its merits and limitations for carrying out analyses are discussed and compared with those of wet methods.

Spectrographic Equipment for Foundry Metal Control. H. W. Dietert and J. A. Schuh. (Transactions of the American Foundrymen's Association, 1945, vol. 52, Mar., pp. 889–902). The application of the spectrograph for rapid analyses in foundries is discussed and descriptions are given of suitable equipment, including such items as moulds for casting the pin and disc types of sample, developing machines, film dryers and means for converting line density readings into the percentage of the element present.

Spectrograph Speeds Analysis Work in New Allis-Chalmers Laboratory. G. W. Birdsall. (Steel, 1944, vol. 115, Dec. 25, pp. 73, 112–114). A description is given of the new spectrographic laboratory of the Allis-Chalmers Manufacturing Company.

The Order of Stating Elements in Analyses. (Jernkontorets Annaler, 1944, vol. 128, No. 12, p. 621). (In Swedish). The Standards Committee of Jernkontoret has recommended that when steel analyses are given in technical journals, certificates and correspondence, the order in which the elements are given should be standardised as follows : C, Si, Mn, P, S, Cr, Ni, Mo, W, Co, V, Ti, Cu, Al, Pb, As, Sb, Sn, Nb, Ta, Zr and B, followed by the gases oxygen, hydrogen and nitrogen.

An Electrically Heated Furnace for the Determination of Volatile Matter in Coal. C. W. G. Ockelford. (Fuel in Science and Practice, 1945, vol. 24, Jan.–Feb., pp. 3–7). The constructional details are given of a small electric furnace for use in determining the volatile matter in coal by the method described in British Standard Specification 1016 of 1942.

SUBJECT INDEX.

[References to the papers read before the Institute are indicated by the word *Paper* following the page number. The letter *P.* denotes a reference in Section I. of the Journal which contains the reports of the proceedings of meetings, the papers read, and the discussions thereon. The letter *A.* denotes a reference to the section dealing with abstracts.

Indexing of Alloy Steels and Other Alloys. In the indexing of alloy steels, carbon and iron are ignored and the alloying elements contained in the steel are arranged in alphabetical order; for example, all references to nickel-chromium-molybdenum steel will be found under the heading chromium-molybdenum-nickel steel. In the indexing of other alloys, carbon and iron, when present, are included in the title; iron, when present, is always mentioned first and the other elements follow in alphabetical order, carbon being in all cases mentioned last. Examples: "Iron-silicon-carbon alloys" and "iron-chromium-nickel-carbon alloys."]

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